AN EXPERIMENTAL AND NUMERICAL ANALYSIS OF HYDROGEN ASSISTED CRACKING AND WELDABILITY TEST METHODOLOGY

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TO MY PARENTS

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ABSTRACT

An Experimental and Numerical Analysis of Hydrogen Assisted Cracking and Weldability Test Methodology

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Hydrogen assisted cracking (HAC) in weld metal and/or heat affected zone (HAZ) is caused by a combination of tensile stress, diffusible hydrogen, and a susceptible microstructure. Hydrogen assisted cracking induced catastrophic failures in bridges, ships, offshore drilling rigs, pipelines, aircrafts, structural fabrication and, more recently in Sea Wolf submarines have been well researched. In spite of extensive research, complete immunity against HAC cannot be guaranteed because of the complex, and incompletely understood, microscopic interactions between metallurgical, environmental and mechanical factors.

The preferred method for increasing resistance to HAC is the application of an adequate preheating temperature, T_{ph} . The suitability of given welding conditions, including T_{ph} , in avoiding HAC is generally assessed through the use of Tekken and Lehigh weldability restraint tests. The safe welding conditions determined from these tests are then applied in industrial fabrication. It is observed that these safe welding conditions do not always avoid HAC in actual weldments. Therefore, it is necessary to evaluate the weldability testing procedure in its entirety against the more general industrial fabrication practice and understand the inherent differences.

The differences arising, at different stages of weldability testing procedure, from weld hydrogen measurement technique, weldability testing procedure, hydrogen diffusion behavior, residual stress development, and, dimensional differences in weldability tests and actual weldments were analyzed in detail using an experimental and numerical approach.

The weld hydrogen measurement results indicated that the existing hydrogen measurement standards do not measure the weld hydrogen levels in actual weldments, and should, therefore, be modified for use in weldability testing procedure. The Tekken and Lehigh weldability test results suggested that weld induced variation at stress concentration locations strongly influences the HAC tendency and crack propagation behavior.

Finite element analysis (FEA) of hydrogen diffusion behavior in weldability tests and actual weld grooves indicated that hydrogen diffusion is a strong function of the groove shape and the weld thermal cycle, and hence, direct applicability of weldability test results to actual weldments can be misleading. Elasto-plastic thermo-mechanical behavior of Tekken and Lehigh weldability tests during welding was carried out using FEA. The heat transfer in these test specimens was highly non uniform resulting in varying HAC susceptibility along the weld length. The residual stress distribution varied significantly along the weld length due to the slot nature of the groove and restraint variation along the weld length.

It was observed from this research work that the existing weldability testing methodology needs to be modified for better applicability to industrial fabrication practice. The weldability tests should be designed so as to simulate weld hydrogen levels, welding process variables, hydrogen diffusion, heat transfer and residual stress levels of actual weldments. It is expected that incorporating these suggestions will lead to a better control over HAC occurrence and prevention.

INTRODUCTION

Successful welding in industrial fabrication requires, among other things, that weldments be resistant to hydrogen assisted cracking (HAC). Hydrogen assisted cracking, or cold cracking, in weld metal and/or heat affected zone (HAZ) is caused by a combination of tensile stress, diffusible hydrogen, and a susceptible microstructure. Hydrogen assisted cracking generally occurs after the weldment has cooled to below about 150°C, and is especially dangerous because of an incubation period of up to several hours or even days after welding.

Hydrogen assisted cracking has caused catastrophic failures in bridges, ships, offshore drilling rigs, pipelines, aircrafts, structural fabrication and, more recently in Sea Wolf submarines. The economical impact of HAC can be understood from the fact that, in Britain alone, the estimated annual cost of repairing hydrogen assisted cracks runs into over 40 million pounds¹. Service failures of welded components cost an additional 140 million pounds annually¹. As a result of its critical effect on the life of welded structures, HAC has received tremendous attention from various researchers. In spite of extensive research, complete immunity against HAC can not be guaranteed because of the complex, and incompletely understood, microscopic interactions between metallurgical, environmental and mechanical factors. Therefore, HAC is avoided through more "tangible" or practical measures.

In industrial fabrication, HAC is controlled through the use of low hydrogen welding practices, less susceptible materials and modification of welding procedures. The preferred method for increasing resistance to HAC is the application of an adequate preheating temperature, T_{ph}, as it lowers the susceptibility of a given microstructure and also the residual diffusible hydrogen. Correct application of preheating requires detailed knowledge of a variety of weld metal and base metal metallurgical and geometrical constraint variables and their interaction as a function of welding parameters and ambient temperature condition. The various relationships needed to correctly determine a given weldments preheating temperature are still not completely defined and/or quantified. Thus practical fabrication preheating temperatures must be estimated either experimentally and/or analytically.

The suitability of given welding conditions, including T_{ph} , in avoiding HAC is generally assessed through the use of weldability tests. The "safe" welding conditions determined from these tests are then applied in actual weldments. It is observed that these "safe" welding conditions do not always avoid HAC in actual weldments. Because of this discrepancy in the laboratory weldability test results and actual fabrication experience, it is extremely difficult to rely on laboratory test results. As a result, the selection of "safe" welding conditions becomes a difficult task and, more importantly, the results can not be relied upon. Therefore, it is necessary to evaluate the weldability testing procedure in its entirety against the more general industrial fabrication practice and understand the inherent differences.

These differences primarily arise, at different stages of weldability testing procedure, because of; 1) existing weld hydrogen measurement methodology, 2) weldability testing procedure, 3) hydrogen diffusion behavior, 4) residual stress development, and, 5) dimensional differences in weldability tests and actual weldments,

It was felt that with a better understanding of these differences, the weldability test results could be applied more closely to actual weldments. The advantages of such a study would be enormous, as better control over the incidence of cracking in industrial fabrication can be exercised. Secondly, it will help design the weldability tests to closely simulate actual welding practices. More importantly, the uncertainty factor due to the delayed nature of the test could be minimized.

The objective of this research work was to experimentally and numerically analyze the weldability testing procedure, with an emphasis on the following:

- Detailed experimental analysis of existing weld hydrogen measurement procedure and its suitability from a weldability testing viewpoint.
- (2) Experimental investigations into self restraint weldability tests, such as, Lehigh and Tekken, and their suitability to HAC susceptibility assessment. A detailed analysis of different preheat temperature prediction methods and their applicability to actual weldments was performed.
- (3) Numerical study of hydrogen diffusion behavior of the weldability tests and different weld joints used in practice.
- (4) Numerical investigation of thermal distribution and residual stress development in weldability tests and a qualitative understanding of the consequences of dimensional differences of actual weldments.

Thesis Format

The thesis is organized in 4 Chapters, each separately focusing on the topics mentioned above. In each Chapter, existing literature is reviewed and the problem is defined, the experimental or the numerical approach is detailed followed by results and discussion. The first Chapter addresses different issues related to weld hydrogen measurement procedure. Weldability testing and several other critical issues are analyzed in Chapter 2. Numerical simulation of hydrogen diffusion behavior in different weld joints is detailed in Chapter 3. Using the finite element method, thermo-mechanical behavior of weldability tests during welding is examined in the last Chapter. In the end, important findings of this research work related are discussed.

CHAPTER 1

WELD HYDROGEN MEASUREMENT STUDIES

1.1 - BACKGROUND

Weld hydrogen measurement provides the means of deciding the amount of weld hydrogen introduced from given welding consumables. This assists in categorization of welding consumable and process combinations. These measurements also provide a starting point for preheating temperature, T_{ph}, and outgassing heat treatment determinations to remove hydrogen after welding. Good mobility of hydrogen at room temperatures and the concentrations, of the order of parts per million level, require special sampling and analysis procedures for weld hydrogen determinations². Standardization of weld hydrogen determination procedure is thus necessary for inter-laboratory comparisons and to develop procedures controlling weld hydrogen levels. American Welding Society, AWS A4.3-86³, International Institute of welding (IIW) testing standard, IIW Doc. No. II-1155-91 (to replace ISO 3690)², British Standard, BS 6693⁴, and Japanese Industrial Standard (JIS), Z 3118⁵, are some of the currently used standards for weld hydrogen determinations. Table 1 lists specimen dimensions and outgassing treatments for some of the most commonly used standards.

The primary focus of this research work was to experimentally and analytically assess weld hydrogen measurement methodologies in the context of understanding/predicting hydrogen assisted cracking (HAC) susceptibility through weldability tests. The weldability testing procedure involves test welding of weldability specimens, such as, Lehigh and Tekken, at preselected welding parameters. The objective of the weldability testing procedure is either to define the welding parameters or the T_{ph} (for a given set of welding parameters) that will avoid HAC. In the latter case, different preheating prediction schemes are used which base their predictions on weld hydrogen content, base metal chemistry and welding conditions (heat input). Prediction of safe T_{ph} is strongly dependent on knowledge of weld hydrogen content. This dependency arises because the cracking index (tendency) for a weld is a logarithmic function of weld hydrogen content⁶. The weld hydrogen content for a given welding procedure is determined by using a suitable standard test methodology.

It is important to note here the basic purpose of these standards. These standards were developed to measure weld hydrogen levels for a given combination of welding consumables and processes. Classification of welding consumables and processes according to their weld hydrogen potential was the underlying principle behind these standards. Subsequent to this initial development the same standards for weld hydrogen measurement were applied for weldability testing. From a weldability testing viewpoint, several crucial issues are not properly attended to in the weld hydrogen measurement standards. Some of these issues are :

- Specimen preparation influence on weld hydrogen levels. The weld hydrogen measurement standards recommend a variety of surface preparation techniques, such as, dry belt sanding, machining, dry shot blasting, surface grinding etc. Measured weld hydrogen levels depend on specimen preparation technique. Weldability test specimens are generally machined.
- 2 Prior degassing of the weld hydrogen measurement specimens to remove any residual hydrogen from the steel making operation. In the weldability testing procedure, the specimens are not subjected to any degassing treatment.
- Base metal chemistry difference effect on hydrogen absorption and hydrogen diffusivity. AWS A4.3-86 recommends that ASTM A36 or SAE 1020 be used as base metal, whereas, in weldability testing, the base metal is determined by fabrication requirement.
- 4 Joint and groove geometry effect on weld hydrogen levels. Weld hydrogen measurement standards use bead-on-plate (BOP) type specimens. In weldability tests, however, groove shape is highly variable and rarely it is of the BOP type.
- 5 Specimen lengths in weld hydrogen measurement standard specimens and weldability specimens are significantly different and may influence measured weld hydrogen levels.

It is clear from the above discussion that the differences between the weld hydrogen standards and the weldability testing procedure can influence hydrogen absorption and diffusion processes. If their effect on measured weld hydrogen levels is significant, then it also suggests that weldability test results are often analyzed based on wrong weld hydrogen levels. The effect of these differences on weld hydrogen levels has been overlooked by welding researchers. The significance/insignificance of these differences needs to be evaluated for a better understanding of the suitability of weld hydrogen measurement techniques, from a weldability assessment viewpoint.

In this chapter, a systematic review of the existing literature on weld hydrogen measurement procedure was carried out. It focussed on the different weld hydrogen measurement techniques, their drawbacks and limitations, the best technique available and the effect of different variables on weld hydrogen measurement.

WELD HYDROGEN MEASUREMENT TECHNIQUES

Weld hydrogen measurement involves welding a test assembly consisting of run on and run-off tab, and a test specimen at preselected welding parameters. This is immediately followed by quenching in iced water to retain as much of initially absorbed hydrogen as possible. The test assembly is then stored in a low temperature bath for further processing before commencing the hydrogen evolution (outgassing) treatment. Hydrogen evolution is carried at a predetermined outgassing temperature. Depending on the standard and the method used, evolved hydrogen is collected over a mercury or glycerine bath or in air-tight capsules in an argon atmosphere when hydrogen measurement is performed by using the gas chromatograph (GC) method.

The Glycerine Method

The non-toxic nature and cost of glycerin, as well as the simplicity of this method, prompted earlier weld hydrogen measurements to be performed exclusively by the glycerin method. Detailed analysis of hydrogen measurement using glycerine as a collecting medium by various researchers^{7,8} has proved beyond doubt that results from the glycerine method are conservative and inaccurate. Additionally, reproducibility can not be guaranteed due to water content, hydrogen solubility⁸, and suspension of small-size hydrogen bubbles⁷ in glycerin.

Ohno and Uda⁷ examined collecting mediums for weld hydrogen determination and found that measured hydrogen content was a function of the rising rate of evolved hydrogen bubbles through the various mediums. Their rigorous experimental and analytical analysis showed that rising rate of gas bubbles in glycerin was the slowest of all mediums studied and resulted in suspension of small-sized bubbles in glycerin. In comparison, the rising rate of bubbles in mercury was 10⁴ times greater than that in glycerin with the result of essentially no bubbles retained in mercury. At present, only the JIS standard allows use of the glycerin method.

The Mercury Method

Weld hydrogen measurements using mercury as the collecting medium have been scrutinized in depth. Bonsizewski and Morris⁹ experimentally found that the mercury method (ISO 3690) is unreliable for hydrogen measurement. They further pointed out that the evolution rate of hydrogen is surface physico-chemical condition dependent and observed that different amounts of hydrogen were collected over a period of three days, the suggested standard time. Morris¹⁰ concluded that the three-day period is misleading as hydrogen evolution is not complete in this time period at ambient temperature; further proposing that evolved hydrogen should be measured regularly until there is no change in consecutive measurements. This suggestion has been subsequently adopted in IIW and BS 6693 drafts. According to Matsada, et al¹¹, the inaccuracies in weld hydrogen measurements by mercury method are caused by individual errors in reading and the occurrence of small bubbles.

Gedeon¹² observed that scatter in weld hydrogen was higher with the IIW specimen compared to the AWS specimen when using gas metal arc welding (GMAW) with globular transfer. Coe¹³ discussed the chronological development of the IIW method and principles and justifications behind it. He also outlined the development of a new IIW draft and defended the mercury method against the criticism that: 1) mercury is an undesirable collecting medium because of the health hazard; 2) the method is slow, inaccurate and imprecise; 3) modern methods are faster and more accurate. He provided experimental evidence with the following conclusions: 1) The standard deviation is 0.28 at 8.00 ml/100 g level for triplicate measurements, 2) No other gases were measured after evolution over mercury, 3) Reproducibility at 3 days is poorer as hydrogen evolution is not complete and surface variations contribute to the scatter. The reproducibility improved after a 14 day evolution period as hydrogen evolution was almost complete. According to Coe¹³, the IIW mercury method should serve as a primary method (21 days evolution at room temperature) against which all other methods should be calibrated.

The Gas Chromatograph Method

For weld hydrogen measurement with higher precision and accuracy, Gas Chromatograph (GC) method was proposed by various researchers. Application of gas chromatography to weld hydrogen measurement was first reported by Sugiyama, et al¹⁴. Pokhodnya and Paltesvich¹⁵ reported excellent agreement between GC method and ISO 3690 mercury method. JIS provided weld hydrogen measurement standard for using GC method for GMAW in 1983. This was later adopted into MIL Spec E 022200/10A(SH) for the high tensile strength electrodes for HY80 and HY100 steels using the IIW specimen. Weld hydrogen measurements using GC found its way into the AWS A4.3-86 and in IIW standard [II-1155-1991] draft for weld hydrogen measurement. High reproducibility, rapid and accurate measurement after complete evolution (it takes about 3 minutes per specimen for measurement) and ease of operation make GC an ideal tool for rapid quality control for weld hydrogen measurement. In this research work, it was, therefore, decided to perform weld hydrogen measurements using the GC method.

Gas Chromatography Principle

Gas chromatography refers to separation of a mixture of compounds into its components. The chromatographic process is the partitioning of a compound between two phases, a stationary phase and a mobile phase. The amount of compound in each phase depends on its relative affinity for the two phases. The proportion to which a compound distributes itself between two phases, termed the partition coefficient, is different for different compounds.

Gas Chromatograph Method Details

In the Yanaco-Oerlikon GC used in this experimental study, the stationary phase is a molecular sieve of type 5A that allows the separation of hydrogen and other gases. This stationary phase (molecular sieve 5A in fine granular form) is packed in a column of 5mm inner diameter and 3 meters long. The mobile phase (carrier gas) is ultra high purity argon. The mobile phase moves the compounds down the column. When the diffusible weld hydrogen evolution is complete the evolved gas mixture is passed through the column. Due to the different rates of migration through the stationary phase hydrogen, nitrogen and the mobile phase (argon) are separated from the evolved gas mixture and are detected. The detector in the GC unit is a thermistor type thermal conductivity detector.

The response of the detector is a linear function of the concentration of the solute (hydrogen) in the carrier gas or the mobile phase (argon) if the flow rate is constant. At high temperatures, the sensitivity of the detector goes down and the response is no longer proportional to the concentration of the solute in the carrier gas. The stability of the detector depends on the temperature of the thermistor, and it varies with carrier gas flow rate. The electrical supply must be stabilized so that it provides constant intensity current to the bridge.

It should be noted, however, that the results are highly sensitive to carrier gas flow rate and temperature fluctuations during measurement. Therefore, calibration before each new measurement is a must. The calibration procedure can take anywhere from 30 minutes to 2 hours.
Furthermore, operator to operator variation can become significant. Many times the manufacturers instructions do not provide a complete understanding of the weld hydrogen measurement procedure and the principles underlying it. Therefore, according to the new IIW standard draft (II-1155-91), the results from the GC method should be calibrated against the primary method, i.e. the mercury method.

COMPARISON OF DIFFERENT WELD HYDROGEN MEASUREMENT TECHNIQUES

Symons¹⁶ reviewed the development of different hydrogen measurement standards and the basic principles related to diffusible hydrogen measurements. According to Symons, a comparative study between IIW and AWS specimens using the GC for hydrogen measurement after evolution at 45°C (72 h) and 150°C (6 h) against the primary IIW/BS 6693 mercury method over a 21 day period was performed by Evans. Evans suggested that the mercury method should be the control method with the GC method providing a rapid quality control tool. Hydrogen measurements according to AWS GC method exhibited slightly lower hydrogen values compared with the primary method at all weld hydrogen levels studied. This, in spite of the fact that the AWS specimen is considerably longer than the IIW/BS 6693 specimens.

Kotecki¹⁷, suggested that the weld hydrogen values using the AWS method are 93% of the IIW method. The AWS A4.3-86 also mentions this type of correlation. Jenkins¹⁸ at The Welding Institute (TWI) compared AWS and other methods. He found that, the AWS specimens (after outgassing at 45°C and 150°C and measurement by GC) exhibited higher hydrogen than the Ytube mercury method using the 30x15x10 mm test block according to the ISO 3690 method at room temperature. When compared with the TWI carrier gas method at 300°C, no difference was found between this method and the AWS method. However, it is interesting to note that the Ytube method (ISO 3690) did not show statistically different results than the TWI carrier gas method at 300°C.

According to Evans¹⁹, the IIW specimen with GC at 45°C and 150°C showed 98.6 % of

hydrogen compared to that measured by the mercury method at room temperature (RT). The AWS method of using GC at 45°C and 150°C exhibited 95.2% hydrogen of the GC-IIW method. This suggests that specimen length is contributing to lower hydrogen levels in the AWS specimen. Effectively, AWS-GC (45 and 150°C) shows about 93.8% hydrogen of that measured by the mercury method at RT. From the above discussion, it can be concluded that: 1) IIW mercury method is the most reliable method and should be used as a primary method as it does not depend on a secondary calibration, 2) The GC technique is equally reliable and can be used as a rapid (depending on the outgassing treatment) measurement tool.

In the present study, the GC method was used. Most of the measurements were carried out at 150°C and 45°C for speedier evolution. The mercury method was not used due to its potential health hazard.

INFLUENCE OF VARIABLES ON WELD HYDROGEN CONTENT

Base Metal Composition Effect

The AWS A4.3-86 recommends use of nonrimming quality of steel grade ASTM A36 or SAE 1020. The IIW/ISO 3690 recommends plain carbon non-rimming steel containing not more than .3% C and .045% S for weld hydrogen measurements. Thus base metal chemistry effect on hydrogen absorption in the weld pool is not addressed in these standards. Though a first guess suggests that the arc atmosphere will be different with different steel chemistries and may influence hydrogen absorption and solubility. In addition, the hydrogen trapping effect is dependent on microstructure.

Gedeon¹² experimentally found that the diffusible hydrogen content of high strength steel (MIL-A-46100) was consistently 2 to 3 ppm higher than the A36 mild steel. It should be pointed, however, that these differences were not analyzed statistically even though the scatter in the weld hydrogen level was considerable.

The chemistry effect becomes significant from a weldability testing viewpoint. As reviewed above, the base metal chemistry influences hydrogen absorption during welding. When assessing A514 steel weldability based on the weld hydrogen content determined from A36 base metal it should be noted that the initial hydrogen content will be different in these two steels. When using preheat prediction schemes with the weld hydrogen content determined using A36 specimens, preheat temperatures corresponding to the A36 weld hydrogen content will be calculated. This preheat temperature may or may not avoid HAC in high strength steel (say A514) as the weld hydrogen content is higher than that determined by using A36 steel. This also will determine preheat/no-preheat scenario for many steels. In this author's opinion, this could one of the major reasons for incorrect preheat temperature determination using prediction schemes.

Different base metal chemistry and prior heat treatments of the base metal also affects microstructure. It is well known that hydrogen diffusion coefficient and solubility change with microstructure²⁰. The results by Asaoka²⁰ show that the solubility of hydrogen in a tempered structure with fine precipitates is the same level of, or in some cases higher than, that in the quenched martensitic structure, contrary to the effect on diffusion coefficients. Asaoka, ²⁰ using autoradiography, observed no trapping effect from alloying element additions but postulated that, since the precipitate-matrix interface and prior-austenite grain boundary characteristics are changed by alloying elements, the hydrogen trapping is influenced by alloying element addition. Thus the amount of measured weld hydrogen is a function of diffusion coefficient and solubility dependence on microstructure. Therefore, from weldability viewpoint, the same base material should be used for weld hydrogen measurement as will be used in weldability tests.

Outgassing Treatment Effect

Hydrogen introduced during welding is present in two states: trapped (residual) and diffusible hydrogen. Hydrogen is dispersed in three different regions of the metal lattice: interstitial lattice sites, vacancies and dislocations, cavities and cavity interfaces. Lattice imperfections; such as dislocations and pores are called traps. As hydrogen is captured by these imperfections it is bonded with particular values of binding energy depending on the type of trap. Trapped hydrogen is released after a critical temperature is reached. This suggests that the amount of trapped hydrogen release is a function of outgassing temperature.

In high-strength steels, the hydrogen dissolved in the metal lattice has an embrittling effect. The diffusible hydrogen is, hence, responsible for cracking. Whereas, the residual trapped hydrogen has no embrittling effect²¹. During the subsequent heat treatment operation, the residual (trapped) hydrogen can become diffusible again^{21, 22} and contribute to cracking or it may outgass from the steel. As the outgassing temperature increases, the amount of hydrogen evolved also increases because of the release of trapped hydrogen. The trapped hydrogen evolution rate exhibits a maximum at certain temperatures. For example, Lee, et al²³, Table 2, found distinct material dependent peaks for trapped hydrogen evolution rate as a function of temperature. They observed that the height and location of these peaks, and hence the binding energies, varied with the amount of cold work and the amount and type of ferrite-carbide interfaces. They attributed temperatures of 115°C and 129°C for hydrogen activation from Fe-C interface traps and temperatures above 200°C for dislocation traps. These reactivation temperatures are different from those found by other researchers, as discussed below.

Otsubo, et al²⁴, Table 2, experimentally found no significant difference in the total amount of evolved diffusible hydrogen at 25, 75, 100, 125 and 150°C but the evolution rates and evolution periods varied widely with temperature. Hydrogen was introduced by charging in hydrogen atmosphere at 950°C in their study. No residual (trapped) hydrogen evolution was observed until the specimen was heated to 250°C and they observed residual hydrogen evolution at temperatures of 310, 380, 480 and 700°C. The peak at 310°C was associated with the ferrite grain boundary. The 380°C peak was attributed to micro-voids caused by cold work. The 480°C peak was related with cementite and/or the microstructure, and the 700°C peak was associated with the interface between the oxide inclusion and the matrix. Habashi, et al²⁵, Table 2, calculated the binding energy for hydrogen introduced through electrolysis at 300°C. They performed hydrogen outgassing at 20, 250, 300 and 400°C to calculate the binding energies for different traps using Oriani's theory²⁶. For a temperature range between 20 - 150°C, only one sort of trap is associated with $\Delta E_x = 53$ kJ/mole of hydrogen and is the dislocation interaction. Between 300 to 400°C grain boundaries or carbide-grain boundary type of trap become active and has a binding energy of 75 kJ/mole of hydrogen.

Recently, Dilthey and Trube ²¹ studied hydrogen effusion behavior from single-pass and multiple pass welds at temperatures between 100°C and 600°C for basic and cellulose coated electrodes. The effusion behavior for a single-pass weld showed dependency on the electrode type. For basic electrode deposited weld metal, they observed a maximum effusion rate of 0.01 ppm/s at 250°C which rose with increasing effusion temperature. At 600°C, a secondary peak, in addition to the primary peak, was observed, and after a further delay a third maximum was also observed. The total amount of evolved hydrogen doubled as the effusion temperature increased from 250°C to 600°C. A similar trend was observed with the rutile electrode deposited weld metal although the amount of evolved hydrogen was 10 times higher than that for the basic electrode. For cellulosic electrode deposited weld metal the effusion trend was similar to that of rutile electrode.

Coe¹³ has strongly admonished against the use of extraction temperatures above 100°C if diffusible hydrogen is to be measured. At higher temperatures residual hydrogen will be released, the amount will be smaller for basic electrode weld metal but can not be guaranteed in other weld metal compositions. Based on their earlier work, Coe²² indicates that, for the IIW specimen, complete evolution at 650°C will require about 1 hour, while at 20°C some 6 days are required. Complete evolution within a working day requires 7 hours at 150°C.

The AWS A 4.3-86 recommends the use of 45 and 150°C outgassing temperatures. The effect of these outgassing temperatures on the amount of reactivated residual and diffusible hydrogen is important as this has a direct bearing on the cracking tendency. According to

Evans¹⁹, the GC results using BS 6693/IIW method showed somewhat higher, though inconsistent, weld diffusible hydrogen at 150°C compared to 45°C, however, no statistical analysis was performed to quantify the difference.

From the above discussion, the temperatures at which trapped hydrogen becomes active are not clear. General consensus on trap activation temperatures can not be found because hydrogen trapping itself is dependent on material, heat treatment etc. The results from trapping studies (external hydrogen charging) can not be directly applied to trapping encountered in welding. During hydrogenation by electrolysis or high temperature charging, some defects are introduced and these may finally affect the trapping density and trap types. Whereas, during welding, the hydrogen introduced does not create additional traps but interacts with the traps introduced during welding due to thermo-mechanical effects. Therefore, residual hydrogen studies involving welded specimens should be relied on. The fact that welding researchers did not find any consistent difference in the amount of hydrogen measured at RT and 45°C or RT and 150°C probably means that there is no release of trapped hydrogen until a temperature of 150° C is reached, or the amount of hydrogen released from traps is so small that it is unrecognizable. Additionally, the variation in welding process and measurement procedure (though small) may affect detection of the trapped hydrogen since the amount of trapped hydrogen is very low. Therefore, same set of specimens should be subjected to different outgassing treatments; starting with complete evolution at lowest temperature and successively increasing the outgassing temperature to study any additional release of trapped hydrogen as a function of outgassing temperature.

Prior Degassing Effect

According to the AWS A4.3-86 standard procedure, the test assemblies are de-gassed at 600°C for an hour to remove any hydrogen in the as-received material introduced during the steel making operation. This residual (trapped) hydrogen becomes mobile when heated to higher temperatures. For example, Lee²³ found activation temperatures of 115, 129, 200, 311 and 405°C

corresponding to the release of hydrogen from different traps, Table 2. Otsubo et al²⁴, found trapped hydrogen evolution at 280, 310, 380, 480 and 700°C corresponding to different types of traps, Table 2. Habashi's ²⁵ results show that trapped hydrogen evolution in the temperature range of 300- 400°C, Table 2. According to Dilthey and Trube²¹ residual hydrogen evolution from single-pass welds was observed at 250°C, 400°C and at 600°C depending on the type of electrode used. Coe and Chano²² observed residual hydrogen evolution from welds which depended on quenching temperature, the total amount of hydrogen in the weld and, weld and base metal composition.

Clearly, any residual hydrogen in the as-received plate has a good chance to become mobile during the test welding because of the high temperatures experienced by the test specimen. When determining weld hydrogen content for use in the determination of T_{ph} , the following issue has to be considered. The weldability specimens, e.g. Lehigh and Tekken, are not prior de-gassed. Therefore, any residual hydrogen present in the steel under consideration will become mobile during welding. From the discussion in the above paragraph, this is especially true for regions that are heated above 200°C. The amount of residual hydrogen will depend on types and kinds of traps, which are determined by the steel making practice, the type of steel plate quality and the thermo-mechanical history. Therefore, the amount of hydrogen in a weldability specimen will be higher than that determined using a prior degassed weld hydrogen measurement specimen. A discrepancy of even 1 ppm can mean the difference between cracking and no-cracking. Hence, the weld hydrogen levels determined using a prior de-gassed specimen may not give a correct idea about the diffusible hydrogen from a weldability testing point of view. Therefore, it is necessary to determine the residual diffusible hydrogen, H_r, content of A36 and A514 steel specimens used in this study in unwelded un-degassed (as received condition) and degassed ($650^{\circ}C$) conditions.

Specimen Preparation Effect

The presence of an oxide film on the specimen surface: Prevents the release of part of the original hydrogen, Reduces the rate of hydrogen release, Reduces the amount of hydrogen recovered in high temperature analysis through a reaction to form water²⁷.

In addition, the stress state at the surface influences hydrogen evolution as well as the amount of hydrogen due to the associated trapping effect. The IIW and AWS standards do not specify a single method of surface preparation, but suggest a variety of methods. The governing idea is to remove the mill scale that can interfere with hydrogen absorption and evolution. The AWS A4.3-86 recommends machining, dry shot blasting and dry belt sanding, the IIW (II-1155-91) draft recommends surface grinding. Boniszewski and Morris⁹ found that the amount of evolved hydrogen with shot blasted specimens was about 10% higher than with wire brushed specimens alone. Coe¹³ found that the hydrogen evolution rate is lowered owing to the surface effects.

Jenkins¹⁸ compared diffusible as well as residual hydrogen from cleaned and as welded specimen using the AWS and ISO mercury method. For diffusible hydrogen measurement by AWS (GC at 150°C) and ISO specimen (TWI Carrier Gas at 300°C), cleaned and uncleaned specimens did not show any significant difference (at a 5% significance level). As the hydrogen extraction temperature increased to above 400°C more diffusible hydrogen was observed for cleaned specimen as compared to uncleaned specimens, although residual hydrogen was unaffected by different surface conditions.

Specimen Length Effect

The weld hydrogen measurement specimens have different dimensions depending on the standard used, Table 1. The fundamental idea is to optimize the specimen size to allow a wider range of welding parameters as well as to enhance outgassing. The IIW (II-1155-91)/BS 6693 standard has two specimen configurations; 30x15x10 mm for lower heat inputs and 15x30x10

mm for higher heat inputs. The JIS (Z 3118) allows different lengths and widths to allow for a wider variation in heat inputs and welding processes. The AWS A4.3-86 recommends 80x25x12.5 mm test specimens to allow for most of the welding processes and heat inputs.

The longer the specimen, including the start and end tabs, the more the time allowed for hydrogen escape before the specimen is quenched. For example, the AWS specimen is 80 mm long, and hence the start part of the test specimen will be at much lower temperature than the end part of the test specimen at the time of quenching, thereby allowing more hydrogen loss. A similar argument can be made for the IIW specimen. Even though the IIW specimen is much shorter than the AWS specimen, 15mm or 30mm depending on the orientation, the end tab is almost 80 to 85 mm long and thus the IIW test specimen (center block) will be at a much lower temperature at the time of quenching.

In theory, none of these specimens yield accurate information about the initially absorbed weld hydrogen, and the measured value is more of a reflection about the hydrogen remaining at the temperature (distribution) of the quench. This has been correctly pointed out by Gedeon¹², and, therefore, the use of weld hydrogen measurement by conventional standards to determine initially absorbed value at solidification, say for Sievert's rule verification, is incorrect. To gain a better understanding of the hydrogen value at solidification the technique used by Terasaki²⁸ should be used.

When measuring weld hydrogen content for preheat temperature determination using weldability tests, the difference in specimen length on weld hydrogen measurement becomes significant. The Tekken test weld is 76 to 80mm long, the Lehigh test weld is 125mm long. The initial weld hydrogen measured using a 15, 30, 40 or 80 mm long specimen will be much different than for a longer specimen. The longer weldability test specimen such as Lehigh will experience significant amount of hydrogen escape before the test weld is completed. Whereas the Tekken test will experience less hydrogen loss because of the shorter test weld lengths (76 to 80mm). If Tekken test and Lehigh test specimens are being used with the same welding

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procedures to assess HAZ and weld metal susceptibility to HAC, respectively, then, the weld hydrogen content determined by AWS or IIW specimen will not provide a correct idea about initial hydrogen content in these weldability tests.

Effect of Groove Shape

Additional contribution to the diffusible weld hydrogen difference between weld hydrogen measurement specimen and weldability test specimen may come from the groove shape difference. The standard weld hydrogen measurement specimens are bead on plate type, whereas, the weldability specimens have different groove geometries, e.g., U-groove in Lehigh test, oblique-Y groove in Tekken and single bevel in single bevel weldability test. The arc stability and weld pool size and shape will be different in these specimens compared to the weld hydrogen measurement specimens. This difference will be accentuated owing to the non-equilibrium nature of the welding process in the weld start and weld end regions of the weldability specimens. In the literature reviewed, these issues have been overlooked. In this author's opinion, an analysis and quantification of these differences is necessary for a better understanding of weldability testing procedure in itself.

OBJECTIVES OF THIS STUDY

This study had following specific objectives.

- 1) To study the base metal composition effect on weld hydrogen content.
- 2) To study the outgassing temperature effect on the amount of evolved hydrogen.
- 3) To study the prior residual hydrogen content effect on weld hydrogen content.
- 4) To study the specimen preparation effect on weld hydrogen content.
- To study the specimen configuration, i.e., specimen length and groove shape, effect on weld hydrogen content.

1.2 - EXPERIMENTAL

WELDING DETAILS

Pulsed gas metal arc welding (GMAW) using shielding gases containing different concentrations of hydrogen was carried out. The composition of shielding gases is given in Table 3. Use of certified cylinders ensured consistent composition of the shielding gas and consistent set of welding parameters throughout the study. The effect of different weld hydrogen levels was studied by maintaining welding parameters at a predetermined level for different shielding gases. Spray type of metal transfer was obtained with the welding parameters given in Table 3. A GE robot was used to get consistent and reproducible welds.

EXPERIMENTAL TEST MATRIX

Using welding conditions described above the effect of following variables on weld hydrogen content was studied.

Base Metal Composition Effect

The effect of base metal composition on weld hydrogen content was studied using ASTM A36 and A514 specimens. Weld hydrogen content was varied by using three different shielding gases with varying amount of hydrogen, Table 3. Outgassing was performed at 45 and 150°C.

Outgassing Temperature Effect

Hydrogen out-gassing was performed at 150°C for 6 hours and at 45°C using both A514 and A36 specimens. The extent of hydrogen trapping at 45°C and 150°C was studied by using the following experimental approach. Hydrogen measurements were carried out for different shielding gases after outgassing at 45°C for 72 hours for both A514 and A36 steel specimens. At the end of this measurement, outgassing was again performed at 150°C for 6 hours using the same sampler containing the same specimens after 45°C outgassing. Weld hydrogen measurements were again carried out at the end of 150°C outgassing treatment. The difference in the hydrogen content after these two outgassing treatments will reveal the amount of hydrogen trapped at 45°C but mobile at 150°C.

An additional study using A36 specimens was carried out with a 0.27% H₂ containing shielding gas. This involved R.T. outgassing (21 days) followed by measurement, 45°C for 96 hours outgassing followed by hydrogen measurement and 150°C for 6 hours outgassing followed by hydrogen measurement. The aim was to study the amount of trapped hydrogen that is released at different outgassing temperatures.

Prior Degassing Effect

In the first part of the study, weld hydrogen content of unwelded, i.e. prior degassed and undegassed (as received), specimen blanks determinations were carried out. Four A514 and A36 steel test specimen blanks were loaded in the specimen holding chamber and hydrogen outgassing was carried out at 150°C for 6 hours. Hydrogen measurements were carried out according to the procedure described earlier.

In the second part of the study, undegassed and de-gassed A514 and A36 steel specimens were welded using the parameters and shielding gases as mentioned in Table 3. The outgassing was carried out at 150°C for 6 hours and the weld hydrogen measurements were carried out as per the measurement procedure described earlier.

Specimen Preparation Effect

After the degassing treatment, twelve A36 test specimen assemblies were surface ground to final dimensions. In addition, twelve A36 test specimen assemblies were descaled by dry belt grinding. During surface preparation by dry belt grinding the specimens experienced temperature rise and were allowed to cool down before commencing further. A comparative study was performed using three different shielding gases on each type.

Specimen Length Effect

The specimen length effect on weld hydrogen content was studied using JIS (40x25x12.5 mm) and AWS (40x25x12.5 mm) specimens. These specimens are similar except that the JIS test specimen is half the length of AWS. Both A36 and A514 specimens were used and outgassing was performed at 150°C. The JIS recommends an outgassing temperature of 45°C, but, to eliminate the effect of outgassing temperature difference, if any, JIS specimen outgassing was carried out at 150°C.

Groove Shape Effect

A new specimen configuration with a U-groove, Figure 3, was used to study the effect of groove shape on weld hydrogen content. The grooved specimen length was 80 mm. No starting weld tab and run-off tabs were used. The U-groove was milled and the specimen was cut along the centerline to create the bottom slit. The two cut pieces were gas tungsten arc welded (GTAW) at the edges using inserts, taking the precaution that 2 mm gap was maintained. The hydrogen introduced from welding was kept at minimum by using pure Argon. The size of these welds was negligible compared to the size of the test weld. Thus, hydrogen introduced from GTAW of the edge welds was insignificant. The U-groove specimen test weld then simulated a Tekken test weld bead length.

1.3 - WELD HYDROGEN MEASUREMENT PROCEDURE

Hydrogen measurement procedure influences the final results significantly; hence the exact procedure used is outlined here. Most of the weld hydrogen measurements in this study were carried out according to the AWS A4.3-86 procedure. For comparative purposes, some of the determinations were performed according to the JIS standard (Z 3118). Each weld test assembly consisted of a starting weld tab, a test specimen at the center, and a run-off weld tab, Figure 1. Four such weld test assemblies constituted a complete diffusible weld hydrogen measurement test. The dimensions of all the three pieces of the test assembly are shown in Figure 1. To maintain a uniform cross section in the test assembly, specimens of dimensions 160x25x12.5 mm, slightly wider than the final dimensions, were cut from 12.5 mm thick rolled plates and processed further. Test specimens were prepared according to the AWS A4.3-86 procedure. After test welding at the preselected welding conditions, the specimens were subjected to hydrogen outgassing at the desired temperature.

Weld Hydrogen Measurement

After completion of the hydrogen outgassing heat treatments, the samplers were connected to the Oerlikon-Yanaco gas chromatograph (GC) for calibration and analysis. The GC specifications are shown in Table 4. The accuracy and reproducibility of the Oerlikon-Yanaco unit depends on calibration of the chromatograph and the samplers (outgassing units). The calibration procedure was carried out for both 150°C and 45°C samplers for AWS as well as JIS specimens.

Calibration Procedure for Gas Chromatograph

Calibration of the gas chromatograph generally is carried out by injecting a known volume of hydrogen (reference sample) into the carrier gas system. After setting up the chromatograph as per the manufacturers instructions, it was left to stabilize for at least 30 minutes. The sampler, after it cooled down to room temperature (in high temperature outgassing), was connected to the chromatograph. The chromatograph incorporates a hydrogen measuring tube that holds 1 ml. of hydrogen. This tube is filled with hydrogen from an external source for 5 seconds. After a wait of 5 seconds, this hydrogen is passed through the capsule/or bypass valve into the detector part of the chromatograph. The measured value is displayed on a digital readout and is then adjusted using a sensitivity dial to the standard temperature and pressure (STP) value supplied by the manufacturer. This procedure is repeated until the adjusted values remain constant.

Using the same calibration procedure, both 45°C and 150°C samplers were assessed for accuracy and reproducibility of each capsule (specimen holding chambers).

Calibration Procedure for 45°C sampler

The 45°C sampler has 4 capsules, Figure 2(a). A valve on the sampler controls the gas flow through any of the 4 capsules or allows bypassing them. The 4 capsules were loaded with blank (unwelded) specimens of 80x25x12.5 mm dimensions. The calibration procedure was then performed with the valve in bypass (B), capsule 1, capsule 2, capsule 3 and capsule 4 positions. The calibration prior to actual measurements is carried out with the valve in the bypass position. Therefore, the values during the sampler calibration with the valve in the capsules 1,2,3 or 4 should match with those when the valve is under bypass position. Reproducibility of the calibration procedure was also studied in this experiment.

Calibration Procedure for 150°C sampler

The 150°C sampler has 5 specimen holding chambers (capsules), Figure 2(b). The additional capsule is for calibration purposes and holds one blank (hydrogen free) specimen of dimensions 80x25x12.5 mm at all times. The gas flow in this capsule is controlled by setting the valve at B position and in other capsules by setting it at the corresponding capsule positions. The remaining 4 capsules were loaded with hydrogen free unwelded specimens of dimensions 80x25x12.5 mm. The calibration procedure was performed with the valve in bypass, 1, 2, 3 and 4 capsule positions.

Calibration Procedure for JIS specimens.

The gas flow rate and the volume of gas inside a capsule affects the measured value, hence a separate calibration procedure was carried out when using JIS specimens, 40x25x12.5 mm in the samplers (these were designed for the longer AWS specimens of dimensions 80x25x12.5 mm). This was performed using the 150°C sampler, as weld hydrogen measurements according to JIS (Z 3118) were carried out at 150°C. The capsule at B position contained blank (hydrogen free) AWS test specimen. In capsules 1-4, two blank JIS size specimens each were added. Thus volume occupied by the specimens was maintained at a constant level in each capsule. Calibration was then performed according to the procedure described earlier. Same procedure was used for the U grooved specimens, Figure 3.

STATISTICAL DECISION THEORY

In attempting to reach decisions, it is useful to make assumptions about the population involved. Such assumptions, which may or may not be true are called statistical hypotheses. If a hypothesis is rejected when it should be accepted, then a Type 1 error has been made. If, on the other hand, if a hypothesis is accepted when it should be rejected, then a Type II error has occurred. In testing a given hypothesis, the maximum probability with which one is willing to risk a Type 1 error is called the level of significance of the test. This probability is generally specified before any samples are drawn, so that results obtained will not influence the choice of error allowed²⁹. In practice, a level of significance of .05 or .01 is customary, although other values are used. If, for example a .05 or 5% level of significance is chosen in designing a test of hypothesis, then there are about 5 chances in 100 that one would reject the hypothesis when it should be accepted. Thus one is about 95% confident that a right decision has been made. In such case, it is said that the hypothesis has been rejected at a .05 level of significance. Some statisticians adopt the terminology that results at the .01 level are highly significant, results at .05 level but not at the .01 level are probably significant.

1.4 - RESULTS

CALIBRATION OF SAMPLERS

The calibration results on 45°C and 150°C samplers using the AWS specimens and on 150°C sampler using the JIS specimens are shown in Tables 5,6 and 7, respectively. They illustrate that consistent and reproducible results are obtained after the application of a proper calibration procedure. This suggests that final results will not be influenced by experimental error arising from calibration procedure and sampler variation.

SHIELDING GAS EFFECT

The weld hydrogen measurement results on A514 specimens after outgassing at 150° C show that weld hydrogen levels increased with an increase in the partial pressure of hydrogen in the shielding gas, Figure 4. This increase gradually dropped with an increase in the shielding gas hydrogen content. Scatter in weld hydrogen content values increased with an increase in the shielding gas hydrogen content because of the unstability in metal transfer mode, Table 8. The scatter increased significantly for the shielding gas containing 1% hydrogen. Hence, subsequent weld hydrogen measurements were performed without using this shielding gas. The weld hydrogen content using 1% H₂ containing shielding gas is about 16 ml/100 gm of DM. Even a scatter of about 1 ml/100 gm of DM made it difficult to isolate effects of different variables.

EFFECT OF DIFFERENT VARIABLES ON WELD HYDROGEN CONTENT

Base Metal Composition Effect

The results show that, Figure 5(a),(b) for both the outgassing treatments, the diffusible weld hydrogen content in A514 is higher than in the A36 specimens. This trend is observed with all the shielding gases used in this study. The difference ranges from 0.5 to 1.4 ml/100 gm of DM for different conditions. The statistical significance of the difference in means was analyzed by student t-tests analysis, Table 9. It is observed that the difference in the mean values for A514

and A36 specimens is significant at 5% significance level. In some cases, this difference is significant at 1% level of significance. These results suggest that the base metal composition influences the diffusible weld hydrogen levels.

Outgassing Treatment Effect

Figure 6(a),(b) suggest that the outgassing temperature does not affect the weld metal hydrogen levels for both the steels. No consistent trend is observed at any hydrogen levels. Student t-test analysis also supports this conclusion, Table 10, i.e., the difference between the means is not statistically significant at 5% level. The effect of outgassing temperature on diffusible weld hydrogen measurements is negligible.

The results after 45°C and 45+150°C outgassing of the same specimens, Figure 7(a),7(b) indicate that the additional outgassing at 150°C releases some of the hydrogen not completely evolved at 45°C. The amount of this additionally evolved hydrogen is not significantly high, Table 11, and is about 0.16 - 0.66 ml/100 gm of DM for different base metal compositions. This probably explains the observed trend that effect of outgassing temperature on the diffusible weld hydrogen level is not significant at 5% level.

An additional study on A36 specimens involved hydrogen measurement at the end each outgassing treatment, i.e., 21 days at R.T., 45°C for 96 hours and 150°C for 6 hours. From the results, Table 12, it is clear that the amount of hydrogen that evolves after 45°C outgassing is negligible and is about 0.03 ml/100 gm of DM. But, after 150°C outgassing, some hydrogen is released, i.e. 0.23 ml/100 gm of DM. This additionally evolved hydrogen is similar to that observed in the previous experiment for A36 specimens with 0.269% shielding gas, Table 11. This additionally evolved hydrogen was immobile or could not be completely removed at RT and 45°C but became diffusible at 150°C. But, again the amount of evolved hydrogen that is released at 150°C is not significantly high. Therefore, the difference in the weld hydrogen levels after 45°C and at 150°C outgassing, under identical welding conditions, will go undetected.

Prior Degassing Effect

Unwelded Specimens

The residual hydrogen content of unwelded A36 and A514 specimens in both degassed (at 650°C for 1 hour) and undegassed condition (as received) is shown in Table 13. The residual hydrogen content for degassed specimens is negligible in both A36 and A514 specimens, i.e., about 0.024 ml/100 gm of DM. This is within the measuring error during the analysis, and hence can be neglected. Undegassed A36 specimens exhibit 0.045 ml/100 gm of DM hydrogen content which is higher than that in undegassed A514 specimens, 0.017 ml/100 gm of DM. Student t-test shows that this difference between the undegassed specimen is significant. The average residual diffusible hydrogen amount at 150°C for undegassed A36 specimens is 0.096 ml for each test specimen. After welding and subsequent measurement, the residual hydrogen will get measured with the total weld hydrogen, i.e., the contribution of the residual hydrogen is 0.096 ml to the final weld hydrogen value.

Welded Specimens

A36 specimens

At all the weld hydrogen levels, the undegassed specimens exhibited hydrogen contents that are at least 1 ml/100 gm of DM higher than the degassed specimen, Figure 8(a). For the weld hydrogen levels being measured, this is a substantial difference. The student t-test analysis also supports that this difference in mean values is significant at 5% level for all the hydrogen levels, Table 14. The calculated contribution of residual hydrogen content from the undegassed base metal in the A36 specimen was 0.35 ml/100 gms of DM (0.096 ml for each specimen) to the final weld hydrogen value. Whereas, the difference between degassed and welded and, undegassed and welded specimens is about 1 ml/100 gm of deposited metal. This clearly suggests that during welding, more residual hydrogen (that is still immobile at 150°C) becomes and stays, mobile in the base metal that reaches temperatures above 150°C thus contributing additional hydrogen to the final weld hydrogen value.

A514 specimens

The results, Figure 8(b), indicate that the difference between degassed and undegassed specimen is insignificant which is further supported by the t-test analysis, Table 14. This suggests that, in A514 specimens, the contribution of residual hydrogen to final weld hydrogen values is negligible.

Surface Preparation Effect

Surface preparation influences the weld hydrogen values, Figure 9. Belt ground specimens show higher weld hydrogen levels than the surface ground specimens for identical welding conditions at all the weld hydrogen levels analyzed. The t-test analysis, Table 15, indicates that, for shielding gas hydrogen levels of 0% H_2 and 0.269 % H_2 , these differences are significant at 1% level, whereas, for 0.51% H_2 the differences are significant at 5% level.

Specimen Length Effect

From Figure 10(a),(b) it is observed that JIS specimen yields more weld hydrogen for 0.269% H₂ and 0.51% H₂, in both A36 and A514 specimens, than the AWS specimen. Whereas, for 0% H₂ shielding gas the opposite is true. This suggests that the specimen length has some influence on the weld hydrogen content. Student t-test analysis, however, indicates that the difference between the weld hydrogen means of AWS and JIS specimen is not significant at 5% level, Table 16. Therefore, for practical purposes, both these specimen will yield same amount of weld hydrogen under identical welding condition. Another important result is the scatter for JIS specimens. From the variance calculations, Table 16, it is observed that the JIS specimen is more prone to variation in weld hydrogen levels as compared to the AWS specimen.

Groove Shape Effect

From Figure 11, it is observed that grooved specimen shows slightly lower weld hydrogen levels compared to the standard AWS specimen. Student t-test analysis, Table 17, suggests that this difference is not significant at 5% level. Although, in this author's opinion, this question is worthy of a detailed study with a larger data base. The measured hydrogen level scatter is higher in the grooved specimen than the bead on plate specimen. Since the grooved specimen weld bead was equivalent to a Tekken test weld bead, the above results suggest that the influence of an actual bead groove shape can be important.

1.5 - DISCUSSION

GAS CHROMATOGRAPH CHARACTERISTICS

It was observed that the calibration procedure was extremely sensitive to carrier gas flow rate and room temperature variation. Before every calibration, the GC unit was left on for about 45 minutes to 1 hour for stabilization. Stabilization is indicated by a zero-level on the digital display that can be adjusted by using a zero adjustment switch. The stabilization time depended on whether the unit was in regular use or not. If the unit was started after a long shut-down, then stabilization took more than 1 hour. In this case, the unit shut down and calibration sequence had to be carried out 3 to 4 times before calibrated values were near the value suggested by the manufacturer. According to the manufacturer, the stabilization (ready) time for the unit is about 10 minutes. In the present work, it always took more than 30 minutes. The stabilization process is critical as the calibrated values drift continuously if the GC is not stabilized.

During the hydrogen measurement procedure, zero level of the unit drifted between successive measurements of each sample and had to be brought back to the zero level before proceeding to the measurement of hydrogen in the next capsule. Enquiry with the manufacturer revealed that generally the zero-level should not change and should be reset to zero if it drifts substantially during measurement. This deviation from the zero-level during measurement depended on the amount hydrogen being measured. For weld hydrogen contents greater than 10 ml/100 gm DM the deviation was significantly higher compared to lower weld hydrogen levels. Even after a waiting period of 2-3 minutes, the zero-level was not achieved and then had to be forced through the use of zero-level adjustment switch. This has also been reported by Gedeon¹².

The valve controlling the carrier gas (argon) flow rate was highly sensitive to touch and slight disturbance caused large fluctuations in the carrier gas flow rate. Additionally, this valve had to be adjusted to set the proper flow rate before each calibration and measurement. This adjustment is required because the detector linearity and sensitivity varies with carrier gas flow

rate. If this valve is made less sensitive to touch then good control over carrier gas flow rate can be maintained easily. In addition, when used at carrier gas pressures slightly different than those provided by the manufacturer, the readings significantly deviated and calibration could not be performed as the hydrogen level being calibrated always fell beyond the instrument readout. Before every measurement, carrier gas pressure and flow rates were maintained at the values suggested by the manufacturer.

The maximum readout on the digital display is 2.00. Therefore, the maximum limit on evolved hydrogen which can be displayed, and hence measured is 2.00 ml. In the present study, the average weight of deposited weld metal varied between 17 to 20 gms. Rough calculation suggests that maximum hydrogen that could be measured varied from 10 to 11.76 ml per 100 gm of DM. If higher hydrogen levels are encountered, then modifications are required in the weld hydrogen measurement procedure. Kotecki³⁰ suggested the use of two 40x25x12.5 mm test blocks, instead of one 80x25x12.5 mm test block, to reduce the volume of hydrogen being measured at a time. Another way to circumvent this problem is to perform measurement selection hydrogen measurement and delays the outgassing treatment. For the lack of a better alternative, the latter procedure was adopted. Alternatively, the manufacturer should increase the hydrogen volume that can be measured.

When outgassing at 45 and 150°C, the outgassing unit (sampler) should be allowed to cool to room temperature before commencing calibration and measurement. Otherwise, the measured value will be incorrect. This arises from the basis that the detector sensitivity and linearity degrades with temperature. In addition, the carrier gas temperature rise (due to the passage through the capsules and tubing) will alter its thermal conductivity, thereby, changing the electrical resistance of the thermistor element.

CALIBRATION ON DIFFERENT SAMPLERS

The calibration and measurement procedure was standardized as follows :

- 1 Cool-down of samplers to RT and then connection to GC
- 2 Stabilization time of at least 45 minutes for the GC
- 3 Calibration and adjustment of calibrated value to that provided by the manufacturer followed by 10 more runs to ensure the stability of the calibrated value.

It was observed that a rigorous control over the calibration procedure was necessary in getting reproducible measurements. The reproducibility could be guaranteed for hydrogen measurements under identical conditions even after a gap of 6-8 months if these calibration procedures were followed.

45°C Sampler

The results for 45°C sampler calibration show that the measured hydrogen value with the gas flow control valve in bypass position matches closely with valve in capsule positions 1, 2, 3 and 4, Table 5. But, the bypass value is always higher by .006-.007 than the value measured at individual capsule positions. Therefore, actual hydrogen measurements will be lower by this amount. The measurements in this study generally involved 18 gms of deposited metal. So, when converting the evolved value for as deposited weld to the standard reported value of per 100 gms of weld metal, the above mentioned error will cause an error of (100/18)x.007 = .035 ml per 100 gm of DM. The error is almost equal to the specified measurement error and, hence, should not be considered as significant. Enquiry with the manufacturer revealed that there should not be any difference between the calibration at bypass position and at other capsule positions. From the above, it can be concluded that calibration of the 45°C sampler with the flow control valve in bypass position should accurately reflect the amount of hydrogen inside the capsules.

150°C Sampler

The calibration results for 150°C are shown in Table 6. It is clear from the readings that measurement of hydrogen can also be made accurately and consistently with the 150°C sampler. In this case, there was no difference in the hydrogen value between the bypass and other capsule position. Each capsule results in an accurate measurement of the amount of hydrogen in it. The standard deviation for these sampler readings is about .0004 suggesting that the calibration procedure is highly reproducible.

JIS Specimens Using 150°C Sampler

The samplers for hydrogen outgassing were ordered specifically for AWS specimens which are longer than IIW and JIS specimens, Table 1. The specific gas volume and flow rate through a capsule containing JIS specimen will result in different hydrogen values compared to the one containing AWS specimen. This problem can be circumvented by making the volume occupied by the specimens inside the capsules equal to that occupied by AWS specimens. The JIS specimens are 40x25x12.5 mm. Thus, the addition of specimen blanks (hydrogen free) of same dimensions in the capsules will give an equivalent of AWS specimen. The same sampler can then be used to measure weld hydrogen in JIS specimens. To check the validity of this argument, calibration was performed with JIS specimen plus specimen blank of same dimension in different capsules and also in bypass capsule using the 150°C sampler. The results of the calibration are shown in Table 7. These calibration results show that this modified technique of hydrogen measurement by JIS standard can be employed successfully by using the sampler designed specifically for AWS specimens.

In the author's opinion, the 150°C sampler should be preferred over the 45°C sampler. In the 150°C sampler, the calibration is performed with the same size specimen blanks as are being used for hydrogen measurements. Unlike the 45°C sampler, there is no difference in measurements in bypass and different capsule positions. In addition, different specimen geometries (such

as JIS and even IIW) can be used for hydrogen measurements. Although no measurements were carried out using the IIW specimens, it is felt that the same procedure as was used for JIS specimens can be adhered to for IIW specimens.

SHIELDING GAS COMPOSITION EFFECT

Increasing the partial pressure of hydrogen in the shielding gas resulted in increasing weld hydrogen content, Table 8 and Figure 4. This trend was observed for both A514 and A36 base materials at 150°C and at 45°C. With the welding parameters used in this study, a stable spray transfer was obtained for all the shielding gases, except for the one containing 1% hydrogen. For the shielding gas containing 1% hydrogen, the welding parameters resulted in welding conditions at the transition between spray and globular transfer. Since the amount of weld hydrogen is a strong function of type of metal transfer¹², this shielding gas exhibited a significant scatter in measured weld hydrogen levels. It was then decided to exclude this shielding gas from the analysis of the effect of different variables on weld hydrogen levels.

It was also observed that scatter increased with increasing hydrogen concentration in the shielding gas, Table 8. Though scatter with the shielding gas containing 0.51% of hydrogen was not significant enough to affect subsequent study of different variables on weld hydrogen content.

BASE METAL COMPOSITION EFFECT

The results in Figure 5(a),(b) show that weld hydrogen levels with A514 base material are higher than those observed with A36 material. This was observed with the 150°C as well as the 45°C outgassing treatment. To study if these results are significantly different, student-t test analysis was carried out. From Table 9, it is observed that the difference in the means of weld hydrogen value for A514 and A36 is significant at 5% level. This trend was also repeated with JIS specimens as shown later, Table 16.

There are two probable explanations for this behavior. The first reason could be the

influence of different microstructures on hydrogen diffusivity. The second reason could be that the hydrogen absorption itself is influenced by base metal composition. Hydrogen trapping because of various line and volume defects is well documented^{20, 23, 25}. The release of hydrogen from these defects occurs at different temperatures, depending on the binding energy between the trap and the hydrogen. For A514 specimens, the microstructure consisted of untempered martensite after quenching the weld hydrogen specimen, while the A36 a mixture of ferrite and pearlite. In this context, Asaoka²⁰ found that as quenched microstructure has a more pronounced hydrogen trapping effect²⁰. Using autoradiography, Asaoka did not observe any trapping effect from alloying additions, but postulated that, since precipitate matrix interface and prior-austenite grain boundary characteristics are changed by alloying elements, the hydrogen trapping is influenced by alloying element additions. In comparison, in A36 specimens, the microstructure will be less hardenable due to the absence of alloying elements. As a result, hydrogen trapping will be less pronounced compared to A514 specimens. By the same reasoning, hydrogen diffusivity will be lower in A514 specimens as compared to A36 specimens.

An expected lower hydrogen evolution rate and higher trapping in A514 suggests that A514 will experience lesser hydrogen release. But the A514 specimens exhibit higher weld hydrogen than A36 specimens. This suggests that the trapping effect is not influencing the observed results.

The diffusivity measurements for A36 specimens at RT, explained in chapter 3, yield a hydrogen diffusivity of 7.06e-07. The extrapolated diffusivity at 45°C and 150°C suggests that 100% hydrogen evolution is ensured at 45°C and 150°C outgassing treatments. The hydrogen diffusivity in A514 is expected to be lower than in A36 because of the microstructure influence. Therefore, from diffusivity considerations alone, A36 specimens should exhibit more weld hydrogen if evolution is incomplete. Since the hydrogen evolution is complete, it can be concluded that base metal composition effect can not be attributed to hydrogen diffusivity differences in A36 and A514 specimens.

Gedeon¹² found that the diffusible hydrogen content of high strength steel (MIL-46100) was consistently 2 to 3 ppm higher than in A36 mild steel specimens. Gedeon further postulated that the increased alloy content of high strength steel should decrease the hydrogen solubility during welding. Since A514 steel specimen contains more alloying elements than A36 specimens, the weld hydrogen levels should be lower in A514 as compared to the A36 specimen by the above reasoning. But the results in Figure 5(a),(b) are in contradiction to this. This suggests that alloying elements do not affect hydrogen absorption during the welding process itself in the present study.

Gedeon also found that increasing the carbon content increased the amount of weld diffusible hydrogen. This is contrary to the observed fact that carbon reduces the solubility of hydrogen by reducing the oxygen partial pressure (oxygen in the arc atmosphere increases hydrogen absorption)¹². Gedeon has tentatively explained this on the basis that carbon lowers the melting temperature and thus will promote monoatomic hydrogen absorption into the cooler outer edges of the weld pool. The difference in carbon content of A36 and A514 is significant to cause a change in the weld hydrogen content. A36 contains about 0.28% carbon compared to 0.19% in A514. If we consider the fact that carbon reduces the solubility of hydrogen by reducing the oxygen partial pressure then A36 specimen should exhibit lower hydrogen content which is in accordance with that observed in this study. Therefore, the carbon content influences the weld hydrogen absorption process during welding.

Clearly, the choice of base material influences the weld hydrogen levels. For classification of welding consumables, the base material should be selected such that it matches the one used in service. For determination of weld hydrogen levels for use in preheating temperature calculations, the same base material should be used in weld hydrogen determination as will be used in actual weldability tests.

OUTGASSING TREATMENT EFFECT

The difference in the outgassing heat treatments at 45° C and 150° C is not significant, Figure 6(a),(b) indicating that hydrogen evolution is complete after these outgassing treatments or remaining hydrogen is insignificant. The above trend is observed in both A36 and A514 specimens. The differences in the mean hydrogen values are not significant at 5% level of significance, Table 10.

From the above, the two outgassing temperatures used in this research work do not affect the weld hydrogen levels, which is contrary to that reported in the literature, Table 2. In these studies, hydrogen was externally introduced by hydrogenation using electrolysis and/or high temperature charging. This mode of hydrogen charging introduces defects and extreme internal hydrogen pressures, and, hence this author postulates that the above results are influenced by the charging methods. As a result, residual hydrogen data from studies using welded specimens should be used as the basis for determining expected hydrogen outgassing characteristics of weldability test.

Recently, Dilthey and Trube²¹ found that the reactivation temperature of residual hydrogen is dependent on the electrode type and on the total hydrogen content of the weld. These authors quote an activation temperature of 80°C for a cellulosic electrode deposited weld metal. Coe and Chano²² found that residual hydrogen formation occurred at temperatures lower than 300°C and was dependent on steel and weld metal composition, the number of trapping sites available and on the total amount of hydrogen present. They proposed that the composition effect is a direct result of microstructure and inclusion type.

It is clear from the above studies, that the results from the residual hydrogen/trapping studies can not be applied to other compositions. This precludes the use of these reactivation temperatures. Hence, a statement about the effect of trapping can not be made based on these results. In addition, it should be realized that the amount of trapped hydrogen is small (depending on the total weld hydrogen). Its detection will be masked by the scatter involved in the weld hydrogen measurement procedure. This difficulty can be overcome if a single set of welded specimens is subjected to outgassing at different temperatures starting with complete outgassing at the lowest outgassing temperature and subsequently increasing the outgassing temperature (ensuring complete outgassing).

A study was conducted that involved weld hydrogen outgassing at 45°C for 72 hours and measurement followed by subjecting the same specimens to outgassing at 150°C for 6 hours and measurement. The results from this study, Figure 7(a),(b) and Table 11, show that additional hydrogen is released after 150°C outgassing in both A36 and A514 specimens. This additionally evolved hydrogen amount is very low and it explains its non-detection due to the masking associated with the scatter involved.

A similar study on A36 specimens with outgassing and measurement at room temperature, 45°C and 150°C was conducted. From Table 12, it is observed that there is no difference in the evolved weld hydrogen values at room temperature and at 45°C, suggesting that no additional hydrogen is released by 45°C outgassing. In comparison, at 150°C evolution of additional hydrogen is observed. The amount of additionally released hydrogen in A36 specimens in this case is remarkably similar to that mentioned in the first part of study (45 and 45+ 150°C outgassing), Table 11.

The additional evolution at 150°C in the above experiments could be because of two reasons. First, the hydrogen evolution at 45°c could be incomplete, hence with 150°C outgassing, additional hydrogen evolution is observed. Second reason could be due to the effect of hydrogen trapping. The completion/incompletion of hydrogen outgassing can be checked by a simple calculation involving the calculation of D Δ t, the thermal factor or the diffusion distance, for 45°C and for 150°C outgassing temperatures. Where, D is hydrogen diffusivity in steel at a given temperature and Δ t is the time spent at that temperature. Hydrogen outgassing is generally characterized by relations of the following type :

$$\frac{H_r}{H_o} = A \exp(B D\Delta t)$$
(1)

Where, H_r is the amount of hydrogen remaining in the specimen, H_o is the initial hydrogen in the specimen. Therefore, $\frac{H_r}{H_o}$ represents residual hydrogen fraction remaining in the specimen. Using the finite element method, hydrogen outgassing relations for the AWS specimens were determined as will be described in Chapter 3 on hydrogen outgassing. The numerically determined hydrogen outgassing relation for the AWS specimen used in this study is as follows.

$$\frac{H_{\rm r}}{H_{\rm o}} = 0.82 \exp(-8.02 \,\mathrm{D}\Delta t\,)$$
(2)

Using this relation, $\frac{H_r}{H_o}$ for 45 and 150°C temperatures were calculated using hydrogen diffusivities at the mean and lower end of the hydrogen diffusivity spectrum. The results are shown in Table 18. It is clear from Table 18 that hydrogen outgassing is a strong function of hydrogen diffusivity. For example, steels with mean diffusivity will experience complete outgassing at 45°C. In this case, if additional outgassing at 150°C evolves more hydrogen then it suggests that it is trapped hydrogen. In comparison at low diffusivities, the 45°C outgassing treatment only removes 20% of the initial hydrogen, H_o. In this case, the subsequent 150°C outgassing treatment will result in additional hydrogen evolution. This additionally evolved hydrogen will now comprise of hydrogen still remaining after 45°C incomplete outgassing and trapped hydrogen, if any. In such case, it will be difficult to distinguish between the type of evolved hydrogen. The exact amount of outgassing as a function of temperature can be determined if hydrogen hydrogen diffusivity at these two temperatures is known. Therefore, in the present present, it is difficult to determine the exact nature of the evolved hydrogen with additional outgassing at 150°C.

In this author's opinion, the outgassing temperature effect on hydrogen evolution should be

studied using a single weld specimen instead of using different welded specimens at different temperatures. This will avoid the error in hydrogen values due to the experimental variation in welding process itself. This technique can then be successfully used to reveal the reactivation temperature and the effect of outgassing temperature on weld hydrogen content.

PRIOR DEGASSING EFFECT

Unwelded Specimens

The residual hydrogen measurements on undegassed and unwelded (as-received) A36 and A514 base materials show that, Table 13, the residual hydrogen content in A36 specimens is substantially higher than that in A514. Student t-test analysis also supports this conclusion. Whereas, prior degassed and unwelded A36 and A514 specimens do not show any residual hydrogen content. This suggests that the prior outgassing heat treatment removes most of the residual hydrogen in the as received material.

Welded Specimens

The weld hydrogen measurements using undegassed (as received) specimens also corroborate the influence of residual hydrogen, Figure 8(a), (b). A36 undegassed specimens show significantly higher (almost 1 ppm) weld hydrogen contents compared to the degassed specimens. The differences are significant at 5% level, Table 14. For A514 specimens, the weld hydrogen content is unchanged in both undegassed and degassed specimens. This could be due to the quenching and tempering heat treatment, of A514 steel. Tempering at 600°C removes most of the residual hydrogen.

Clearly, the residual hydrogen content influences the weld hydrogen content depending on the material, the microstructure and the thermo-mechanical history. Enquiry with the local steel producer revealed that the hydrogen level during steel making is maintained at about 5-7 ppm for all the steels. Therefore, A36 and A514 ingots should have similar residual hydrogen content before they undergo any further thermo-mechanical treatment. The subsequent hot-rolling operation further lowers the residual hydrogen content. After the hot-rolling operation, the A36 plate is not subjected to any heat treatment, whereas, the A514 plate undergoes quenching and tempering heat treatment. This tempering treatment is carried out at 600°C for 1 hour. At this tempering temperature, hydrogen is highly mobile, and therefore, this tempering treatment results in additional hydrogen removal. This tempering treatment is in fact equivalent to prior degassing treatment of AWS specimens which is carried out at 600–650°C for 1 hour. As a result, in the as received condition, the A514 steel plate has considerably lower amounts of residual hydrogen than the A36 steel plate. This is also supported by the residual hydrogen measurements on the as received A514 and A36 base materials, Table 13. The quenched and tempered A514 steel is expected to contain a significantly high density of trapping sites, which is expected to reduce the diffusible hydrogen content even further.

The residual hydrogen content for undegassed (as received) A36 test specimens is 0.0963 ml and will remain constant for the A36 steel used in this study. The average weight of the weld deposit in this study is 18 gms. Therefore, the residual hydrogen value of 0.0963 will contribute $0.0963 \times (100/18) = 0.5$ ppm to the final weld hydrogen content.

The undegassed and welded A36 specimens have weld hydrogen levels that are 1 ppm higher than the degassed and welded A36 specimens. The contribution of residual hydrogen to the weld hydrogen value as calculated in the previous paragraph is 0.5 ppm. About 0.5 ppm of hydrogen is still unaccounted for by residual hydrogen contribution to final weld hydrogen value. This can be explained on the basis that additional residual hydrogen becomes mobile hydrogen during the test welding cycle. The residual hydrogen in undegassed and unwelded specimens was determined by outgassing at 150°C. Therefore, residual hydrogen mobile at temperatures lower than 150°C was outgassed and measured but residual hydrogen with higher activation temperature remained in the specimen.

During welding of these undegassed specimens, more residual hydrogen became mobile as the weld specimen is heated to temperatures above 150°C. In A36 undegassed specimens, this type of conversion contributes about 0.5 ppm hydrogen. Whereas, in A514 specimens, because of the tempering heat treatment, this effect is negligible.

This change in weld hydrogen content as a function of prior residual hydrogen content is important in weldability assessment as a change of 1 ml/100 gm DM can change estimated T_{ph} from a no crack to a cracking situation. For example, while assessing A36 HAC susceptibility using Tekken tests and with the T_{ph} from a suitable preheat prediction method, the predicted T_{ph} will vary markedly depending on the initial hydrogen content. If the weld hydrogen content is determined by using A36 prior degassed AWS specimens then the predicted T_{ph} will correspond to this hydrogen value. But, in A36 Tekken specimens, which are in as received condition, the resulting weld hydrogen value will be higher than that determined from the prior degassed AWS specimens. As a result the predicted T_{ph} may be too low to avoid cracking in Tekken specimens. This variation in weld hydrogen content depend on the steel making operation, the type of steel and the thermo-mechanical history. The weld hydrogen content variation due to prior residual content could account for the failure of different preheat prediction methods. From this perspective, when determining weld hydrogen content for HAC susceptibility assessment the AWS specimens should not be prior degassed to obtain more relevant weld hydrogen contents.

SPECIMEN PREPARATION EFFECT

The AWS A4.3 procedure does not specify a single specimen preparation technique but recommends different methods. In this study, A36 degassed specimens were prepared by surface grinding and dry belt grinding. From Figure 9 it is observed that weld hydrogen content in dry belt ground specimens is higher than that in surface ground specimens by at least 0.8 ppm. The student-t test analysis suggests that this difference in weld hydrogen mean values is significant at a 5% level, Table 15.

During surface grinding, compressive stresses develop at the surface of the specimen and tensile stresses develop just below the compressive stress layer³¹. These stresses will affect the diffusion of hydrogen. It will lead to increased trapping where the stresses are tensile, and thus less diffusible hydrogen will be available.

It can be concluded from the above that different specimen preparation techniques, as suggested by different weld hydrogen measurement standards, will affect the weld hydrogen level. For example, AWS A4.3-86 suggest surface preparation by dry belt grinding, dry shot blasting, dry grit blasting, or dry power wire brushing. The IIW standard (II-1155-91) recommends use of a surface grinder and JIS 3118 recommends grinding. Therefore, the exact specimen preparation will vary depending on the availability of proper equipment. Weld hydrogen content in specimens prepared by two different techniques as per these standards will be different. Therefore, the surface preparation technique should be kept constant, when comparing weld hydrogen contents using different standards, to eliminate the effect of surface preparation on weld hydrogen content.

Different preheat prediction schemes use weld hydrogen levels in terms of the IIW specimen. But, if the weld hydrogen is determined by using JIS and AWS specimens, then conversion to IIW weld hydrogen values using established correlations can be misleading because the effect of surface preparation is not considered in these correlations. Therefore, the conditions under which these correlations were derived should be cited and be strictly adhered to. To avoid variation in weld hydrogen content sue to different surface preparation techniques, all hydrogen measurement standards should exactly specify the surface preparation techniques and these should be unique.

The specimen preparation technique should be fast, easy and economical and should not alter the physical, chemical and mechanical state of the specimen (especially the surface). The above two criteria are met by dry belt grinding. The specimen preparation by dry belt grinding is easier and faster than that by surface grinding and/or machining as it does not any involve set-up. Surface grinding alters the stress state of specimen surface, and hence it should be avoided³¹.

SPECIMEN LENGTH EFFECT

This effect was studied by performing weld hydrogen measurements using the AWS and JIS specimens. Outgassing of the weld hydrogen was performed at 150° C. Even though JIS standard recommends use of 45° C outgassing treatment, 150° C was used to avoid the potential effect of different outgassing temperatures. From Figure 10(a) (b), it is observed that JIS specimens exhibit somewhat higher hydrogen than AWS specimens for shielding gases containing 0.269% and 0.51% H₂. In comparison, in shielding gases containing no hydrogen the opposite is true. The student-t test analysis, Table 16, however, shows that there is no significant difference in the weld hydrogen levels measured by these two methods. When comparing the standard deviation values, the JIS specimens exhibit larger scatter compared to the AWS specimens.

AWS and JIS specimens are identical except that the AWS specimens are twice as long. The specimen length affects the weld hydrogen levels in two ways. The longer the specimen, the more the hydrogen loss before quenching the specimens. The welding parameter variation, and hence, the scatter will be higher in shorter specimens. It is felt that these two factors qualitatively explain the results obtained in this study. Additional studies are needed for quantification of these differences. Other researchers have performed similar comparisons between IIW and AWS standards. Evans¹⁹ observed that AWS specimens outgassed at 45°C and 150°C using the GC exhibited 95.2% of weld hydrogen as compared to the IIW specimens. Kotecki¹⁷ also reported that AWS weld hydrogen values are 93% lower than those from IIW weld hydrogen values. This difference is most likely caused by the higher hydrogen loss before quenching the specimen in AWS specimens compared to the IIW specimens.

These results suggest that hydrogen loss before quenching the specimens controls the weld hydrogen levels. Similarly, the JIS specimens should show higher hydrogen than the AWS specimens because of the influence of weld hydrogen content. The results in the present study point out to a specimen length effect on weld hydrogen content.
SPECIMEN GROOVE SHAPE EFFECT

The grooved specimen used in this study was similar to the Tekken test bead in bead length. It also incorporated the effects of weld start and weld end and closely simulated actual Tekken test conditions. The results show that the AWS specimen always exhibits slightly higher weld hydrogen contents than the grooved specimen, Figure 11. The student-t test analysis, however, suggests that these differences are not significant at 5% level for two out of three shielding gases, Table 17. The scatter in weld hydrogen levels is higher for the grooved specimens indicating the influence of weld start and weld end effects. It has been observed by Evans³² that start and end perturbations affect weld hydrogen content. The scatter at all the shielding gas hydrogen content variation due to start and end perturbations. This is more important from the weldability test viewpoint. The difference in weld hydrogen levels in a weldability test and those from the standard AWS specimens will be more pronounced in weldability tests with a longer weld length, such as Lehigh test specimen where the test weld length is 125 mm.

The above results, though qualitative, suggest a need for a change in the weld hydrogen specimen configuration as applied to weldability tests. When the weld hydrogen results are required for preheat temperature prediction, the bead length should be same as the weldability test specimen bead length. This will incorporate all the perturbations and also the inherent bead characteristic to get realistic weld hydrogen values.

The grooved specimen is expensive to machine, and hence can not be used frequently. In addition, melt-through problems at different welding conditions prevents its use for different welding parameters and processes. If, instead of a grooved specimen, a bead-on-plate (BOP) specimen (similar to AWS but without run on and off tabs) is used then initial hydrogen values will not be significantly different as the groove shape only affects the remaining diffusible hydrogen. Therefore, for weld hydrogen values to be used in weldability tests, AWS type specimen should be used with same bead length as the weldability test bead length.

1.6 - CONCLUSIONS

- Gas chromatograph method can be used to consistently and precisely measure weld hydrogen contents. Because of this, GC method can be used to evaluate the effect of different variables on weld hydrogen contents.
- 2) The base metal composition significantly influences the weld hydrogen content through a chemical composition effect on the hydrogen absorption process itself. Therefore, from a weldability testing view point, the base metal should be kept constant in both weldability testing and weld hydrogen measurement.
- 3) Residual hydrogen in the as-received condition content increased the final weld hydrogen content. This effect is significant from weldability testing viewpoint since the weldability test specimens are not subjected to residual hydrogen removal compared to the weld hydrogen measurement specimens. The residual hydrogen content depends on the steel making practice and the subsequent thermo-mechanical history.
- 4) The outgassing temperature did not influence the weld hydrogen content significantly.
- 5) The specimen configuration effect on weld hydrogen content was insignificant. It was observed that with a shorter specimen the scatter in weld hydrogen content increased. Similar effect was observed in the case of effect of groove configuration on weld hydrogen content.
- 6) It can be concluded from this study that the weldability tests are conducted based on weld hydrogen content which are not representative of those in actual weldability test specimens. This will result in a wrong assessment of HAC susceptibility of different steels as applied to industrial fabrication.





* DIMENSIONS IN mm

Figure 1. Weld Hydrogen Measurement Specimen Dimensions (a) AWS Specimen, (b) JIS Specimen



Figure 2. Specimen Holding Chambers for Hydrogen Outgassing (a) 45 $^{\rm o}{\rm C},$ (b) 150 $^{\rm o}{\rm C}$



Figure 3. Grooved Specimen Configuration



Figure 4. Weld Hydrogen Variation as a Function of Shielding Gas Hydrogen Content



Figure 5. Base Metal Composition Effect on Weld Hydrogen Content (a) Outgassing at 45°C, (b) Outgassing at 150°C



Figure 6. Outgassing Temperature Effect on Weld Hydrogen Content (a) For A36 Specimens, (b) For A514 Specimens



Figure 7. Hydrogen Evolution after 45°C and 45+ 150°C Outgassing (a) For A36 Specimens, (b) For A514 Specimens



Figure 8. Prior-degassing Effect on Weld Hydrogen Content (a) For A36 Specimens, (b) For A514 Specimens



Figure 9: Specimen Preparation Effect on Weld Hydrogen Content

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Figure 10. Specimen Length Effect on Weld Hydrogen Content (a) For A36 Specimens, (b) For A514 Specimens



Figure 11. Groove Effect on Weld Hydrogen Content for A514 Specimens

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SPECIMEN SIZE	OUTGASSING TREATMENT
IIW Standard	(a) Primary method :
(to replace ISO 3690)	At 25±5°C until
	no increase in calculated hydrogen
30Lx15Wx10T mm for heat	on successive days (approx. 14 days)
input below 2 kJ/mm	(b) Rapid methods :
	At 45°C for 72 hours
15Lx30Wx10T mm for heat	At 150°C for 6 hours
input above 2 kJ/mm	(Does not recommend temperature > $150^{\circ}C$)
AWS A4.3-86	45±3°C for 72 hours
80Lx25Wx12.5 mm	150±3°C for 6 hours
JIS Z 3118	Gas Chromatograph - 45°C for 72 hours
40Lx25Wx12.5 mm for the	Glycerine method - 45°C for 75 hours
heat input used in this study	

Table 1 : Weld Hydrogen Measurement Standards

Trapping Site Classification	Reactivation Temperature ^o C	Reference
grain boundary (pure iron)	118	
Fe-C interface (0.49% C Steel)	129	
Fe-C interface (AISI 4340 Steel)	115	
dislocation (pure iron)	253	
dislocation (0.49% C steel)	205	
dislocation (AISI 4340 Steel)	272	Lee (23)
microvoid (pure iron)	311	
microvoid (AISI 4340 Steel)	338	
405°C peak (0.49% C steel)	387	
405°C peak (AISI 4340 Steel)	405	
MnS interface peak (AISI 4340 Steel)	495	
ferrite grain boundary (pure iron)	310	
microvoids due to cold work (plain carbon steel)	380	Otsubo (24)
cementite and/or microstructure (low alloy weld metal)	480	
inclusion-matrix interface (single-crystal iron)	700	
dislocation interaction (0.012% C steel)	20-150	Habashi (25)
grain boundary, carbide-boundary (")	300-400	

Table 2 : Reactivation Temperatures for Hydrogen Release from Different Traps

Welding Pa	rameters	Shielding Gases
Current	330 amps	Ar- 2%O ₂
Voltage	31 volts	Ar- 2.53%CO ₂ - 0.269%H ₂
Welding speed	5.28 mm/sec	Ar- 2.5%CO ₂ - 0.27%H ₂
Efficiency	0.80	Ar- 2.65%CO ₂ - 0.51%H ₂
Heat Input	1.55 kJ/mm	Ar- 2.7%CO ₂ - 1%H ₂

Table 3 : Welding Parameters and Shielding Gases Used

Table 4 : Specifications of Gas Chromatograph - G 1006 - H_{DM}

Principle	Gas Chromatography
Detector	Thermistor type Thermal Conductivity Detector (TCD)
Analysis Column	Carrier Gas - Ar @ 300 ml/min Molecular Sieves 5A - 5 mm (ID) x 3 m
Reference Specimen	Incorporates a 1 ml Hydrogen measuring internal tube
Analysis Time	up to 5 minutes per specimen
Measured Values	H ₂ value - 0 - 1.999 ml N ₂ value - 0 - 1.999 ml
Measured Error	Within ±0.02 ml for 1 ml

Reading No.	Bypass	Chamber 1	Chamber 2	Chamber 3	Chamber 4	Test Conditions
1	0.979	0.973	0.973	0.974	0.973	$R.T. = 23^{\circ}C$
2	0.979	0.972	0.972	0.973	0.972	Pressure = 760 mm Hg
3	0.979	0.972	0.973	0.972	0.972	Hydrogen at
4	0.979	0.972	0.973	0.972	0.973	STP = 0.979 ml

Table 5 : Results from 45°C Sampler Calibration (AWS Specimen)

Table 6 : Results from 150°C Sampler Calibration (AWS Specimen)

Reading No.	Bypass	Chamber 1	Chamber 2	Chamber 3	Chamber 4	Test Conditions
1	0.979	0.977	0.978	0.979	0.980	$R.T. = 23^{\circ}C$
2	0.978	0.978	0.977	0.981	0.979	Pressure = 760 mm Hg
3	0.979	0.977	0.977	0.981	0.981	Hydrogen at
4	0.979	0.977	0.977	0.981	0.981	STP = 0.979 ml

Table 7 : Results from 150°C Sampler Calibration (JIS Specimen)

Reading No.	Bypass	Chamber 1	Chamber 2	Chamber 3	Chamber 4	Test Conditions
1	0.988	0.990	0.988	0.989	0.988	R.T. = 20.75°C
2	0.988	0.990	0.988	0.989	0.989	Pressure = 760 mm Hg
3	0.988	0.990	0.989	0.988	0.989	Hydrogen at
4	0.989	0.989	0.990	0.989	0.990	STP = 0.988 ml

Table 8 : Weld Hydrogen Content Variation with Shielding Gas Hydrogen Concentration for A514 Steel - Outgassing at 150° C

Shielding Gas	Weld Hydrogen ml/100 gm DM	Mean ml/100 gm DM	Standard Deviation, s
Ar-2%O ₂	2.87,2.8,2.68,2.7	2.77	0.07
Ar-2.65%CO ₂ -0.27%H ₂	7.8, 7.6, 7.9, 7.76	7.79	0.12
Ar-2.53%CO ₂ -0.269%H ₂	8.19, 7.95, 8.25, 8.19	8.14	0.11
Ar-2.65%CO ₂ -0.51%H ₂	11.7, 11.4, 11.8, 11.3	11.57	0.21
Ar-2.7%CO ₂ -1%H ₂	15.1, 16.1, 14.65, 14.87	15.18	0.56

Table 9 : Base Metal Composition Effect on Weld Hydrogen Content

Tamantan	Shinking Con	A514		A36			tuoluo			
	Silleidilig Gas	x	s	N	x	S	N	t value	L.995	L.975
	Ar-2% O ₂	2.77	0.07	4	2.2	0.32	4	2.99	± 3.71	± 2.45
150°C	Ar-2.53% CO ₂ 0.269 % H ₂	8.14	0.11	4	7.1	0.04	4	14.75	± 3.71	± 2.45
	Ar-2.65% CO ₂ 0.51 % H ₂	11.57	0.21	4	10.13	0.33	3	3.9	± 4.03	± 2.57
	Ar-2% O ₂	2.64	0.18	3	1.8	0.09	3	5.7	± 4.60	± 2.78
45°C	Ar-2.53% CO ₂ 0.269 % H ₂	7.7	0.28	4	7.26	0.07	4	2.90	± 3.71	± 2.45
	Ar-2.65% CO ₂ 0.51 % H ₂	11.63	0.25	3	10.4	0.4	4	3.7	± 4.03	± 2.57

Rose Shielding G		150°C			45°C			tualua		
Material	Smelding Gas	x	S	N	x	S	N	t value	1,995	1.975
	Ar-2% O ₂	2.20	0.32	4	1.81	0.09	3	1.73	± 4.03	± 2.57
A36	Ar-2.53% CO ₂ 0.269 % H ₂	7.13	0.03	4	7.26	0.07	4	-2.86	± 3.71	± 2.45
	Ar-2.65% CO ₂ 0.51 % H ₂	10.63	0.33	3	10.4	0.43	4	0.65	± 4.03	± 2.57
	Ar-2% O ₂	2.77	0.07	4	2.64	0.18	3	1.09	± 4.03	± 2.57
A514	Ar-2.53% CO ₂ 0.269 % H ₂	8.14	0.11	4	7.74	0.27	4	2.3	± 3.71	± 2.45
	Ar-2.65% CO ₂ 0.51 % H ₂	11.57	0.21	4	11.63	0.25	3	-0.29	± 4.03	± 2.57

Table 10 : Outgassing Temperature Effect on Weld Hydrogen Content

Base Material	Shielding Gas	H _{DM} After 45°C Outgassing	H _{DM} After 150°C Outgassing	Released Hydrogen	x
	Ar- 2%0 ₂	1.68 1.87 1.89	1.84 2.03 2.06	0.16 0.16 0.17	0.16
A36	Ar- 2.53%CO ₂ 0.269% H ₂	7.20 7.21 7.38 7.26	7.41 7.46 7.61 7.54	0.21 0.25 0.23 0.28	0.24
	Ar- 2.65%CO ₂ 0.51% H ₂	10.40 10.64 10.84 9.71	10.82 11.10 11.27 10.06	0.42 0.46 0.42 0.35	0.41
	Ar- 2%O ₂	2.45 2.60 2.89	2.85 3.05 3.22	0.40 0.45 0.33	0.39
A514	Ar- 2.53%CO ₂ 0.269% H ₂	7.90 8.11 7.50 7.44	8.33 8.53 7.94 7.83	0.43 0.42 0.44 0.39	0.42
	Ar- 2.65%CO ₂ 0.51% H ₂	11.44 11.99 11.47	12.09 12.62 12.16	0.65 0.63 0.69	0.66

Table 11 : Outgassing Temperature Effect on Release of Trapped Hydrogen

Table 12 : Outgassing Temperature Effect on Trapped Hydrogen Release in A36 Steel

Base Material	H _{DM} After R.T. Outgassing	H _{DM} After 45°C Outgassing	H _{DM} After 150°C Outgassing	Trapped H ₂ below 150°C
	7.30	7.34	7.51	0.21
A36	6.82	6.87	7.06	0.24
	7.03	7.08	7.26	0.23
	6.88	6.91	7.10	0.22

Dere Maral		A514			A36		tualua			
condition	x	S	N	x	s	N	t value	1,995	1.975	
As Received	0.037	0.002	4	0.09	0.012	3	8.44	± 4.03	± 2.57	
Degassed Unwelded	0.026	0.003	4	0.02	0.013	4	0.26	± 3.71	± 2.45	

Table 13 : Prior Degassing Effect on As Received Material's Residual Hydrogen Content

Table 14 : Prior Degassing Effect on Weld Hydrogen C	Content
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	Shielding goo	Degassed			Unc	legassed		tuolue		
	Smelding gas	x	S	N	x	S	N	tvarue	1,995	5.975
	Ar-2% O ₂	2.20	0.32	4	3.39	0.29	4	-4.8	± 3.71	± 2.45
A36	Ar-2.53% CO ₂ 0.269 % H ₂	7.13	0.04	4	9.8	0.07	4	-45	± 3.71	± 2.45
	Ar-2.65% CO ₂ 0.51 % H ₂	10.63	0.33	4	11.7	0.25	4	-4.3	± 4.03	± 2.57
	Ar-2% O ₂	2.77	0.07	4	2.29	0.15	3	-4.7	± 4.03	± 2.57
A514	Ar-2.53% CO ₂ 0.269 % H ₂	8.14	0.1	4	8.11	0.14	3	-0.26	± 4.03	± 2.57
	Ar-2.65% CO ₂ 0.51 % H ₂	11.57	0.2	4	11.18	0.38	3	-1.5	± 4.03	± 2.57

Shielding Cas	Bel	t Ground	Surfa	ace Grour	nd	tualua		t	
Smelding Gas	x	S	N	x	S	N	tvalue	1,995	L.975
Ar-2% O ₂	2.20	0.32	4	1.47	0.019	4	3.94	± 3.71	± 2.45
Ar-2.53% CO ₂ 0.269 % H ₂	7.13	0.036	4	6.27	0.17	3	6.108	± 4.03	± 2.57
Аг-2.65% CO ₂ 0.51 % H ₂	10.63	0.328	3	9.62	0.258	4	3.85	± 4.03	± 2.57

Table 15 : Specimen Preparation Effect on Weld Hydrogen Content in A36 Steel

Table 16 : Specimen Length Effect on Weld Hydrogen Content

Page	Shielding Geo	ЛЅ				AWS		tuoluo		
Material	Smelding Gas	x	S	N	x	s	N	t value	L.995	L.975
	Ar-2% O ₂	2.42	0.256	3	2.77	0.074	4	-2.176	± 4.03	± 2.57
A514	Ar-2.53% CO ₂ 0.269 % H ₂	8.72	0.709	4	8.14	0.113	4	1.399	± 3.71	± 2.45
	Ar-2.65% CO ₂ 0.51 % H ₂	12.13	0.56	4	11.57	0.208	4	1.624	± 3.71	± 2.45
	Ar-2% O ₂	1.72	0.03	3	2.2	0.320	4	2.19	± 4.03	± 2.57
A36	Ar-2.53% CO ₂ 0.269 % H ₂	7.49	0.10	4	7.13	0.036	4	-5.87	± 3.71	± 2.45
	Ar-2.65% CO ₂ 0.51 % H ₂	11.12	0.195	4	10.63	0.328	3	-2.08	± 4.03	± 2.57

Shielding Gas	AWS			GR	OOVED		tvolue	t		
Silleiding Gas	x	S	N	x	s	N	t value	1,995	¢.975	
Ar-2% O ₂ Ar-2.53% CO ₂ 0.27 % H ₂	2.77 7.79	0.07	4	2.67 7.15	.15 0.18	4	1.05 5.0	± 3.71 ± 3.71	± 2.45 ± 2.45	
Ar-2.65% CO ₂ 0.51 % H ₂	11.57	0.21	4	11.44	0.69	4	0.25	± 3.71	± 2.45	

Table 17 : Groove Shape Effect on Weld Hydrogen Content in A514 Steel at 150° C

Table 18 : Numerically Developed $\frac{H_r}{H_o}$ for the AWS Specimens

Outgassing Treatment	Hydrogen Diffusivity, $\frac{cm^2}{sec}$	D∆t cm ²	$\frac{H_r}{H_o}$
45°C for 72 hours	TWI Mean D _H 2.179e-06	0.565	0.009
150°C for 6 hours	TWI Mean D _H 2.016e-05	0.435	0.025
45°C for 72 hours	TWI Minimum D _H 1.79e-08	0.005	0.79
150°C for 6 hours	TWI Minimum D _H 1.061e-05	0.23	0.13

CHAPTER 2

WELDABILITY TESTING

2.1 - BACKGROUND

INTRODUCTION

Hydrogen assisted cracking in weldments is avoided in industrial fabrication through the application of an adequate preheating temperature, T_{ph} . Correct application of preheating requires detailed knowledge of a variety of weld metal and base metal metallurgical and geometrical constraint variables and their interaction as a function of welding parameters and ambient temperature condition. The various relationships needed to correctly determine a given weldments T_{ph} are still not completely defined and/or quantified. Thus practical fabrication T_{ph} must be estimated. Presently, there are a variety of methods for estimating preheating temperature both experimentally and analytically.

The usefulness of the T_{ph} prediction schemes is assessed through the application of different weldability tests. Lehigh and Tekken weldability tests are used to assess weld metal and HAZ susceptibility to HAC, respectively. The weldability tests are conducted at T_{ph} from different prediction schemes and are then observed for HAC occurrence. The important issue then is the application of weldability test results to actual weldments. Specially important are the dimensionality and groove geometry and crack initiation and propagation differences between weldability tests and actual weldments.

In this chapter, the above issues are reviewed to gain a global understanding of HAC phenomenon. First, the various factors responsible for HAC occurrence are reviewed followed by an analysis of application of preheating to avoid HAC. Lehigh and weldability tests and weldability testing methodology are then analyzed. Finally, crack propagation behavior in Tekken test specimens is discussed.

HYDROGEN ASSISTED CRACKING

Hydrogen assisted cracking, or cold cracking, is caused by a combination of tensile stress, residual hydrogen, and a susceptible microstructure. Hydrogen assisted cracking generally occurs after the weldment has cooled to temperatures lower than 150°C, and is characterized by an incubation period of up to one or two days after welding. Longer delay times have also been reported.

Hydrogen assisted cracking can occur in either weld metal or HAZ depending on the weld restraint stress, microstructure, yield strength of the weld metal and the location of the stress concentrator notch. Stress concentration and restraint stress across the weld are a function of joint and bead geometry, and weld groove length, respectively.

Hydrogen assisted cracking is also described as cold cracking, toe cracking, root cracking or underbead cracking depending on the crack location. Hydrogen assisted cracking may also occur in the weld metal, but here it is easily controlled by using low carbon and low hydrogen electrodes. For structural steels, weld metal cracking seldom occurs as the HAC susceptibility of HAZ is generally higher due to higher carbon content and carbon equivalent. When welding higher strength steels (greater than 100 ksi, 690 MPa) with matching strength filler metal or using high hydrogen electrodes weld metal cracking commonly occurs.

HAC Susceptible Microstructure

In general, the coarse grained HAZ (GCHAZ) of a weld has the most HAC susceptible microstructure. In addition, hydrogen tends to concentrate in the CGHAZ. According to Granjon³³, weld metal generally has a lower carbon content and hence higher Ac₃ temperature than the base metal. He postulated that austenite transformation in weld metal occurs prior to that of the HAZ. Hydrogen is thus rejected into the austenite in the coarse grained HAZ next to the fusion zone from the already transformed weld metal, since solubility of hydrogen in ferritic transformation products is lower than in austenite. The HAZ austenite transformation on cooling

proceeds from the base metal side of the HAZ towards the fusion zone. It begins on the periphery of the austenite containing region of HAZ, as the region nearest to the base metal initiates austenite decomposition first. Austenite transformation then proceeds toward the weld metal, whereas, hydrogen diffusion takes place from weld metal to HAZ. Therefore, the last region of the HAZ to transform (CGHAZ) is already bounded by previously transformed material and, as a result, has the highest concentration of hydrogen, Figure 12. Hence HAC generally occurs in coarse grained HAZ, but when yield strength of the weld metal is high, i.e., 600 MPa or more, weld metal cracking is also a possibility³⁴

The susceptibility of the HAZ microstructure is strongly dependent on the steel's chemistry (usually assessed using carbon equivalency), carbon content, peak temperature attained during welding, and the cooling rate through the austenite transformation temperature range. Micros-tructures listed in order of decreasing crack susceptibility are listed below ³⁵

- 1. Internally twinned martensite (found in medium and high carbon steel),
- 2. Martensite without internal twins (found in low carbon steel),
- 3. Mixtures of martensite and softer phase such as bainite or ferrite and
- 4. Non-martensitic structures.

The higher the HAZ hardness the more susceptible is the microstructure to HAC. The hardness of the HAZ microstructure depends primarily on carbon content, the peak temperature in the HAZ and the cooling rate. The carbon content decides the martensite-start temperature Ms as well and, hence, the hardness and brittleness of martensite formed. Peak temperature controls the HAZ grain size and chemical homogeneity of the prior austenite grains and thus influences transformation kinetics and hardness of the HAZ microstructure after transformation. The cooling rate in the HAZ is affected by many factors, including joint geometry, heat input, welding process, T_{ph}, type of preheating (local or uniform) and plate thickness³⁶. It has been observed by several researchers that nonmetallic inclusions affect HAC susceptibility of the HAZ, but the exact role played by non metallic inclusion is still being debated. Early theory postulated that sulfides acted as traps or sinks for diffusible hydrogen and thus reduced its detrimental effects in the HAZ³⁷. This was supported by the observation that very low-sulphur steels show increased HAC susceptibility. Kikuta and co-workers have also supported this observation³⁸. Another significant observation is that sulfides and other inclusions act as nucleation sites for intragranular ferrite transformation from austenite, thereby precluding the formation of hard and brittle martensitic structures³⁹. These observations reinforce the concept that the presence of inclusions are beneficial and result in lower susceptibility towards HAC. In contradiction to the above hypothesis, it has been observed by Suzuki⁴⁰ that low sulphur contents do not adversely affect HAC susceptibility.

Hydrogen Content of the Weld Metal

Hydrogen in the weld metal is derived from hydrogenous chemical compounds that are dissociated in the arc column. Hydrogen is absorbed by the weld pool from the arc atmosphere according to Sievert's law. The principal sources of hydrogen in welding consumables are¹:

- Moisture in the coating of Shielded Metal Arc Welding (SMAW) electrodes, submerged-arc welding flux, and in the flux used in flux cored wires,
- 2. Any hydrogenous compounds in the coating or flux,
- 3. Oil, dirt, and grease either on the surface or trapped in the surface layers of welding wires and electrode core wires, including electrode drawing agents, and
- Hydrated oxide, e.g., rust on the surface of welding wires.

The principal sources of hydrogen from the material to be welded are¹:

 Oil, grease, dirt, paint, rust, etc., on the surface and adjacent to the weld preparation that can break down to produce hydrogen in the arc atmosphere, and Degreasing fluids used to clean surfaces before welding may likewise dissociate to produce hydrogen.

The hydrogen that is potentially available from the welding consumables is different than the actual weld hydrogen because not all the potentially available hydrogen is absorbed by the weld pool; although, the higher the potential hydrogen the higher will be the actual weld hydrogen. Typical weld hydrogen levels for various welding processes are shown in Table 19³⁴.

Hydrogen continuously evolves out of the weld metal as the solidified weld metal cools, thus hydrogen content at the temperature of cracking is different than initial weld hydrogen content. Since HAC generally occurs at temperatures lower than 150° C, the residual diffusible hydrogen at 100° C, H_{r100}, is better related with critical stress for hydrogen cracking than initial hydrogen²⁸. Even at the same initial weld hydrogen contents, H_{r100} will be different for two welds cooling at different rates because of variations in heat input and other influencing factors, such as local versus global preheat.

Weld Induced Stress Concentration

The residual restraint stresses associated with weld shrinkage concentrate dissolved hydrogen as hydrogen diffuses to regions of high tensile stress concentration along stress gradients. The severity of residual stresses acting on the weld is dependent on stress concentration and restraint intensity. Localized plastic strains occur because of the stress concentration effect even where the average residual stress level is below yield. Stress concentration and restraint level across the weld are analyzed in the following sections.

Stress Concentration Factor

The initiation of HAC in a weld depends on the localized stresses at the crack initiation location. It is generally observed in weldability tests that HAC initiates from a weld induced notch such as, weld root and toe. The magnitude of stress concentration due to a geometrical

notch at the root of the weld depends on various factors such as⁴¹: Throat depth, Root gap, Global joint geometry, Eccentricity of the weld about the neutral axis, Local joint geometry in the vicinity of notch root, Plate thickness and Concavity of the weld metal surface.

The stress concentration factor value, K_t , increases with a decrease in root angle and with an increase in both plate thickness and throat depth⁴². The stress concentration at the notch root gives rise to local stresses and strains which increase the tendency to HAC. Values of K_t for various joint geometries encountered in weldability tests have been determined by experimentation as well as by Finite Element Analysis, Figure 13. The local stress at a weld notch, σ_{local} , is given by a product of restraint and the K_t^{43} ;

$$\sigma_{\text{local}} = \sigma_{\text{restraint}} \quad K_{\text{t}} \tag{3}$$

where $\sigma_{restraint}$ is the mean restraint stress acting on the weld.

Restraint Intensity

Weld metal is subjected to restraint stresses as it cools due to the ensuing shrinkage and the forces opposing this shrinkage. The mean stress, σ_w (MPa), acting on the weld metal is dependent on the joint restraint intensity, R_F (MPa), where the R_F of a joint is "the magnitude of force per unit weld length that is necessary to displace elastically the root gap by a unit length in the direction perpendicular to the weld."⁴⁴ Restraint intensity is a measure of restraint offered by a particular welding joint. It is a function of restraining length, plate thickness, root gap and joint geometry. It should be noted that R_F is a value defined before welding and does not incorporate the shape of groove or the size of weld metal⁴⁵. The σ_w stress acting on the weld is given by following relations for a single-pass root weld.

 $\sigma_{\rm w}$ (butt) = 0.050 R_F, For R_F $\leq \sigma_{\rm y}$ (4)

 σ_w (butt) = σ_y + 0.0025 (R_F - 20 σ_y), For R_F >= 20 σ_y (5)

where σ_v (MPa) = yield strength of weld metal

Therefore, the higher the R_F the higher the mean stress acting on the weld for a given σ_v .

Restraint intensity, R_F (MPa), is a function of plate thickness, h (mm), and coefficient of restraint intensity, r_f^{46} . Table 20 shows typical restraint intensity values for different types of welding joints⁴⁷. The R_F mentioned above deals implicitly with transverse restraint, because transverse residual stress causes root cracking and toe cracking, whereas longitudinal residual stress in a weld is always as high as the yield strength level, independent of the restraint intensity⁴⁸. It also should be noted that the restraint intensity is relevant only for the first pass, but not in multipass welding, because a weld is completely restrained in the transverse direction only after root pass welding⁴⁸.

MEASURES TO AVOID HAC

Hydrogen assisted cracking in weldments can be avoided by:

- 1. Controlling the hydrogen level,
- 2. Controlling the residual tensile stress level,
- 3. Avoiding stress concentrations, and
- 4. Lowering the susceptible microstructure content.

The weld metal hydrogen level can be controlled by using low hydrogen consumables. The potential hydrogen in a given set of consumables can be lowered by baking the consumables before welding, as consumables experience an increase in moisture content during storage. The weld hydrogen content also depends on the welding process and on welding process variables. Welding process control, hence, plays a crucial part in deciding weld hydrogen content. Tensile stress across the weld can be reduced by controlling joint misalignment and/or by selecting a joint that minimizes R_F. Theoretically, one would like to minimize the formation of HAC-susceptible weld metal and HAZ microstructures. However, these factors are essentially fixed once the selection of base metal and filler metal is made for a selected welding condition. In actual practice, preheating is the most widely used method of avoiding HAC.

PREVENTION OF HAC BY PREHEATING

Preheating decreases the weld cooling rate, which increases hydrogen diffusivity time and can also decrease the formation of HAC susceptible microstructures. Preheating a weldment significantly decreases the 1500 to 100° C cooling rate, $t_{15/1}$, thus lowering the amount of H_{R100} . Preheating also decreases the 800 to 500° C cooling rate, $t_{8/5}$, which determines austenitic transformation products. Lower cooling rates in the transformation range means that less brittle microstructure formation is favored. Preheating also lowers the magnitude of residual stresses⁴⁹. The practical fabrication question which arises is not whether to preheat, but what is the minimum T_{ph} required to assure resistance to HAC.

Selection of the most suitable T_{ph} is important from the joint strength, toughness, distortion, residual stress and economy point of view. Unnecessarily high localized T_{ph} will result in lower joint strength and lower toughness and will be uneconomical. Whereas, lower than the required T_{ph} leads to HAC. Thus knowledge of the optimum T_{ph} for a given welding situation is essential. Various guidelines, experimental techniques and predictive schemes have been defined by governing agencies throughout the world. Some of these procedures and preheating temperature prediction schemes are discussed below.

Fabricators have several choices when selecting a T_{ph} . These choices are:

- 1. Use preheating guidelines from welding codes,
- 2. Perform weldability tests and use the results as preheating guidelines, and
- Use complex, empirically-based, calculation schemes based on material chemistry to predict a safe T_{ph}.

Preheating Guidelines from AWS Welding Codes

Generally acceptable generic preheating guidelines have been developed and published by American Welding Society (AWS) based upon years of generally successful field experience. The AWS Structural Welding Code D1.1⁴⁹ and AASHTO/AWS Code D1.5⁵⁰ specify minimum preheating temperatures for various thicknesses of bridge steel. These predictions are for general welding conditions and do not take into account specific weld hydrogen content and restraint across the weld. For welding conditions other than those defined in the welding codes, use of hydrogen control method or hardness method is recommended.

Weldability Tests

Weldability tests have been devised as experimental methods to evaluate the complex interactions taking place in weldments. One needs a method of predicting the effects of welding on joint integrity for a given joint configuration, plate size, chemistry, consumable combination, welding technique, and parameter set. This is not a trivial task, particularly when one wants to utilize some type of subsize "weldability" test. These types of tests have, historically, been used to study such variables as steel composition, electrode type, heat input, T_{ph} and degree of restraint responsible for cracking^{51,52}.

A major practical reason for conducting weldability testing is to determine the resistance of the weld and HAZ to HAC, or cold cracking, and more than 42 different cold cracking weldability tests have been developed³⁶. There is no single universally acceptable test because of the complex nature of HAC and interactive factors that cause HAC. The objective of these tests is to assess the susceptibility of a given steel to HAC.

Weldability tests can be separated into two categories, self-restraining and externally loaded. Self-restrained tests are those in which the stresses acting on the weldment result solely from thermal contraction of the cooling weld, phase transformations, and inherent joint restraint intensity. In externally loaded tests, the stress is applied externally during or shortly after completion of welding. Only two specific self-restraint weldability tests will be discussed here: the Lehigh restraint cracking test and the Tekken restraint test.

Lehigh Restraint Test

The Lehigh weldability test was developed by Stout, et al, in 1945⁵¹. The Lehigh restraint test specimen contains a U-groove geometry, which is supposed to represent actual weld geometry, and is sufficiently large so that the degree of restraint and cooling rate are "representative" of actual weldments, Figure 14. In this test, a weld bead is deposited in the 5-inch-long (127mm) groove, that simulates the root pass at mid thickness in a U-groove weldment in the presence of a mechanical notch. If the material is susceptible to HAC, cracking generally occurs in the weld metal at the weld root and propagates into the weld metal. Konkol, et al ³⁶ observed HAZ cracking in Lehigh specimens. This test also assesses hot cracking susceptibility so the exact cracking mechanism must be decided after metallographic examination of cross sections of the weld³⁶. Finite element analysis results combined with experimentation indicate that the coefficient of restraint, rf, is 430.h N/mm³ for Lehigh specimen without saw cuts, where h is equal to the plate thickness⁵³. In real structures (such as bridges, frames and vessels), the observed restraint coefficient is around 394.h N/mm^{347, 53}. These restraint coefficient values suggest that the Lehigh weldability specimen closely simulates the weld metal response to HAC of real structures.

Tekken Test (Oblique y-groove test)

The Tekken restraint test was designed to primarily assess HAC susceptibility of the HAZ. It exhibits severe restraint conditions that are approximately twice what is typically seen in bridge fabrication weldments. The Tekken test has been widely used in Japan since 1960 to study HAZ cold cracking susceptibility of steels⁵⁴. It is used as a standard test to qualify newly developed steels⁵⁵. The groove length in Tekken specimen is only 80mm, Figure 15, and hence, gives rise to a very high restraint coefficient value of 690.h N/mm³, where h is the thickness in mm⁴⁷. In Tekken test specimen, because of the asymmetrical y-groove, higher stress concentration is located near the HAZ region of the weld root on the double bevel side of the specimen. This

forces HAC initiation from this side of the weld notch root depending on the weld root shape. The Tekken test is conducted in a manner similar to the Lehigh restraint test.

Use of Weldability Test To Determine Preheating Temperature

Weldability specimens are welded with a constant set of welding conditions as a function of T_{ph} . They are then allowed to set at room temperature (RT) for a minimum of 48 hours prior to being examined for weldment cracking. The specimens can then be separated into cracked and non-cracked groupings and the minimum T_{ph} to prevent weldability specimen cracking can then be determined for the actual plate (type), consumables, welding technique and parameters expected to be used in actual fabrication.

The Lehigh specimen configuration dictates that it be used to assess weld metal susceptibility to HAC while the Tekken specimen configuration dictates it be used to assess HAZ susceptibility to HAC. Thus the recommended experimental methodology for determining butt weld preheating requirements is to test both Lehigh and Tekken specimens.

It is recommended that the Lehigh test results be used directly to determine weld metal preheating requirements. Thus a required minimum T_{ph} would be determined from the maximum preheating temperature where Lehigh specimen cracking was observed. Note that the Lehigh specimen needs to be used without any restraint-reducing saw cuts for this application, as this configuration yields restraint factors comparable to typical fabrication weldments. In addition, one needs to use a bottom slot versus the standard rounded groove for heat inputs above about 1.7kJ/mm to prevent melt-through.

HAZ preheating requirements would be determined from Tekken test results. The actual fabrication preheating requirement may, however, not necessarily be the maximum preheating temperature exhibiting Tekken specimen cracking due to severe restraint encountered in a Tekken test. Yurioka has developed an empirical relationship between Tekken test temperature, weld metal yield strength, type of welding and fabrication preheating requirements. This is a

preheating "forgiveness" factor for low tensile strength materials and low constraint weldments. This relationship is presented in Figure 16.

EFFECT OF WELDABILITY SPECIMEN SIZE ON HAC SUSCEPTIBILITY

The Tekken and Lehigh tests use the same thickness plate as used in actual weldments but, in general, the plate dimensions are much smaller than encountered in service. Since weldability test specimens are small, their cooling behavior and the heat distribution may be significantly different than the larger plates encountered in heavy fabrication. The higher heat extraction effect associated with large plates used in fabrication may result in faster weld cooling rates than those experienced by the small weldability specimens. This would be expected to result in higher H_r content in actual welding, causing an increased tendency to HAC in actual fabrication than that predicted by the weldability tests. The magnitude of potential increase in weld hydrogen content in actual weldments needs to be quantified.

PREDICTION OF PREHEATING USING THE NOMOGRAPHS

Two nomographic methods for prediction of preheating will be discussed herein. One method has been developed by TWI and published as British Standard BS5135:1974⁵⁶. The other has been developed by Yurioka. The British Standard method is based on the controlled thermal severity (CTS) test and simulates a fillet weld. However, Yurioka and Suzuki suggest that its use in assessing the potential for toe cracking in a butt weld with reinforcement may be justified based on the CTS test fillet toe having a similar stress concentration as a typical multipass butt weld toe⁴⁸. The use of this preheat temperature prediction methodology is discussed in depth in the British Standard. It is not directly applicable to butt welds, and is not presently applicable to low carbon microalloyed steels. Cottrell has recently published suggested changes which broaden the model's applicability and take into account effect of sulphur on hardenability and hence HAC susceptibility⁵⁷. Cottrell's method exhibited better correlation with actual CTS test results, and hence this method is preferred over the BS method.

Yurioka has also recently published a nomographic methodology for predicting expected Tekken test cracking temperature as a function of plate chemistry and residual hydrogen level^{55,58}. Yurioka's nomographic method is based on Tekken test results, and predicts the preheating test temperature that separates cracked from uncracked Tekken tests. The master nomograph, Figure 17, presents the minimum preheating needed to assure no cracking in a Tekken test welded at a temperature of 10°C, a heat input of 1.7kJ/mm, and an initial weld hydrogen content of 5 ml/100gm of deposited metal as per the International Institute of Welding (IIW) method.

Application of Yurioka's Scheme for Practical Welding

In Japan, the Tekken test is conducted under a standard set of welding conditions which are 1.7kJ/mm heat input, weld hydrogen content of 5 ml/100gm weld metal and 10°C ambient temperature. In practice, however, the welding conditions in a Tekken weldability test may deviate from the standard welding conditions to simulate actual welding. As the tendency towards cold cracking is influenced by the welding heat input and weld metal hydrogen, the effect of this variation from the standard conditions on HAC should be taken into account. This is done on an empirical basis as shown in Figures 18(a) and 18(b). In these Figures, the extent of deviation from the standard conditions is represented not by the critical preheating temperature, but indirectly by the difference in the carbon equivalent number (CEN).

The effect of chemical composition, i.e., alloying elements, is taken into account by using "carbon equivalent," that is closely related to the hardenability of a given steel. Multiple carbon equivalent formulas have been proposed, see Table 21. Yurioka's proposed carbon equivalent formula, CEN, is based on the examination of weldability test results for a wide range of steel composition.

By substituting CEN + Δ CEN for different heat inputs and hydrogen contents into the abscissa in Figure 17, the preheating temperature, T_{ph}, for a y-groove restraint test can be

predicted for arbitrary conditions of CEN, plate thickness, heat-input and weld metal hydrogen. When applying the preheating temperatures from oblique y-groove Tekken test to actual welding situation it should be remembered that Tekken test is much more stringent than other cold crack-ing tests, for the following reasons³⁴:

- 1. Restraint intensity is extremely high because of slit welding,
- 2. Very acute notch at the weld root,
- 3. Increased HAZ hardness because of short bead and single pass welding and
- Less evolution of hydrogen from the single pass weld root because of no post heating due to subsequent passes.

Therefore, the critical T_{ph} s obtained in the y-groove Tekken are not employed directly in actual welding fabrication. Figure 16 shows the reduction in actual preheating temperature as a function of weld metal yield strength. In the case of fillet welding, the plate thickness should be that for the thicker plate.

Calculation Schemes for Predicting Preheating Temperatures

Calculation schemes for predicting required T_{ph} have been developed for both weld metal and HAZ. These preheating methodologies allow prediction of preheating without the need for fabrication and analysis of weldability specimens. The fundamental idea behind various predictive schemes is to predict reasonably and accurately the most optimum $T_{ph}s$ for given welding conditions which will avoid HAC in weld metal as well as HAZ. Various calculation approaches to T_{ph} calculation were also proposed by Suzuki^{44, 59-62}. Similar T_{ph} prediction methodology was proposed by Yurioka, et al⁶. The prediction methodology by Yurioka for avoiding HAZ hydrogen assisted cracking is based on weldability test results.

Most of the preheat prediction schemes estimate T_{ph} for avoiding HAC in HAZ. Whereas, for avoiding HAC in weld metal only a few T_{ph} prediction scheme are available. The reason
behind this is the rule of thumb that the T_{ph} which avoids HAC in the HAZ also avoids HAC in the weld metal. This observation is valid only for conventional steels. With the recent development of low-carbon steels that are resistant to HAC, the HAC occurrence has shifted back to the weld metal. It is then necessary to develop a T_{ph} prediction scheme for avoiding HAC in weld metal. To the author's knowledge, only Nippon steel is working in this direction. In the mean time, any existing prediction scheme for weld metal HAC should be evaluated for its applicability. The only existing prediction method is for avoiding HAC in multi-pass weld metals developed by Nippon steel researchers^{34, 63-65}. This prediction scheme relies on tensile strength of the weld metal, weld hydrogen content and the throat depth. For its applicability to single pass welding situation this procedure was modified as follows. The weld throat for the welding conditions in this study was smaller than the applicable range of this scheme. The lowest weld throat to which this method is applicable is 10 mm and it was used in the present study even though the weld throat in the experimental study was smaller than 10 mm.

CRACK PROPAGATION BEHAVIOR IN TEKKEN TEST SPECIMENS

Three types of crack propagation behavior have been observed in the Tekken specimens:

- 1) Type 1: The crack propagates all the way in the HAZ, Figure 19(a).
- Type 2: The crack propagates some distance into the HAZ and then changes its crack propagation direction into the weld metal, Figure 19(b).
- 3) Type 3: The crack initiates and propagates into the weld metal all the way, Figure 19(c).

The last mode of crack initiation and propagation is not expected generally. It occurs when the root gap becomes 2.5 mm or larger or when weld metal is highly susceptible to HAC⁶⁶⁻⁶⁸. An additional reason is the shape of the notch on the double-bevel side. If this notch is not sharp then crack initiation occurs from the single bevel notch side. With careful control over the bead profile and location, the last mode of crack propagation can be avoided. It is not clear why the crack propagation is not always as per Type 1. The most commonly observed Type 2 crack propagation behavior could be because of microstructural changes or mechanical force changes. It is thus necessary to identify the exact reason behind this type of crack propagation behavior.

Simple reasoning tells us that Type 2 crack propagation can not be caused by sudden microstructural changes. This suggests that changes in mechanical conditions/forces operating on the crack probably govern Type 2 propagation behavior. After careful analysis of crack propagation modes in the literature, it was observed that the crack propagation in the initial stages most commonly occurs at an angle with respect to the y-axis, Figure 19(b). This also means that the crack propagation occurs at an angle with respect to the tensile residual stresses. If the crack is at an angle to applied/residual stresses then it experiences both Mode I, tensile, and Mode II, shear, loading. In such cases, the crack is under mixed mode loading situation.

According to Chen and Nakasa⁶⁹, since HAC is controlled by mechanical factors, such as stress or stress intensity factor, and chemical factors, such as hydrogen concentration, it is possible that the superposition of Mode II loading on Mode I loading affects the interaction and influences HAC phenomenon. crack propagation under mixed mode loading conditions has been studied by various researchers⁷⁰⁻⁷⁵ The results from these studies have proved that, under mixed mode loading, the crack propagation angle changes as Mode II contribution increases. After a certain threshold, the mode II type loading governs crack propagation. At this stage, the crack changes its propagation direction. This is also called "kinking". In many other cases, crack branching also occurs as a result of mixed mode loading conditions. It is thus necessary to study the mechanism of crack propagation in Tekken test specimens under the influence of Mode I and II loading.

OBJECTIVES OF THIS STUDY

From the literature survey, it was clear that some important issues from HAC viewpoint were not addressed in detail. HAC occurrence in steels can be prevented provided that factors responsible for it are avoided. In spite of extensive research in this area, guaranteed preventive measures are still lacking. A prime example is the recent failure of HSLA welds on the Sea-Wolf submarine. It demonstrates the limited understanding of the factors avoiding HAC. Developing steels with lower carbon contents and carbon equivalents does not always guarantee immunity from HAC occurrence. Until recently, preheating and post weld heat treatment were the most widely used industrial methods for high strength steels, as extra low hydrogen electrodes, $H_{DM} < 2$ were not available. With the development of extra low hydrogen consumables an additional preventive measure can be employed. The specified weld hydrogen contents will vary in actual welding practice due to hydrogen pick up during welding. In such cases, T_{ph} needs to be calculated for the welding procedure used. Hence, T_{ph} prediction schemes are still necessary. It is also necessary to evaluate these prediction methods for their advantages, drawbacks and limitations. The overall objective of weldability testing program was to analyze different issues which affect HAC. The specific objectives were defined to address each important issue. The specific objectives of this study were :

- To determine heat affected zone (HAZ) and the weld metal HAC susceptibility of ASTM A514 steel using Tekken and Lehigh weldability test specimens, respectively.
- To assess the applicability of different preheating temperature prediction methods to avoid HAC occurrence in HAZ and weld metal and their applicability to practical joints.
- To locate the source and time of crack initiation using acoustic emission monitoring (AEM) in Lehigh and Tekken test specimens.
- To analyze the crack propagation behavior in Tekken test specimens by stress intensity factor calculations using the finite element method.

2.2 - EXPERIMENTAL

INTRODUCTION

In this study, Lehigh and Tekken weldability tests were performed with different weld hydrogen contents and at different preheating temperatures calculated from Cottrell and Yurioka's method. The weld metal hydrogen content was varied by using shielding gases containing different amounts of hydrogen. In addition, acoustic emission monitoring (AEM) was carried out to study crack initiation location and to study crack propagation behavior as a function of weld hydrogen content and T_{ph}. Crack initiation location was analyzed using metallography. Fracture modes in cracked specimens were studied in detail using metallography. The crack propagation behavior in Tekken specimen was analyzed by stress intensity factor calculations using the finite element method

CONSUMABLES

Base Material Description :

An ASTM A514 Grade B quenched and tempered steel was used throughout this study. The chemical composition, prior heat treatment and the mechanical properties are given in Table 22. The microstructure consisted of tempered martensite and is shown in Figure 20(a).

Filler Metal Description :

Solid wire conforming to ER120-S1 of 1.6 mm diameter was used. The chemical composition and mechanical properties of the welds deposited using this filler wire are given in Table 22. The microstructure of the as deposited weld metal was mostly acicular ferrite type, Figure 20(b).

Shielding Gases :

Four shielding gases containing different concentrations of hydrogen were used in this

study to deposit welds with different weld hydrogen contents. The compositions of shielding gases is given in Table 23. The welding parameters shown in Table 23 were maintained for different shielding gasses.

WELDING DETAILS

Robotic pulsed GMAW was performed to get consistent and reproducible welds using the welding parameters shown in Table 23. The weld hydrogen content corresponding to these shielding gases and welding parameters was determined according to the AWS A4.3-86 procedure, Table 23. With the knowledge of weld hydrogen, heat input and chemical composition, T_{ph} from Cottrell's, AWS D1.1 and Yurioka's prediction schemes were calculated for the Tekken specimens. In case of Lehigh specimens, Yurioka's prediction scheme for avoiding HAC in multi-pass welding was modified and T_{ph's} for different weld hydrogen contents were determined. The weldability tests were conducted at these preheating temperatures.

LEHIGH WELDABILITY TESTS

Lehigh weldability specimens were used to assess the weld metal HAC susceptibility. The specimens were machined and the welding groove was cleaned using ethanol and acetone and was blow dried prior to welding in case of room temperature preheat, and prior to preheating in case of preheated specimens. The specimens were uniformly preheated. Thermocouples were attached to the specimens to monitor the preheating temperature and to record the thermal history.

A single-pass test weld was performed. Immediately after welding, the specimen was kept on an asbestos sheet and acoustic emission (AE) sensors were attached as shown in Figure 21. Acoustic emission monitoring was initiated immediately. In the case of preheated specimens, the temperature in the vicinity of AE sensors was allowed to drop to 125°C before AEM was started. Acoustic emission monitoring was continued until the specimens cracked completely or for 72 hours. This procedure was repeated for different amounts of weld hydrogen.

Thermal Cycle Measurements :

Thermal history for each Lehigh test specimen was monitored by capacitor discharge welding (Alumel -Chromel) K-type thermocouple in arrays as shown in Figure 22. The preheating temperature was monitored and welding was started as soon as the preheating temperature reached the desired temperature. Thermal cycles were recorded for further analysis using a data acquisition system.

Acoustic Emission Monitoring

An AET 5500 computer-based AE system developed by Hartford Steam Boiler Inspection Technologies was used for AE characterization and linear source location. This system in dual channel mode can be used for Linear source location. The AE set up is shown in Figure 21. The AE sensors were omni-directional and differential type with a resonant frequency of 375 KHz and have a sensitivity of better than -70 dB referred to 1V/ Bar. The diameter of the transducers is 25.4 mm. Received AE signals were preamplified using 60 dB preamplifiers with a flat frequency response between 1 KHz and 2 MHz. The filtered signals receive post amplification of 10 dB in a Signal Processing Unit (SPU). The total system gain was 70 dB except for Tekken specimens for which the total system gain was 60 dB. Threshold voltage was set at 1 V fixed. Thus, amplified signals over a 1 V fixed threshold were detected as valid events. Ring Down Event Module (REM) and Peak Amplitude Rise Time Module (ARM) are used to characterize the incoming preamplified signal in the SPU. The various AE parameters were set according to those listed in Table 24.

Linear Source Location :

The AET 5500 system contains a Time Differential Module (TDM) which is used for Linear source location. The TDM compares the time of arrival of an AE event at two or more sensors. The difference in the time of arrival of a signal at one sensor and the time of arrival of the same signal at another sensor is referred to as Δt .

A simple two sensor array set up for linear location locates the source of emissions along the straight line between the two sensors. The manufacturers procedure for linear source location was used. The accuracy of source location is dependent on the maximum Δt value. Maximum Δt value occurs when an event outside the array excites sensor # 1, travels the entire length of the specimen, and then excites sensor # 2. The maximum Δt thus measured can then be verified through the use of the following formula :

Velocity = Distance (measured)/ (Max Δt x Locator Clock Period) (6)

A Raleigh wave velocity of 3140 m/sec was used to verify the maximum ΔT for the array shown in Figure 21. The maximum ΔT for Lehigh weldability test specimen from the above equation was found to be 95 msecs for LOCATOR CLOCK PERIOD of 1000 ns and distance of 305 mm. This matched accurately with the maximum ΔT determined during calibration.

Calibration for Linear Source Location :

Calibration of the system for linear source location was carried out as per the manufacturers instructions. The AEM sensors were arranged in a logical array as shown in Figure 21 on the weld line but outside the welding groove. A pulsar sensor with a pulsing frequency of 1 Hz was used and placed outside the array on the welding line. The distance between the two AEM sensors was noted and the calibration was carried out for 5 minutes with sensors in this position. At the end of the calibration, the averaged ΔT was noted and compared with the one calculated from equation 1. This procedure was repeated for different locations of AEM sensors and the values were compared with those determined from equation 1. This procedure was also used to calculate maximum ΔT . After the determination of the maximum ΔT the source location accuracy was checked by breaking a pencil lead at different places in the linear array and verifying the location displayed with the actual location. If the acoustic emission event occurs inside the array then the location is determined by the following formula :

Location = Maximum Δt - Sample Δt x Sensor Spacing / (2 x Max ΔT) (7)

This verification procedure gave confidence in linear source location capability of AET 5500. Additional AE features such as ring down counts, event duration, rise time, slope, energy and peak amplitude were also recorded during each test.

TEKKEN WELDABILITY TESTING

Tekken weldability test specimens were used to assess the HAZ susceptibility to HAC using the same procedure as described for Lehigh weldability tests. Test welding of Tekken test specimens was not possible with the established welding parameters when using shielding gases containing 0.269% H_2 and 0.51% H_2 . Even after different approaches and trials, successful welding of Tekken tests using these two shielding gases could not be achieved. Therefore, Tekken test results with shielding gases containing no hydrogen and containing 1% H_2 were only considered in this study.

In addition, the AEM procedure was slightly different than that for the Lehigh test specimens. In this case, the AE sensors were placed on the specimen top surface instead of on the plate edges. This lowered the linear source location accuracy by the diameter of the AE sensor. Additional weld cooling data for Tekken test specimens was obtained using the graphs provided by Kasuya based on a 3-dimensional heat transfer equation⁷⁶.

CRACKING RATIO DETERMINATION

The weldability test specimens were sectioned after cracking or after a minimum waiting period of 5 days after welding. Longitudinal crack ratios were calculated by dividing the longitudinal crack length by the weld bead length. Sectional crack ratios were also determined by cross sectioning the specimens at five locations. The height of the crack divided by the weld throat was used to express the sectional crack ratio.

HARDNESS MEASUREMENTS

Knoop micro hardness measurements were performed with a Leco hardness tester using a 500 gm load. The HAZ and weld metal hardness as a function of T_{ph} was measured in Lehigh and Tekken specimens.

METALLOGRAPHY

Optical microscopy was performed using Nikon epiphot optical microscope. Crack initiation location was observed for both Lehigh and Tekken weldability specimens. The crack initiation site, if in the HAZ or weld metal, was then located.

FRACTOGRAPHY

Categorization of fracture modes was performed by observing the type of fracture at different locations along the weld length using a Zeiss scanning electron microscope at 10 and 30 kV. Detailed examination of fracture surface was carried at the weld notch root locations.

CRACK PROPAGATION STUDIES

To study the crack propagation behavior from Tekken and Lehigh tests, simple calculations using the FEM were performed. Linear elastic fracture mechanics capability of ANSYS, a commercial Finite Element Analysis (FEA) program, was used. This program uses nodal displacements to determine stress intensity factors, K, for a given crack. The general procedure for K calculations is described below.

The displacements and stresses in the vicinity of the crack-tip/front are expressed in terms of K using the equations proposed by Paris and Sih⁷⁷. These equations, after modification, yield K at the crack face in terms of incremental displacements. The FE results are compared with theoretical equations and K is extracted. The approximate crack face displacements are obtained through a linear curve fit of nodal displacements. From these values, K is then calculated. In the ANSYS program, all these calculations are performed using a single command. The stress singularity is simulated using singular elements and a fan shaped mesh at the stress concentration

location. Various benchmarks have established the applicability of ANSYS to K calculations, and hence it can also be used to calculate K for different crack configurations. The exact procedure to calculate K for cracks originating from the root notch is outlined below.

Numerical Procedure

The crack propagation behavior in Tekken specimens was analyzed. These calculations were made under the assumption that a small crack has already originated from the double beveled side of the Tekken specimen. The K calculations were then performed by varying the tensile stresses applied on the specimen.

A two-dimensional Tekken specimen as shown in Figure 23 was assumed. The model assumed a pre-existing crack emanating from the double beveled side notch root. The nucleated crack then propagates under the action of tensile residual stresses. To simulate the crack propagation under tensile stresses, the Tekken test specimen was subjected to external tensile stresses on one of the specimen edges while restraining the other edge, Figure 23. The resulting stress distribution is different than the expected residual stress distribution in a Tekken test specimen. But for the K calculations, it does not matter if the applied stress type is different or not as it will only cause a change in the magnitude of resulting stress intensification but will not affect the K calculations. Since the final objective was to calculate the K for different crack configurations, this simple model was expected to yield useful results.

The pre-existing crack configuration was varied in the following manner. For a given crack angle, cracks of different lengths were introduced and K calculations were performed. In addition, the crack angle itself was varied from 15, 30 and 45 degrees. The crack angle was defined by the angle between the crack plane and the Y-axis, Figure 23. The K calculations depend on the nodal displacement values, and hence on the mesh fineness. Therefore, the mesh density was kept constant for different crack configurations. The final K calculations were independent of mesh density. After application of tensile stresses, K₁, K in mode I, and K₁₁, K in mode II, calcu-

lations were performed. K_I and K_{II} values as a function different crack-lengths for different crack angles were then calculated.

Similar calculations were performed on cracks emanating from Lehigh specimens. In the case of Lehigh specimens, the crack propagation occurs exclusively in the weld metal. Therefore, the crack plane was parallel to the Y-axis and into the weld metal. This also meant that the crack plane was perpendicular to the applied tensile stresses.

2.3 - RESULTS

LEHIGH WELDABILITY TEST RESULTS

Observed critical T_{ph} along with longitudinal cracking ratio and AE data is listed in Table 25 for Lehigh weldability test specimens. The sectional cracking ratio, i.e. the height of crack to throat depth ratio, was always equal to 1 in all these cases.

Root Penetration

After cross sectioning the specimens, it was observed that root penetration varied significantly along the weld length for most of the specimens, Figure 24(a). For the initial 20 mm or so the penetration was lowest and for the remaining length it was mostly constant. But the root penetration was full in some specimens and incomplete in others, Figure 24(b). It was also observed in a separate study involving submerged arc welding that heat inputs of more than 1.7 kJ/mm could not be used as melt-through occurred in Lehigh specimens.

Effect of Increasing Weld Hydrogen Content.

The critical T_{ph} increased as the weld hydrogen content increased. The critical preheat temperatures are shown in Table 25. The critical T_{ph} increases almost linearly with weld hydrogen content, H_{DM} . The preheat temperatures from AWS D1.1, Cottrell's and Nippon steel prediction schemes are plotted against the observed critical preheat temperatures in Figure 25.

Acoustic Emission Results

Effect of Preheating Temperature on Crack Initiation Time.

For a given weld hydrogen content, Lehigh weldability crack initiation results are shown in Table 25. It was observed that crack initiation time increased dramatically as the preheating temperature increased. With weld hydrogen content of 16.18 ml/100 gms of hydrogen and a preheat temperature of 150°C the cracking initiated after 3 days.

Crack initiation times for different weld hydrogen contents with no preheat shows that, Figure 26, Table 26, crack initiation time decreases rapidly as the weld hydrogen content increases. Time for complete cracking decreases as the weld hydrogen increases, Table 26, Figure 26.

Crack Initiation Temperature

From the simultaneous study of AEM and thermal cycle measurements, the temperature of the near weld zone at the time of crack initiation was determined, Table 27. It is observed that no cracking occurred before the weld had cooled down to 80°C. The crack initiation temperature in non-preheated specimens was always lower than 80°C but higher than 60°C. In preheated specimens, cracking occurred when the weld had cooled down to temperatures lower than 55°C.

Crack Propagation Rate

Complete cracking takes place in shorter times as the weld hydrogen content increases, Table 25. It is clear that crack propagation rate increases as H_{DM} increases. A rough estimate of crack propagation rate can also be made from these results. If we compare the crack completion times for specimens at room temperature for different weld hydrogen contents, then crack propagation rate can be calculated, Table 26. From, Figure 27, it is observed that the longitudinal crack propagation rate increases with increasing weld hydrogen.

Crack Initiation Region

From AEM, it was observed that crack initiation location always fell in the second half of the weld, Table 28. Whereas, physical crack propagation occurred from weld start to weld end.

Acoustic Emission Characteristics :

Different hydrogen contents and preheating temperatures did not cause a significant effect on the AE signature. Different features, such as, ring-down counts, energy and peak amplitude did not exhibit any change as a function of of weld hydrogen content and preheating temperature, Table 28. The slope of AE events increased substantially for a H_{DM} of 16.18. The number of events decreased with increasing T_{ph} . For specimens cracked at RT, the number of events decreased with increasing H_{DM} . This agrees well with the fact that cracking completion time also decreased with increasing H_{DM} .

Cracking Ratio Determination

After cross-sectioning the specimens, longitudinal and sectional cracking ratio's were determined. Most of the cracked specimens showed longitudinal cracking ratio of 1, Table 25, suggesting that specimens cracked completely in the longitudinal direction. Also, all the specimens, except the specimen welded at 150° C with a H_{DM} 16.18, showed a sectional ratio of 1. This one exception had a sectional ratio of 0.8. It should be noted that crack initiation occurred 5 days after welding in this specimen.

TEKKEN TEST RESULTS

Table 29 lists the cracking ratio and observed T_{ph} for the two H_{DM} contents used with Tekken tests. Weld hydrogen contents of 2.7 ml/100 gm DM did not cause HAC even with no preheating. For weld H_{DM} of 16.18, a preheating temperature of 195°C was required.

Root Penetration

Root penetration was lowest at the weld start and steadily increased along the weld length and was maximum at the weld end. The effect of varying root penetration is not significant in Tekken specimens because the stress concentration due to the root face induced mechanical notch is still operative.

Effect of Increasing Weld Hydrogen

As mentioned in the experimental procedure, Tekken tests could only be conducted with two weld hydrogen contents. Therefore, the data on Tekken tests is quite limited. In spite of this limitation, it is clear from Figure 28 that T_{ph} required to avoid HAC increases with an increase in weld hydrogen content. The predicted preheating temperatures from AWS-D 1.1, TWI/BS and Cottrell's prediction schemes are also compared with the observed T_{ph} in Figure 28. Clearly, the predicted preheat temperatures from these schemes are inadequate to avoid HAC in Tekken specimens.

Acoustic Emission Characteristics

Source location revealed that crack initiation occurred between the mid-length and 3/4 distance from the weld start. It should be noted that the accuracy of source location in Tekken specimens was much poorer than that for Lehigh specimens. Therefore, crack initiation locations can not be placed accurately. Crack initiation and completion time increased markedly with an increase in T_{ph} , Table 29. Weld metal thermal cycles were recorded only for Tekken test specimens welded at room temperature. The thermal cycle data in combination with the AE data shows that crack initiation temperature was about 80°C for Tekken specimen welded with a H_{DM} of 16.18 and at RT. The peak amplitude and energy were unchanged with an increase in T_{ph} , Table 30. Slope of AE events for H_{DM} of 16.18 were similar to those observed in the Lehigh specimen at the same H_{DM} level, Table 30. More number of events were observed in the second half of the weld metal as compared to the first half in all cracked specimens, Table 30.

HARDNESS

Knoop's hardness number (KHN) variation across the weld zone is shown in Figure 29 for Tekken test specimens. As expected, the coarse grained HAZ (CGHAZ) and maximum HAZ hardness decreased with increasing preheating temperature, Figure 30(a), (b). The weld metal did not show any significant effect of increasing T_{ph} , Figure 31. Similar trend was observed with the Lehigh specimens.

METALLOGRAPHY

Lehigh Specimens

Microscopic observation revealed that crack initiation and propagation occurred exclusively in the weld metal in all the Lehigh specimens. A typical crack propagation mode in Lehigh specimens is shown in Figure 32(a), (b). After initiating, the nucleated crack propagated almost parallel to the thickness direction.

Tekken Specimens

In Tekken specimens, crack initiation initially occurred along the coarse grained HAZ for some distance before entering the weld metal in specimens welded at RT and 130°C. In the specimen welded at 175°C, crack initiation occurred along the fusion line. The crack then propagated along the CGHAZ for a short distance before entering the weld metal. The weld metal profile in Tekken specimens in this study was such that it created a sharp notch root almost parallel to the thickness direction, Figure 33, 34 and 35.

FRACTOGRAPHY

Lehigh Specimens

Two distinct fracture features were observed in Lehigh specimens. At some of the notch root locations, the predominant fracture mode comprised of micro-void coalescence (MVC), similar to that shown in Figure 36(a), while at remaining locations, it was quasi-cleavage (QC) type. The MVC, i.e., ductile type fracture, occurred over a small area ranging from 30 to 120 microns along the weld length, Figure 36(b). EDS analysis of the big inclusion in Figure 37(a), revealed it to be an oxide inclusion of Mn, Al, Ti and Ca. The crack propagation occurred mostly in QC fashion, Figure 37(b). Intergranular type fracture mode was rarely observed, Figure 38(a), (b). The overall fracture exhibited columnar crack propagation texture in the weld metal.

Tekken Specimens

In Tekken specimens, fracture mode at notch root location was mostly micro-void coalescence type, Figure 36(a) and (b). Crack propagation occurred entirely under quasi-cleavage mode. No intergranular crack propagation was observed in this case. The crack initiation region revealed equi-axed dimples of varying size. Fish-eye type features were also observed. The plastic zone size at different locations also varied from 30 to 120 micros.

WIDE PLATE STUDY

The thermal cycles for the standard and wider Lehigh specimens show that weld cooling rate in both the specimens is almost identical until the weld metal reaches 100° C, with the standard Lehigh specimen cooling at a much slower rate below 100° C. The weld cooling times to 100° C, 45° C and to RT are shown in Table 31. The exact effect of these weld cooling rates on H_r content is analyzed in detail in section 3 using the FEM.

CRACK PROPAGATION STUDY

The numerically calculated stress intensity factor calculations are shown in Table 32 and Figure 39. In Figure 39, $\frac{K_{II}}{K_I}$ ratio is plotted against crack length for cracks at different angles. In addition, experimentally observed crack angles and crack lengths in Tekken specimens welded under different conditions are also shown in Figure 39. It is observed that $\frac{K_{II}}{K_I}$ ratio increases as the crack length increases. This increase becomes more pronounced with increasing crack angle. It is observed from Figure 39 that a crack originating at 60° angle will experience mixed mode loading and that the $\frac{K_I}{K_{II}}$ ratio for even a 0.0005 meter long crack will be high enough to cause kinking. It was also observed that mode II type loading, K_{II} component, is non-existent in Lehigh specimens since the crack is perpendicular to the loading direction.

2.4 - DISCUSSION

ROOT PENETRATION

Lehigh Test Specimen

The Lehigh weldability tests in this study resulted in variable penetration. Figure 24(a). The welding start location showed least penetration and it increased as the weld progressed. The variation in root penetration was caused by two factors. The test specimen is cold, i.e. it has not "seen" any welding heat, at the start of welding. Secondly, the welding parameters are not stablized in the beginning. A combination of these two factors caused lowest penetration at the weld start in all the specimens, Figure 24(b).

The Lehigh test weld simulates the root pass at mid-thickness in a double V-groove joint in the presence of a mechanical notch³⁶. The stress concentration effect at the weld root due to this notch is instrumental in initiating HAC. For successful application of the Lehigh test, it is necessary to obtain a test weld bead such that the mechanical notch at the weld root is maintained along the weld length. This is achieved by maintaining the root penetration at a constant level. The variation in root penetration, then, suggests that the stress concentration effect is non-uniform along the weld length. In some specimens, Figure 24(a) and Figure 40(b), complete penetration was observed as the weld progressed. The mechanical notch induced stress concentration effect was either completely absent or the stress concentration factor was much lower at these locations in these specimens. The locations with complete penetration will, hence, exhibit a lower tendency to HAC compared to the locations with in-complete root penetration.

The deviation of the weld bead from the mid-thickness or the neutral axis, also known as eccentricity, causes a simultaneous variation in the stress state at the weld root. The variation in restraint intensity, R_F , with weld eccentricity has been studied in detail by Matsui⁷⁸ and Suzuki⁵⁹. Since the stresses acting on the weld are dependent on R_F , it is likely that variation in root penetration also affects the stresses acting on the weld.

It was also observed that the weld metal area, in general, increased as the weld progressed. It was lowest at the weld start and gradually increased towards the weld end, Figures 24(a),(b) and 31(a),(b). As a result, the restraint stresses acting on the weld will be non uniform along the weld length. Similar variations in restraint stress with heat input were also observed by Kirihara et al⁷⁹. and Bretz and Hoffmeister⁸⁰. This is contrary to the results of Satoh et al⁸¹. which suggested that reaction stress is independent of heat input. This is due to the fact that as the heat input increases the thermal contraction also increases, but the restraint stresses remain constant, since the restrain force is now distributed over a larger weld metal mass, and, hence, weld metal area. The resultant reaction stress or the restraint stress, σ_R , is expressed by the following equation by Satoh and Matsui⁸².

$$\sigma_{\rm R} = S R_{\rm F} / h_{\rm w} \qquad (MN/m^2) \qquad (8)$$

Where h_w is weld throat height (mm) and S is the total free contraction.

Variation in restraint stresses due to variation in weld metal area was also reported by Wingrove et al⁸³. for Tekken specimens. These authors reported an increase in restraint stress with an increase in the welding speed. Wingrove et al⁸³. proposed that the weld metal cross sectional area decreased disproportionately to nullify the effect of decreased thermal contraction due to the reduction in heat input. Similar arguments hold for Lehigh weldability test specimen due to its slot test weld nature. In view of this, the root penetration variation observed in Lehigh specimens becomes important as the heat input is constant for the test weld used in this study. According to Masubuchi and Ich⁵³, the R_F is higher at the weld ends for Lehigh test specimens and decreases parabolically as the weld center is approached. Therefore, the restraint stress acting is not uniform along the weld length even if the term S/ h_w is assumed to be independent of heat input. This effect, in combination with variation in the weld metal cross section, suggests that the final restraint stresses will be non uniform across the weld length. The stress concentration variation with root penetration is more pronounced compared to the restraint stress variation. Because, at the location of incomplete penetration, K_t equals 4 compared to 1.5 at the complete penetration location, Figure 13.

Tekken Test Specimen

The Tekken test specimen also experienced root penetration variation along the weld length. However, its effect on HAC is minimized because stress concentration at the notch root is always present in Tekken specimen. Therefore, incomplete root penetration is not an influencing factor in HAC initiation phenomenon in Tekken tests. The root penetration variation will, however, affect the transverse residual stress state at the weld root. This effect will be dependent on the magnitude of penetration variation. As discussed in the previous section, the restraint stress will be a direct function of R_F , if we consider that S/ h_w is constant along the weld length. Nevertheless, similar arguments to those made for Lehigh specimens also hold in case of Tekken specimens.

ASSESSMENT OF PREHEAT PREDICTION METHODS

Lehigh Test Results

The critical T_{ph} necessary to avoid HAC increased with an increase in weld hydrogen content expressed in H_{DM}, Table 25. Some explanation is necessary when the predicted T_{ph} is lower than 20°C (RT). In such cases, the weldability test was conducted at RT only, instead of cooling the weldability test specimen to the predicted T_{ph} . Since HAC was avoided at RT, no additional tests were performed at temperatures lower than RT. It is generally a recommended practice to bring the plate temperature up to RT if the plate is at a much lower temperature. Therefore, in the present study, whenever the predicted T_{ph} was lower than RT, then it was made equal to RT in all the Figures.

The predicted preheating temperatures from AWS D-1.1, Yurioka's (Weld Metal - Nippon Steel) and Cottrell's schemes are compared with the observed T_{ph} in Figure 25. It should be noted that AWS D1.1-88 and Cottrell's prediction schemes estimate T_{ph} for avoiding HAC in

HAZ and not in weld metal. The purpose of presenting these preheating temperatures was to demonstrate certain limitations of these schemes. It is generally assumed that T_{ph} which precludes HAC occurrence in HAZ will also avoid HAC in weld metal⁴⁹. This was not observed with some of the prediction schemes. Predicted T_{ph} 's from Cottrell's prediction scheme were not sufficient to avoid weld metal HAC in Lehigh test specimens at H_{DM} content of 2.87 and 8.67, Figure 25 and 28. Whereas, T_{ph} 's from AWS D1.1-88 were not sufficient to avoid weld metal HAC in Lehigh 50.11 and 50.

It should be noted that Cottrell's T_{ph} prediction scheme is dependent on H_{DM} levels of 5, 10, 15 and 25. Therefore, for the H_{DM} levels used in this study, the appropriate next higher H_{DM} level was used. For example, for a H_{DM} value of 16.18 used in this study, H_{DM} was assumed to be equal to 25 when using Cottrell's prediction schemes. This results in higher T_{ph} from Cottrell's scheme. In the author's opinion, such a broad classification of H_{DM} levels will lead to unsafe T_{ph} predictions. For example, when welding two specimens under identical welding conditions, but with H_{DM} of 16 for one specimen and 25 in the other specimen, Cottrell's T_{ph} will be identical. In reality, HAC susceptibility will vary widely for these two H_{DM} levels. T_{ph} from Yurioka's method to avoid HAC in HAZ was sufficient enough to avoid weld metal HAC in Lehigh specimens at all the H_{DM} levels, Figure 25 and 28.

The observed critical T_{ph} was compared with the predicted T_{ph} from Nippon steel method^{34, 63-65} for avoiding HAC in weld metal, Figure 25. As mentioned earlier, this prediction method is for multi-pass welds and was modified in this study by the author to apply it to single pass Lehigh test specimen as follows. The weld throat for the welding conditions in this study was smaller than the applicable range of this scheme. The lowest weld throat to which Nippon steel method is applicable is 10 mm and it was used in the present study even though the weld throat in experimental Lehigh test specimens was smaller than 10 mm.

From Figure 25, it is observed that predicted preheating temperatures are on safer side for H_{DM} levels of 2.87, 8.67 and 11.79. For these three H_{DM} levels, the predicted preheating

temperatures are conservative by about 12 to 20° C. This is an excellent agreement between observed and predicted T_{ph}. A better agreement can be obtained if the prediction scheme had provided T_{ph} calculations for weld throat depths of less than 10 mm.

In contrast to this, at a H_{DM} level of 16.18 the predicted T_{ph} from the Nippon steel method is slightly lower than the observed critical T_{ph} . The reason for prediction of unsafe T_{ph} at high H_{DM} levels is unknown. It can be attributed to the the fact that effect of H_{DM} on critical T_{ph} is supposed to be logarithmic. Therefore as the H_{DM} content increases, the corresponding increase in critical T_{ph} becomes gradual. As a result, even if the H_{DM} changes from 11.79 to 16.18 the T_{ph} increases by just 5°C. This 5°C increase in T_{ph} will not alter weld cooling characteristics and residual stress distributions significantly enough to offset the H_{DM} increase from 11.79 to 16.18.

It should be noted that Nippon steel weld metal preheating prediction scheme only considers weld hydrogen, tensile strength of the weld metal and thickness. It does not consider the influence of weld microstructure. This has been correctly pointed out by Kasuya⁷⁶. According to him, non consideration of weld metal microstructure will render this prediction scheme less applicable. In addition, this scheme was developed exclusively for multi-pass welding situation, and hence can not be applied to single-pass welding situations, such as Lehigh weldability test. In view of this, preheat predictions from Nippon steel method should be used with caution for single pass applications. Currently, the researchers at Nippon steel are working towards developing a comprehensive preheat prediction scheme for avoiding weld metal HAC⁷⁶. In the mean time, existing Nippon steel method can be used to successfully avoid HAC in weld metal.

Tekken Test Results

The weld hydrogen content effect on T_{ph} was as expected. A T_{ph} of 195°C was required to avoid cracking in the Tekken test specimen with a H_{DM} of 16.18. Unfortunately, because of the experimental difficulties, as mentioned the experimental procedure section, additional data with intermediate weld hydrogen levels is not available. In spite of the limited data, certain features are noteworthy.

From Figure 28 it is observed that for H_{DM} of 2.87 no cracking was observed in the Tekken specimen welded at RT. It should be noted that T_{ph} was not observed at RT in Tekken test specimen with H_{DM} of 2.87. Actual T_{ph} could be lower than RT. Cottrell's scheme predicted T_{ph} of 0°C. But, Tekken tests were performed at RT instead of at 0°C. This is in accordance with the recommended practice that welding be carried out at RT.

Yurioka's preheating prediction method predicted a T_{ph} of about 140°C for H_{DM} of 2.87. In comparison, Cottrell's scheme predicted T_{ph} of RT and the AWS D1.1 Table recommended a T_{ph} of 52°C. The AWS D1.1 hydrogen control method predicted a T_{ph} of 110°C. The T_{ph} from Cottrell's scheme were found to be in good agreement with the observed results. The T_{ph} from AWS-D1.1 for all the weld hydrogen levels is 52°C. This preheating temperature, though, is recommended for welding procedures incorporating low weld hydrogen content. When low hydrogen conditions can not be maintained, AWS D1.1-88 recommends the use of hydrogen control method. Preheating temperatures according to the hydrogen control method were calculated and are shown in Table 33. It is clear that at a low H_{DM} of 2.87, the predicted T_{ph} from AWS hydrogen control method is unnecessarily high compared to the observed critical T_{ph} . These results suggest that Yurioka's method predicts unnecessarily high T_{ph} at this weld hydrogen level.

In contrast to this, the results at a H_{DM} of 16.18 are significantly different. The observed T_{ph} of 195°C is 5°C lower compared to that predicted from Yurioka's method. The predicted T_{ph} of 175°C from Cottrell's scheme did not avoid HAC at this weld hydrogen level. Whereas, the AWS-D1.1 hydrogen control method predicted T_{ph} of 160°C was significantly lower than the observed critical T_{ph} .

The above results suggest that use of AWS-D1.1 preheating tables will not avoid HAC in

Tekken test welds. The AWS-D1.1 preheating tables are based upon years of fabrication experience and are used in structural fabrication. Structural fabrication generally encounters lower restraint levels than the Tekken test. In addition, most weld joints are fillet type in structural fabrication. Therefore, it is expected that T_{ph} from AWS-D1.1 preheating tables will not avoid HAC in a high restraint situation, such as a Tekken specimen. The restraint levels encountered in bridge construction welding are similar to those in Lehigh test specimens. But, the T_{ph} recommended in AWS-D1.1 preheating tables do not avoid HAC in Lehigh specimens at H_{DM} levels higher than 8, as discussed in the previous section. AWS-D1.1 preheat Table recommends T_{ph} which are independent of weld hydrogen levels or restraint levels. This is in contradiction to the observed fact that critical T_{ph} increases as weld hydrogen levels increase, since the material becomes more embrittled with increasing hydrogen content. In comparison, AWS D1.1 hydrogen control method predicts T_{ph} which are unsafe only at a H_{DM} of 16.18 in Lehigh specimens.

Cottrell's prediction scheme estimates critical T_{ph} reasonably well at lower H_{DM} of 2.87. Whereas, at higher H_{DM} of 16.18, the predicted preheat temperatures are insufficient to avoid HAC in Tekken test specimens. In fact, application of TWI/BS or Cottrell's method to Tekken test T_{ph} is not recommended since these methods are based on CTS test results. The CTS test simulates a fillet weld and has lower restraint than the Tekken test specimen. It is therefore expected that predicted T_{ph} from these methods will not avoid HAC in Tekken test specimens. In spite of these limitations, the results in Figure 28 clearly suggest that the predicted preheating temperatures from Cottrell's method are in good agreement with the observed critical T_{ph} . Cottrell modified the existing TWI/BS method by including the effect of inclusions and cooling rate below 300°C. But, the estimated T_{ph} from Cottrell's scheme are only 5 to 10°C higher than those from TWI/BS method.

From the Tekken test results and Yurioka's predicted T_{ph} , it is clear that Yurioka's method predicts unnecessarily high preheating temperatures for lower weld hydrogen contents. This was also observed with Tekken test specimens welded using FCAW for a base metal with CEN of 0.43 and for 12.7, 25.4 and 38 mm thick specimens. The welding details are in Table 34. The results also suggest that at low hydrogen levels Yurioka's nomograph method predicted too conservative temperatures for 38 mm thick specimen, Figure 41. Cottrell's predicted T_{ph} were in good agreement with the observed T_{ph} .

At high weld hydrogen contents with $H_{DM} = 16.18$, Yurioka's predicted T_{ph} agrees well with the observed T_{ph} . Since, weldability test data are unavailable for intermediate weld hydrogen contents, it is difficult to comment about T_{ph} for these weld hydrogen levels. In addition, the weld hydrogen levels at which Yurioka's method shifts from being too conservative to being applicable is also unknown. These results clearly suggest that even Yurioka's method, in spite of being the most comprehensive, has some limitations.

After several discussions in this regard, Kasuya⁸⁴ indicated that the original prediction method ⁶ was based on weldability test results with H_{DM} levels higher than 5. Very few weldability test results were based on H_{DM} levels of 3 or lower. Therefore, T_{ph} prediction at lower H_{DM} levels is not based on a large database of weldability test results. Another reason for too conservative T_{ph} at lower H_{DM} could be the use of weld hydrogen contents as determined by the JIS glycerine method. It is well established that hydrogen measurements using glycerine are generally lower due to hydrogen solubility in glycerine, especially at lower H_{DM} contents. This results in reduced hydrogen evolution, and hence measurement. Even though the JIS H_{DM} values were expressed in terms of IIW values, fluctuations are inherent. This will lead to an incorrect estimate of IIW H_{DM} values. As a result, the effect of H_{DM} at lower levels was probably in error.

In general, high strength steels, because of inherently higher HAC susceptibility, are welded using weld hydrogen contents of lower than 5 ml/100 gm of DM. Therefore, use of Yurioka's nomograph method is not recommended in these situations. The above mentioned drawback limits the applicability of this method to a wide section of industrial fabrication. A larger Tekken weldability test database at lower weld hydrogen contents is needed to improve the accuracy of predictions from Yurioka's method. Yurioka's nomograph method predicts T_{ph} for steel chemistries with CEN values lower than 0.6. Most of the high-strength steels have CEN values much higher than 0.6. Yurioka's method, as well as other prediction methods, do not cater to these steels. As steel chemistries richer in alloying elements are being developed, this area of research is becoming increasingly important. More work with these steels is required before a comprehensive T_{ph} prediction method can be developed.

ACOUSTIC EMISSION RESULTS

Lehigh Weldability Tests

Crack initiation in most of the specimens occurred in the second half of the weld, Table 28. Only in few instances multiple crack initiation sites were observed, Table 28. Crack initiation takes place at the weld notch root due to the stress concentration associated with the sharp notch and can occur at different locations along the weld length. The location of first AE events in the mid-length region indicates that first crack nucleation events occurred near the mid-length at the weld root notch. Following this, crack initiation at weld notch root location was observed at other locations along the weld length. These nucleated cracks at the weld root notch then were joined together to cause complete cracking. The crack propagation during this stage occurred from the weld start region to the weld end as observed physically. It should be noted that crack initiation and crack propagation are separate events.

Crack propagation physically occurred from weld start to weld end in all the specimens. A possible reason for crack propagation to occur from weld start to weld end is discussed below. In Lehigh specimens, by the time the test welding is complete, the weld start location has already cooled down to low temperatures. This consolidates the residual stresses in this region, whereas, the latter part of the weld is at higher temperatures (lower yield strength), and hence does not experience high residual stress build-up. This consolidation of transverse residual stresses in the start region will then propagate the initiated crack in the weld start region. Additional influence

of residual stress variation along the weld length on crack propagation is discussed in Chapter 4.

The fractographic examination, discussed in a later section, revealed that micro-void coalescence type of fracture at weld root notch was observed intermittently throughout the weld length. Micro-void coalescence type fracture indicates plastic straining suggesting that at this location the von-Mises stresses (equivalent stresses) exceed the yield strength of the local microstructure. The observation of MVC at weld root notch along the weld length suggests that crack initiation occurred by MVC at several places. These nucleated cracks then propagated subsequently under the influence of developing transverse residual stresses causing complete failure.

Tekken Specimens

In contrast to the Lehigh specimens, multiple crack initiations were observed in Tekken specimens. Most of these sites were located in the seconds half of the weld, Table 30. It should be also noted that the source locations were not as accurate as that in the Lehigh specimens because of the AE configuration employed. The higher restraint intensity in Tekken specimens promotes higher transient as well as residual stress levels along the weld lengths. This promotes crack initiation at multiple sites.

HARDNESS RESULTS

Figure 29 shows the hardness results for a Tekken specimen welded at RT using a shielding gas containing 1% H₂. Similar hardness variation was also observed with Lehigh specimens. It is observed that coarse grained HAZ (CGHAZ) does not exhibit the maximum hardness. The hardness slowly increased with a transition from the weld metal into the CGHAZ. Maximum hardness was observed in the HAZ region that was 1.7 mm away from the fusion line. This region corresponds to inter critically reheated HAZ, i.e. this region experienced peak temperatures which were between the AC₃ and AC₁ critical temperatures. This observation suggests that the use of maximum HAZ hardness as an indication of HAC susceptibility can be misleading, especially in Tekken specimens.

The notch at the weld root on the double bevel side of the Tekken specimen introduces stress concentration. The magnitude of the localized notch stresses under the action of residual stresses is high enough to cause yielding. The size of this plastic zone is of the order of 100 microns, Figure 36(b). Clearly, the maximum hardness region, which is 1.7 mm away from the fusion line, does not influence HAC initiation or propagation at all in Tekken specimens since crack initiation occurred along the fusion line in Tekken specimens as will be discussed in later sections. Similar explanation also holds for Lehigh test specimens. The above mentioned argument suggests that the maximum HAZ hardness is not indicative of actual HAC susceptibility, at least in A514 steel. Qualitatively, though, maximum hardness is indicative of HAC susceptibility, since, as the maximum HAZ hardness increases the CGHAZ hardness also increases, Figure 30(a), 30(b).

The hardness further decreased with a transition from the inter critically reheated HAZ into the overtempered HAZ region in the RT specimen, Figure 29. Similar trend was also observed with increasing T_{ph}, although, the location of the maximum hardness HAZ region from the fusion line increased.

The variation of CGHAZ hardness and maximum hardness with increasing T_{ph} is shown in Figure 30(a), (b), respectively, for Tekken specimens. The hardness variation with increasing T_{ph} is essentially similar to that predicted by Yurioka's hardness prediction equation³⁴. Referring to the CCT diagram for a similar composition A514 steel, Figure 42⁸⁵, it is clear that $t_{8/5}$ cooling times lower than 8 seconds promote martensite formation. At higher $t_{8/5}$ cooling times upper bainite formation, with lower hardness than martensite, is preferred. Therefore, an increase in T_{ph} , which increases $t_{8/5}$, results in decreased hardness as observed in Figure 30(a) and (b).

It should be noted that the hardness values shown in Figure 30(a) and (b) varied with different locations along the weld throat. This also suggests that the HAZ typically experiences different thermal cycles along the weld throat, as confirmed by the varying width of the HAZ across the weld throat. Similar trend was also observed with Lehigh test specimens, though, the individual hardness values differed slightly compared to the Tekken test specimens. This is probably caused by the groove shape induced cooling rate differences.

The weld metal hardness did not vary significantly with increasing T_{ph} , Figure 31. This suggests that the weld metal transformation is less sensitive to weld cooling rates. This is because the lower carbon content of the filler metal lowers the formation of high hardness martensite. As a result, the weld metal hardness variation with increasing T_{ph} is less pronounced. Both Lehigh and Tekken test specimen exhibited the same trend.

METALLOGRAPHY RESULTS

Lehigh Specimens

It was observed that crack initiation occurred in the weld metal of Lehigh specimens. In all the cracked Lehigh specimens in this study, crack initiation as well as propagation occurred entirely in the weld metal. The crack initiated from one of the weld shape induced notch roots, Figure 32. This suggests that the mechanical notch at the weld root is instrumental in initiating the crack. As discussed earlier, root penetration is significant from HAC initiation viewpoint in Lehigh specimens. The crack propagation then occurred in a direction parallel to the thickness direction except for the Lehigh specimen welded at RT with $H_{DM} = 16.18$, the nucleated crack changed its direction into the weld metal at an angle of about 45°.

Konkol et al³⁶ observed that HAC occurred in the HAZ in a few Lehigh test specimens. This is contrary to the common experience of HAC in weld metal in Lehigh specimens. These authors did not explain the reasons behind this observation. The results in the present study, however, clearly suggest that Lehigh weldability test specimen only assesses HAC susceptibility of the weld metal and not the HAZ. In addition, the effect of root notch induced stress concentration effect is instrumental in crack initiation. Therefore, complete root penetration variation will substantially alter HAC susceptibility in Lehigh tests.

Tekken Specimens

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In all the cracked specimens, HAC occurred from the weld notch root on the double bevel side, Figures 33,34 and 35. Cracking in Tekken specimens showed crack initiation in either HAZ or along the fusion line. Tekken specimens with H_{DM} of 16.18 and welded at RT and 130°C cracked in the HAZ, Figure 33 and 34. In comparison, in the Tekken specimen welded at 175°C HAC initiation occurred along the fusion line, i.e, between weld metal and HAZ, Figure 35. This suggests that the CGHAZ was less susceptible to HAC in the specimen welded at 175°C. This is also supported by the hardness values for CGHAZ as a function of T_{ph}, Figure 30(a). It is observed from Figure 30(a) that the CGHAZ hardness did not decrease substantially from the RT value until a T_{ph} of 175°C was reached. At this T_{ph}, the CGAHZ average is 375 KHN. The weld metal hardness is around 355 KHN for all the specimens irrespective of the T_{ph} used. As a result, both the CGAHZ and weld metal exhibit almost equal susceptibility to HAC in this specimen from hardness perspective alone, and, hence, HAC initiation occurred along the fusion line under the influence of stress concentration at the notch root. In Tekken specimens welded at RT and 130°C, higher average CGHAZ values of 420 and 410 KHN, respectively, indicating that CGHAZ in these specimens is highly susceptible to HAC compared to the weld metal. HAC initiation, hence occurred along the HAZ in these specimens.

The above discussion suggests that HAC initiation location in Tekken specimens is primarily a function of microstructure as well as the notch root shape and location. Suzuki⁴⁴ noted that the restraining stress on the weld metal and especially the stress concentration factor at the notch shaped root are the factors directly responsible for root cracking. It was also observed by Bala et al⁸⁶. that only weld metal cracking was observed in A710 steel even when undermatched filler metal was used⁸⁶. Weld metal cracking is preferred over HAZ cracking in this case because the weld root notch on the double bevel side has a very high root radius. As a result, the stress concentration effect at this location is absent. The cracking then initiated from the single bevel weld root notch which generally propagates into the weld metal. In contrast to this, Sawhill, JR. et al⁸⁷. stated that in y-groove Tekken tests, the joint geometry does not favor either the HAZ or the weld metal, even though, the exact nature of weld root notch shape from crack initiation viewpoint was not discussed by these authors. It is clear form the above discussion that HAC initiation in Tekken test depends primarily on the local stress state at the weld root notch on the double bevel side. If this notch has a large root radius, i.e., no stress concentration effect, then crack propagation from other single bevel side notch or from the weld metal mid section is preferred. It is thus necessary to exercise a good control over the weld shape in Tekken specimens in order to correctly assess the HAZ susceptibility to HAC. This can be achieved by performing a number of trials on Tekken specimens with the desired welding conditions and optimizing the welding set-up to generate a sharper notch at the weld root locations on the double bevel side.

FRACTOGRAPHY

Lehigh Specimens

The fracture surface showed two distinct features at the weld notch root. Micro-void coalescence (MVC) as well as quasi-cleavage (QC) type fracture mode was observed intermittently at the weld root. The width of the MVC region varied from 30 to 125 microns along the weld length. At some notch root locations, dimples with embedded inclusions were observed. In addition, typical fish-eye features were also observed, Figure 37(a). The electron dispersive spectroscopy analysis indicated that these inclusions were oxide inclusions of Al, Mn, Ca and Ti. The large size of the inclusions and the oxide type suggest that they belonged to the weld metal. Oxide inclusions are not observed in the base metal or HAZ. The above observations suggest that crack initiation at these notch root locations occurred by plastic straining.

In a self-restraint weldability test, such as Lehigh, von-Mises stresses above the yield point will only occur at the weld root notch locations when the specimen is in fully restrained condition. As soon as crack initiation occurs, the accompanying load relaxation will cause the crack tip stress intensity factor to drop. The magnitude of this drop will depend on the competition between the stress intensity factor increase due to crack length increase, and the load drop caused by the stress relaxation. The above discussion suggests that the MVC region at the weld root experienced highest stress concentration and is the crack initiation region. The nucleated crack then propagated under QC mode because of the decreased stress intensity factor. Similar fracture characteristics were observed by Vasudevan et al⁸⁸. when assessing HAC susceptibility using Lehigh slot and Implant weldability tests. It should be noted that the Lehigh slot test was developed for assessing field weldability of pipeline steels and is different than the Lehigh Ugroove test used in the present study. These authors observed MVC fracture mode at the crack initiation site in Lehigh slot weldability test. The fracture mode later changed to QC and intergranular (IG) type supporting the view that the crack tip stress intensity decreases with continued crack propagation in a self restrained Lehigh slot test. From the y-groove Transverse restraint cracking test results, Matsuda, et al⁸⁹. proposed that QC plus MVC regions formed at higher stress intensities.

The fracture surface also exhibited a columnar texture, Figure 38(a). This is caused by the columnar grains of the weld metal. This was also observed by ⁹⁰ with Tekken specimens. Only at a few locations in the weld metal, IG fracture mode was observed, Figure 38(b). It is well established that ⁹¹⁻⁹³ HAC occurs intergranularly at the lowest stress intensity factor but is not favored over QC and MVC fracture modes because it is kinetically less favorable. It was observed that basic fracture mode was almost constant for different weld hydrogen contents as well as T_{ph} 's. As a result, no significant change in the AE signature was observed as a function of weld hydrogen content and T_{ph} 's.

Tekken Specimens

Cracks always initiated at the weld notch root along the double bevel side, as commonly observed in Tekken tests. The cracked weld notch root predominantly exhibited MVC fracture mode. A QC type fracture mode was observed at few other weld root notch locations. The MVC

region was found to vary from 25 to 120 microns. As compared to the Lehigh test specimen, the incidence of MVC type fracture was more frequent in the Tekken test specimen. This suggests that higher von-Miscs stresses are observed in the Tekken test specimen as compared to the Lehigh specimen. This directly follows from the fact that the higher R_F in the Tekken specimen causes the over all residual stress magnitude to be higher compared to the lower intensity Lehigh specimens. As a result, there is a higher incidence of micro-void coalescence at the weld notch root location in the Tekken test specimen compared to the Lehigh specimen.

Similar arguments to that presented for Lehigh specimens also explain the MVC fracture mode at the weld notch root. The crack initiated by MVC and then propagated under QC mode at lower stress intensity factors. Dimples embedded with inclusions were also observed at a few locations. The inclusion size and the round shape suggests that these inclusions belong to the weld metal. This suggests that crack initiation probably occurred in the weld metal. After a careful optical microscopic examination on the cross sections, it was observed that the crack initiation took place along the fusion line for about 10-15 microns before entering the HAZ in specimen welded with H_{DM} of 16.18 at a T_{ph} of 175°C.

The crack propagation occurred by QC fracture mode. No IG fracture was observed. In contrast to this observation, Matsuda⁹⁰ observed all types of fracture modes in Tekken specimens. The IG type fracture occurs at low stress intensities. Therefore in Tekken specimens, the means stress acting is high enough so that that QC fracture mode is preferred over the IG type cracking. In contrast, the Lehigh specimen shows a few occurrences of IG fracture due to lower means stresses, and hence lower stress intensities. Similar observations were made by Vasudevan et al⁸⁸. when assessing HAC susceptibility using Lehigh slot test and implant test. They observed MVC mode at the crack initiation sites. The fracture mode later changed to QC and IG as the stress intensity dropped due to stress relaxation in the self restraint test. Whereas, in implant specimens, the crack propagation mode changed from IG to QC to MVC with an increase in the crack tip stress intensity concomitant with an increase in the applied load.

This fundamental difference in self restraint weldability tests and externally restrained weldability tests becomes significant with respect to practical joints. The practical joints will experience either self restraint or externally restrained loading conditions. Therefore, the HAC initiation as well as the propagation behavior in these cases will differ depending on the loading situation. In order to assess HAC susceptibility of practical joints, the weldability tests should be selected which simulates the actual loading condition of practical joints. According to Yurioka and Suzuki⁴⁸, a self-restraint test simulates HAC in a weld (practical joints) more accurately than does an external restraint test. However, any quantifiable data on critical conditions for HAC can not be obtained. The selection of weldability test should, then, be based on the required information and practical relevance.

APPLICABILITY OF WELDABILITY TEST RESULTS TO PRACTICAL JOINTS

The basic purpose of weldability tests is to assess HAC susceptibility of a weld joint being used in practice. Small size weldability tests were developed to simulate HAC response of larger practical weld joints to, 1) conserve material, and 2) to avoid the difficulties associated with testing large joints for different test variables. Since, HAC depends on weld hydrogen, tensile residual stresses and susceptible microstructure, it is necessary that these factors remain constant in order to correctly assess HAC susceptibility of practical joints. Most of the commonly used weldability tests are root or single pass cracking tests. According to Konkol et al³⁶ experience suggests that weld cracking is most likely to occur in the root pass even though the cooling rate may be lower due to the smaller heat sink provided by the surrounding base metal. Root cracking is enhanced because the root pass is the first weld pass, thus the T_{ph} is usually at the specified minimum, and there is no buildup of interpass temperature; the heat input is usually lower than for the fill passes; and high restraint and stress concentration usually exist. Therefore, welding procedures that are developed to prevent HAC in the root pass are usually adequate and may be overly conservative for the remaining passes in the weldment. It was suggested by Yurioka^{6, 34} that the T_{ph} from Tekken tests are too conservative, and hence should be reduced because of the factors mentioned in the background section.

Therefore, critical T_{ph} obtained from the Tekken test results should corrected as function of restraint and weld metal yield strength as shown in Figure 16. Similar restrictions apply for Lehigh weldability test results. The corresponding correction will be lower in Lehigh specimens because R_F is considerably lower than the Tekken test.

If the purpose of a weldability test is to develop suitable welding procedures for a particular application, it is important to use a test that closely simulates the actual weldment in terms of welding process, joint geometry, restraint, cooling rate, welding consumables, and welding conditions, and permits determination of the effects of variations in welding procedures³⁶. This is one of the basic reasons why there is a plethora of weldability tests to simulate different joints. On the basis of joint geometry and restraint conditions, fillet welds should be assessed by using CTS tests. Similarly, Lehigh and Tekken weldability tests were developed to simulate root pass weld in a butt joint at mid-thickness. If Lehigh and Tekken tests are used to simulate V-groove joints, then it should be noted that the stress state at the weld root will be different in these two cases. For example, the bead eccentricity also affects the residual stress state at the weld root. According to the FEM results from Suzuki, Figure 43⁶², as the weld root deviates from the neutral axis (N.A.) of the plate the transverse residual stress state changes to either compressive or tensile in nature, due to the plate bending along the N.A. When the weld bead is below the N.A., the plate bending causes the transverse residual stress state to be compressive at the weld root.

The Lehigh and Tekken test specimens, hence, do not simulate HAC behavior of practical joints, such as V-groove, single bevel groove root passes. In these joints, the root pass is at the bottom of the groove, and, according to Figure 43, the transverse residual stresses will be compressive. In contrast, in Lehigh and Tekken test specimens, transverse residual stress distribution will be tensile. This difference in transverse residual stress distribution at weld root will influence HAC susceptibility significantly. In fact, one of the prerequisites for HAC is tensile

stresses. If tensile stresses are absent then HAC initiation will be prevented. This suggests that the weldability tests should be selected such that it simulates hydrogen, residual stress and microstructure distribution of actual weldments as closely as possible.

Determination of Interpass Temperatures in Multipass Welds

The preheating temperatures determined from the root-pass cracking tests are too conservative to be used as interpass temperatures in multipass welding. This is because of two reasons:

- The restraint on the root pass is maximum and subsequent weld passes do not experience large restraining force.
- Hydrogen accumulation in the weld decreases and so does the hardness of the microstructure in the weld due to the tempering effect of the post pass heating due to subsequent passes.

These effects lower the cracking susceptibility of the subsequent passes in multipass welds. The interpass temperature can thus be lower than the critical preheating temperature for avoiding root cracking. Suzuki(26) indicates that interpass temperature can be 25°C lower than that determined for avoiding root cracking in the y-groove (Tekken) test. Yurioka et al. suggest that toe cracking and underbead cracking are prevalent in multipass welds. Therefore, interpass temperature can be determined for avoiding toe cracks using the same technique as developed for predicting preheating temperature for fillet welds(32). In case of high-strength weld metal, Yurioka suggests that the method described earlier for avoiding weld metal hydrogen cracking can be used for determining safe interpass temperatures.

Application of Uniform Versus Local Preheating

Preheating can be applied locally or uniformly. Uniform preheating is usually not practical in welding of large sections, and hence localized preheating is applied. Extra precaution is necessary when preheating locally, as localized preheating will have higher weld cooling rates as com-
pared to uniform preheating because of the heat extraction effect of the surrounding mass which is at ambient temperature. Thus hydrogen evolution from the welds is suppressed resulting in more residual hydrogen at 100°C.

Most of the prediction schemes predict uniform preheating temperatures based on the results of uniformly heated weldability specimens. Thus when using experimentally determined, or predicted preheating temperatures in actual welding practice the localized preheating temperatures should be adjusted such that $t_{15/1}$ in local preheating becomes equal to $t_{15/1}$ in uniformly preheated condition. Charts for determining $t_{15/1}$ for various preheating conditions have been developed by Nippon steel³⁴.

APPLICABILITY OF PREHEAT PREDICTION TECHNIQUES

A major problem with using non-experimental techniques to determine T_{ph} is that one must make questionable assumptions about a variety of input parameters. Quite probably one of the most critical is the assumed concentration of hydrogen. The first problem that is encountered is whether one uses "initial" pre- solidification weld metal hydrogen content or post-weld residual (weld metal or HAZ) hydrogen content. The British Standard Method is based on initial hydrogen content, while the modified Cottrell methodology along with Yurioka's techniques indirectly considers the effect of residual hydrogen.

Weld hydrogen measurements with some of the commercially available welding consumables are presented in Table 34 along with the welding conditions employed. Table 34 clearly shows that there is marked difference in the expected H_{DM} values suggested by the manufacturer and actual measured H_{DM} values. This difference is substantial at all the H_{DM} levels analyzed. The reasons for these differences may be different, but the purpose of this exercise was to show that H_{DM} levels in actual weldments will be different than the expected H_{DM} values. In such cases, if the preheating temperatures are determined using the expected H_{DM} values, then the predicted preheating temperatures may be too low to avoid HAC in actual weldments. This suggests that it is necessary to determine H_{DM} content under actual welding conditions before proceeding to calculate T_{ph} using preheat prediction schemes.

To predict actual weldment T_{ph} , the preheating prediction method should also consider all the variables encountered in an actual weldment, such as, joint size, restraint, preheating method, ambient temperature, bead eccentricity, H_r , stress concentration, filler metal strength (under-over matching). Yurioka's method considers the effect of all these variables and hence is the most comprehensive. In this respect, it is superior to the TWI/BS, Cottrell's and AWS D1.1 prediction schemes.

CRACK PROPAGATION IN TEKKEN SPECIMENS

Numerical Study

The results in Table 32, indicate that the $\frac{K_{II}}{K_I}$ ratio is independent of applied pressure for a particular crack configuration, i.e. for a given crack length at a given crack angle. This is in accordance with the fact that the stress intensity factors are unique for a given crack configuration. This suggests that the simple method used in this study of applying tensile stresses to simulate crack propagation behavior under the residual stress distribution in the Tekken specimen is valid. It is observed from Table 32 that as the crack length increases the $\frac{K_{II}}{K_I}$ ratio also increases. This also suggests that the mode II component increases with increasing crack length. The rate of this increase is a function of the crack angle, Figure 39. For cracks at an angle of 15°, the $\frac{K_{II}}{K_I}$ ratio increases initially and then becomes almost constant, Figure 39. In comparison, cracks at 30° and 45° angles the $\frac{K_{II}}{K_I}$ ratio continually increases with the crack length, Figure 39.

As mentioned in the background section, the crack undergoes a change in the crack propagation direction under mixed mode conditions. Several equations have been proposed which

predict the angle of the kink as a function of the $\frac{K_{II}}{K_{T}}$ ratio. The exact angle at which kinking occurs is determined by the fracture criterion operating. Several different fracture criteria have been proposed. It was suggested by Gdoutos and Aifantis⁷⁴ that the crack propagation occurs in the direction of maximum hydrostatic tensile stress. Chen and Nakasa⁶⁹ also proposed a similar crack propagation and crack branching criterion. In a later study, however, they found that the crack propagation angle in HAC is different than that proposed by the maximum hydrostatic tensile stress criterion for higher $\frac{K_{II}}{K_I}$ ratios. They proposed a new criterion, according to which crack propagation in HAC occurs when the hydrogen assisted cracking function reaches the cohesive strength of the material. The crack propagation angles predicted by their new criterion matched closely with the experimentally determined crack propagation angle. Their analysis suggests that the crack propagation angle depends not only on the $\frac{K_{II}}{K_I}$ ratio but also on KI, material constants, such as yield strength and elastic modulus, and hydrogen concentration. Similar equation predicting the crack propagation angle under mixed was proposed by Gdoutos and Aifantis⁷⁴. These authors also proposed that crack propagation occurs in the direction of maximum concentration of corrosive species. Maccagno and Knott⁷² observed that maximum tensile stress criterion for crack propagation is not valid for the case when crack tip plasticity is involved. They concluded that criteria based on the assumption of linear elastic fracture do not apply to engineering alloys.

From the above discussion, it is clear that an exact fundamental explanation for crack propagation under mixed mode loading conditions is still being debated. Nevertheless, the effect of mixed mode loading on crack propagation angle is clear. This suggests that cracks emanating along the HAZ from the double-bevel side notch root of the Tekken specimen will be under mixed mode loading conditions. The angle at which the crack propagates initially will be decided by the shape of the weld fusion line as this determines the shape of the CGHAZ region, since crack will prefer the more susceptible CGHAZ region. The nucleated crack, depending on the angle with respect to the thickness direction, will be subjected to a varying degree of mixed mode loads arising from the residual stress distribution. The crack propagation direction will change into the weld metal after a critical $\frac{K_{II}}{K_{I}}$ ratio has been reached. This hypothesis explains the most commonly observed, Type 2, crack propagation behavior.

The question then arises is why Type 1 crack propagation mode is observed at all. This can probably be explained as follows. The conditions are such that the nucleated crack makes only small angle with respect to the plate thickness direction, e.g. less than 15°. As can be seen from Figure 39, the $\frac{K_{II}}{K_{I}}$ ratio in this case does not rise above 0.13. Therefore, for these type of crack orientations, the K_{II} component is not high enough to influence the crack propagation behavior at all. Another reason could arise from microstructural viewpoint. If the coarse grained HAZ is highly susceptible to HAC then the crack propagation will not be influenced by mode II loading to a great extent. This is especially true when the weld microstructure is less resistant to HAC. A third factor could be the crack propagation rate. If the crack propagation rate is high then the crack will propagate along the original direction before mixed mode loading can take part in crack propagation.

Experimental Results

It was observed from the macrographs, Figure 33-35, that crack propagation occurred in a direction parallel to the thickness direction along the fusion line on the double bevel side in specimens welded at RT and 175° C with a H_{DM} of 16.18. This type of crack propagation behavior is different from the most commonly observed Type 2 crack propagation behavior, Figure 19. In Tekken specimens welded with a H_{DM} of 16.18 at RT and at a T_{ph} of 130°C, type 2 crack propagation mode was observed Figure 33 and 34.

As discussed earlier based on the metallography results, the location of weld notch root in combination with the microstructure primarily determines the crack initiation in Tekken specimens. In the present study, the filler metal used had tensile strength matching with the base metal, i.e. 110 ksi (757 MPa). Therefore, the weld metal microstructure is also highly susceptible to HAC. In addition, the hydrogen concentration in the weld metal is higher in the beginning until it diffuses into the HAZ or escapes to the atmosphere under the action of concentration gradient. This also makes the weld metal equally susceptible to HAC. The presence of the sharp notch at the fusion line causes stress concentration, and hence HAC initiates at the fusion line. The nucleated crack then passes through the HAZ for some distance before entering the weld metal again.

The above discussion suggests that mechanical factors are equally influential in HAC initiation location along with the microstructural factors. The shape of the notch root on the double bevel side controls whether the crack initiates in the HAZ or in the weld metal. This suggests that for successful application of Tekken test, one needs to carefully select the welding parameters such that the weld notch root shape will not promote cracking in the weld metal. The Tekken test in itself does not cause HAC to occur in the HAZ, rather the interaction between the mechanical factors introduced through the weld notch root and the difference in the microstructure of HAZ and the weld metal governs the crack initiation site. The whole philosophy behind Tekken testing, therefore, should be analyzed in detail.

The present JIS 3158 standard⁹⁴ does not specify the welding procedure in detail for a successful application of the Tekken test to HAC susceptibility assessment of the HAZ, although in Japan, Tekken test is conducted under the standard conditions of $H_{DM} = 5$, Heat input = 1.7 kJ/mm and ambient temperature of 10°C. Wingrove et al^{67,68}. after a detailed parametric study of the Tekken test, found that welding voltage, current, speed, notch root shape and the root gap affected the HAC occurrence. In addition, they attributed the weld metal microstructure being the main reason behind HAC occurrence in the weld metal. They concluded that just the

specification of weld heat input in Tekken weldability test is not enough, rather all the welding parameters should be specified.

Detailed metallographic examination indicated that in the specimens welded at RT and 130°C, Type 2 crack propagation mode was initially observed followed by a change in the crack propagation direction. For the Tekken specimen welded at RT, the initial crack angle was 65° and the crack propagated 0.2 mm before experiencing a change in the crack propagation direction. In the Tekken specimen preheated to 130°C, the initial crack angle was 50° and the crack propagated for 0.7 mm before undergoing a change in crack propagation angle. Type 2 crack propagation mode was also observed with Tekken specimens welded with SAW and FCAW, Figure 44(a) and (b). It is observed from Figure 44(b) that crack initiated on the single bevel side of the Tekken specimen in the weld metal. The crack then entered the HAZ and propagated along the HAZ for a certain distance before changing its direction into the weld metal. Figure 44(a) and (b) demonstrates the case where crack initiation occurred under mixed mode and, as a result, the crack experienced kinking. In this case, the $\frac{K_{II}}{K_{I}}$ ratio for the sharp notch was high enough to force a change in crack propagation direction.

All these results are plotted along with the numerical results in Figure 39. It is observed from Figure 39 that the crack propagation angles and crack lengths in experimental Tekken tests closely follow the numerically predicted behavior. The variation in the threshold $\frac{K_{II}}{K_{I}}$ values as a function of different crack configuration suggests that other variables, such as microstructure and joint mis-match, also play a significant role in determining the crack propagation angle.

2.5 - CONCLUSIONS

- The weldability test results indicate that Tekken and Lehigh weldability test specimens can be successfully used to assess HAZ and weld metal susceptibility to HAC, respectively.
- Safe preheating temperatures for avoiding single pass weld metal HAC can be be predicted by modifying the existing Nippon steel method.
- 3) Yurioka's method for avoiding HAC in the HAZ is too conservative at lower weld hydrogen contents, but predicts reasonably well the safe T_{ph} to avoid HAC in the HAZ in case of Tekken specimens. Yurioka's method is most applicable in practice as it incorporates the effect of over/undermatching of filler metals and the restraint levels.
- 4) Preheating temperatures from AWS D1.1 and Cottrell's prediction methods do not avoid HAC occurrence in Tekken specimens. In addition, these preheating temperatures are too low to avoid even weld metal cracking in Lehigh specimens.
- 5) Weld root penetration variation along the weld length markedly influences the localized stress concentration, and hence the HAC susceptibility. This effect is more pronounced in Lehigh specimens as the stress concentration at crack initiation location was almost negligible with complete root penetration.
- 6) Crack initiation in Tekken specimens was highly influenced by the microstructure and weld root notch shape in the vicinity of weld root notch on the double bevel side.
- 7) Acoustic emission signature was unaffected with a variation in weld hydrogen content and T_{ph} for Tekken and Lehigh specimens. This was also supported by the observation that the fractographic modes were unchanged with weld hydrogen content and T_{ph}.
- 8) Finite element analysis indicated that crack propagation in Tekken specimens is under mixed mode loading condition for the most typical crack propagation mode, Type 2. The crack propagation angle changes depending on the microstructure and hydrogen content.

The experimentally observed crack propagation behavior matches well with numerically predicted behavior.

9) This study effectively demonstrated the differences between weldability testing procedure and actual weldments. The weldability test results can not be directly extrapolated to actual weldments and in this sense the present day weldability testing procedures are inadequately designed.







Groove type	Kt
y (root)	4
Double-Vee (root)	3.5
Y (root)	 4~5
Single-bevel (root)	6~8
V (root)	1.5
y, x, Y, V, U (toe)	1.5

Figure 13. Notch Stress Concentration Factor

From Yurioka (6)



Figure 14. Lehigh Restraint Test Specimen (without saw cuts)



Figure 15. Tekken oblique y-groove Restraint Test Specimen



Figure 16. Correction of Preheat Temperature by Restraint and Strength From Yurioka (55)



Figure 17. Master Curves for Determining Preheat Temperature from CEN and Plate Thickness, From Yurioka (55)





(b)

Figure 18. Correction of CEN Depending on, (a) Heat Input (b) Weld Hydrogen Content





:



(b) Type 2



(c) Type 3

Figure 19. Different Types of Cracking Modes Observed in Tekken Specimen (a) Type 1, (b) Type 2, (c) Type 3





(b)





Figure 21. Acoustic Emission Monitoring Set-up



Figure 22. Schematic of Weldability Test Set-up



Figure 23. Tekken Specimen Configuration& Boundary Conditions to Simulate Crack Propagation Behavior Using FEM



Figure 24. (a) Typical Root Penetration in Welded Lehigh Specimens



Figure 24. (b) Root Penetration in Welded Lehigh Specimens



Weld Hydrogen Content, H_{DM}









Figure 27. Longitudinal Crack Propagation Rate Variation with Weld Hydrogen Content for Lehigh Specimens Welded at RT



Weld Hydrogen Content, H _{DM}

Figure 28.

Comparison between Test and Predicted Preheating Temperatures for Tekken Test Specimens



Figure 29. Hardness Variation across the Weld Zone - Tekken Specimen Welded at RT with 1% H2 Shielding Gas



Figure 30. (a) Coarse Grained HAZ Hardness Variation with Preheating Temperature - Tekken Specimens







Figure 31. Weld Metal Hardness Variation with Preheating Temperature - Lehigh Specimens



Figure 32.

(a) Macrograph Showing a Typical HAC Occurrence, X10
(b) Higher Magnification Macrograph at the Same Location, X50
- Lehigh Specimen Welded with a H_{DM} of 16.18 at 150°C







(b)

Figure 33.

(a) Macrograph Showing a Typical HAC Occurrence, X10
(b) Higher Magnification Macrograph at the Same Location, X100
- Tekken Specimen Welded with a H_{DM} of 16.18 at RT





Figure 34.

(a) Macrograph Showing a Typical HAC Occurrence, X10
(b) Higher Magnification Macrograph at the Same Location, X100
- Tekken Specimen Welded with a H_{DM} of 16.18 at 130°C





(b)

(a)

Figure 35. (a) Macrograph Showing a Typical HAC Occurrence, X10
(b) Higher Magnification Macrograph at the Same Location, X200
- Tekken Specimen Welded with a H_{DM} of 16.18 at 175°C



(b)

Figure 36. Fractograph Showing Crack Initiation Region near the Weld Root for a Tekken Specimen Welded with a H_{DM} of 16.18 at RT (a) X1000, (b) X500





(b)

Figure 37. (a) Fractograph Showing Crack Initiation Region near the Weld Root for a Lehigh Specimen Welded with a H_{DM} of 16.18 at RT, X1000 (b) Quasi-Cleavage Region next to the Crack Initiation Region in Tekken Specimen Welded with a H_{DM} of 16.18 at RT, X 500





(b)

Figure 38.

(a) Fractograph Showing Columnar Texture in a Lehigh Specimen Welded with a H_{DM} of 11.79 at RT, X200
(b) Quasi-Cleavage and Intergranular Fracture Mode in a Lehigh Specimen Welded with a H_{DM} of 11.79 at RT, X 500



Figure 39. $\frac{K_{II}}{K_{I}}$ Variation with Crack Angle for Different Crack Lengths (Tekken)





(b)

Figure 40. (a) Root Penetration at Weld Start, X7.5 (b) Root Penetration at Weld End, X7.5






Figure 42. CCT Diagrams for A514 Grade B Structural Steel
(a) 0.5°C/sec Heating to 900°C, 15 Minute Hold at 900°C
(b) 300°C/sec Heating to 1250°C, 12°C/sec Cooling to 900°C

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Figure 43. Residual Stresses Across Weld Throat, From Suzuki (62)



(a)



(b)

Figure 44. (a) Macrograph of A588 Steel - 16° Crack Angle, 2.7 mm Long Crack, X7.5
(b) Macrograph of a Quenched and Tempered Steel - FCAW, 30° Crack Angle, 5.3 mm long Crack, X7.5

Welding Method	Welding Material	H _{JIS} (Glycerin) (ml/100 g DM)	H _{IIW} (Mercury) (ml/100 g DM)
Shielded Metal Arc Welding	Cellulosic Ilmenite Low-hydrogen	35 25 1-5	40 30 2-10
Gas Shielded Arc TIG MIG CO ₂		0.2	1.0
Submerged Arc	Ultra-low-hydrogen Conventional	0.3-1.0 1-6	2-10

Table 19 : Diffusible Hydrogen Content in Weld Metal

Table 20 : Restraint Coefficients in Practice

Coefficient of Restraint kgf/ mm ³	Restraint Severity	Practical Examples
70.h	Severely Restrained	Pipe Welding
(Tekken Test)	Root Pass	Repair Welding
40.h	Normally Restrained	Normal Construction
(Lehigh Test)	Root Pass	Bridge Construction
20.h	Weakly Restrained Root Pass	

h = Thickness in mm

Table 21. Carbon Equivalents to Assess Weldability

Group	Formula
	$CE (IIW) = C + \frac{Mn}{6} + \frac{Cu + Ni}{15} + \frac{Cr + Mo + V}{5}$
А	$CE (WES) = C + \frac{Si}{24} + \frac{Mn}{6} + \frac{Ni}{40} + \frac{Cr}{5} + \frac{Mo}{4} + \frac{V}{14}$
	$CE (Stout II) = C + \frac{Mn}{6} + \frac{Cu}{40} + \frac{Ni}{20} + \frac{Cr}{10} + \frac{Mo}{10}$
	$P_{CM} = C + \frac{SI}{30} + \frac{Mn}{20} + \frac{Cu}{20} + \frac{Ni}{60} + \frac{Cr}{20} + \frac{Mo}{15} + \frac{V}{10} + 5B$
В	$CE (Graville) = C + \frac{Mn}{16} + \frac{Ni}{50} + \frac{Cr}{23} + \frac{Mo}{7} + \frac{Nb}{8} + \frac{V}{9}$
	$CE (D\tilde{u}ren) = C + \frac{Si}{25} + \frac{Mn}{16} + \frac{Cu}{16} + \frac{Ni}{60} + \frac{Cr}{20} + \frac{Mo}{40} + \frac{V}{15}$
	$CE (Stout I) = 1000 \cdot C \cdot (\frac{Mn}{6} + \frac{Cr + Mo}{10} + \frac{Ni}{20} + \frac{Cu}{40})$
С	$CEN = C + A(C) \cdot \left\{ \frac{SI}{24} + \frac{Mn}{6} + \frac{Cu}{15} + \frac{Ni}{20} + \frac{Cr + Mo + Nb + V}{5} + 5B \right\}$
	Where $A(C) = 0.75 + 0.25 + \tanh \{20(C - 0.12)\}$

MATERIAL	С	Mn	Р	S	Si	Cu	Ni	v	Al	Cr	Мо	Ti	В
A514	.19	.94	.008	.004	.31	.19	.10	.03	.039	.52	.21	.030	.0018
Filler Material	.06 .08	1.20 1.35			.30 .40		2.40 2.70			.30 .50	.40 .50		

 Table 22.
 Chemical Composition, Mechanical Properties and Heat Treatment Details

MATERIAL	Y.S. (ksi)	T.S. (ksi)	% Elongation	% Reduction
A514	117	126	32	43
Filler Material	110 115	120 125	15 18	50 65

HEAT TREATMENT SCHEDULE FOR A514-GRADE B

AUSTENITIZE AT 1700°F FOR 60 MINUTES - WATER QUENCH

TEMPER AT 1240°F FOR 60 MINUTES

Welding Pa	rameters	Shielding Gases	H _{DM}
Current	330 amps	Ar- 2%O ₂	2.87
Voltage	31 volts	Ar- 2.53%CO ₂ - 0.269%H ₂	8.67
Welding speed	5.28 mm/sec	Ar- 2.65%CO ₂ - 0.51%H ₂	11.7
Efficiency	0.80	Ar- 2.7%CO ₂ - 1%H ₂	16.18
Heat Input	1.55 kJ/mm		

Table 23 : Welding Parameters and Shielding Gases Used

Table 24 : Acoustic Emission Parameters Used in this Study

Sensors	
Resonant Frequency	375 kHz
Threshold	1.00 Volts - Fixed
Clock Period of	
Locator Clock	1000 Nanoseconds
Event Duration Clock	1000 Nanoseconds
Rise Time Clock	250 Nanoseconds
Range of	
Amplitude	50-117 dB
Rise Time	0-65520 microseconds
Slope	0-65520 mv/microseconds
Event Duration	0-65520 microseconds
Ringdown Counts	0-4096
Energy	50-165
Fixed Preamplifier Gain	60 dB
Total System Gain	70 dB

Weld Metal Hydrogen, H _{DM}	Test Temperature ^o C	$ \begin{array}{c} Cracking Ratio \\ \hline l_c \\ \hline l_w \\ \hline h_w \\ \hline \end{array} $		Time for Crack Initiation	Crack Completion Time
2.87 ml	R.T.	0	0	-	-
	R.T.	1	1	360 Seconds	2288 Seconds
8.67 ml	75°C	.6	1	3780 Seconds	52 Hours
	100°C	0	0	-	-
11.701	R.T.	1	1	284 Seconds	1214 Seconds
11.79 mi	125°C	0	0	-	-
	R.T.	1	1	172 Seconds	458 Seconds
16.18 ml	125°C	1	1	2 Hours 16 min	6 Hours
	150°C	1	.8	> 72 Hours	N.A.
	170°C	0	0	-	-

Table 25 : Lehigh Weldability Test Results

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Weld Metal	Cracking	Crack	Crack Propagation
Hydrogen, H _{DM}	Time, Seconds	Length, mm	Rate, mm/Sec
2.87 ml 8.67 ml 11.79 ml 16.18 ml	1988 930 286	0 125 125 125	0.063 0.134 0.437

Table 26 : Longitudinal Crack Propagation Rate Variation with $H_{\mbox{\scriptsize DM}}$ at R.T.

Table 27 : Crack Initiation Temperature for Lehigh Specimens Welded at R.T.

Weld Metal Hydrogen, H _{DM}	Preheating Temperature ^o C	Crack Initiation Temperature ^o C
8.67	R.T. 75°C 100°C	62°C 54°C No Cracking
11.79	R.T. 125°C	64°C No Cracking
16.18	R.T. 125°C 150°C 170°C	80°C 40°C R.T. No Cracking

Welding Details	Events Statistics Accepted Events	Peak Amplitude dB	Energy	Slope	Events Vs Time
0.269% H_2 at Room Temp.	1382 More in 2nd Half	60 - 98	55-140	1-180	First event in 2nd Half
0.269% H ₂ at 75°C	238 More in 2nd Half	60 - 98	72-126	1-225	First events from both Half
$\begin{array}{c} 0.51\% \text{ H}_2\\ \text{at Room Temp.} \end{array}$	937 More in 2nd Half	60 - 100	62-150	1-180	First event in 2nd Half
1% H ₂ at Room Temp.	846 More in 2nd Half	60 - 98	40-126	1-1900	First event in 2nd Half
1% H ₂ at 125°C	713 More in 2nd Half	60 - 98	64-126	1-1700	First event in 1st Half
1% H ₂ at 150°C	N.A.	N.A.	N.A.	N.A.	N.A.

Table 28 : Acoustic Emission Results from Lehigh Weldability Tests

Table 29 : Tekken Weldability Test Results

Weld Metal Hydrogen, H _{DM}	Test Temperature ^o C	$\begin{array}{c} Cracking \\ Ratio \\ \hline l_c \\ \hline l_w \\ \hline h_w \end{array}$	Time for Crack Initiation	Crack Completion Time
2.87	R.T.	0 0	-	-
16.18	R.T.	1 1	164 Seconds	400 Seconds
16.18	130	1 1	2 Hours 7 Minutes	4 Hours
16.18	175	1 .97	> 24 Hours	N.A.
16.18	195	0 0	-	-

 $\frac{l_c}{l_w}$ - Longitudinal Crack Ratio, $\frac{h_c}{h_w}$ - Sectional Crack Ratio

Welding	Events Statistics	Peak	Energy	Slope	Events Vs
Details	Accepted Events	Amplitude dB			Time
1% H ₂	822	50 - 100	55-128	1-1900	First event
at Room Temp.	More in middle				in 2nd Half
1% H ₂	713	50 - 86	55-128	1-1800	First event
at 130°C	More in 2nd Half				in 2nd Half

Table 30 : Acoustic Emission Results from Tekken Weldability Tests

Table 31 : Plate Width Effect on Weld Cooling Time

Weld Cooling Time, Seconds	Standard Lehigh 300x200 mm	Wide Lehigh 300x400 mm
1500 to 100°C	170	100
1500 to 45°C	2690	584
1500 to 25°C	4500	1100

CRACK	PRESSURE	$\frac{K_{II}}{K_{I}}$		
	IVII a	15°	30°	45°
.0005	.9 E+09	.005	.14	.29
	.8 E+08	.005	.14	.29
	.7 E+07	.005	.14	.29
.001	.9 E+09	.048	.20	.36
	.8 E+08	.048	.20	.36
	.7 E+07	.048	.20	.36
.002	.9 E+09	.105	.25	.40
	.8 E+08	.105	.25	.40
	.7 E+07	.105	.25	.40
.003	.9 E+09	.13	.28	.45
	.8 E+08	.13	.28	.45
	.7 E+07	.13	.28	.45
.004	.9 E+09	.14	.30	.47
	.8 E+08	.14	.30	.47
	.7 E+07	.14	.30	.47
.005	.9 E+09	.145	.32	.50
	.8 E+08	.145	.32	.50
	.7 E+07	.145	.32	.50
.006	.9 E+09	.147	.34	.53
	.8 E+08	.147	.34	.53
	.7 E+07	.147	.34	.53

Table 32 : $\frac{K_{II}}{K_{I}}$ Ratio for Cracks Oriented at different angles from the Notch

Weld Hydrogen Content, H _{DM}	Hydrogen Grouping	Susceptibility Index	Tekken Restraint High	Lel Rest Medium	high raint Low
2.87	H1	D	138°C	110°C	80° C
8.67	H2	E	150°C	130°C	110°C
11.79	H3	F	160°C	150°C	138°C
16.18	H3	F	160°C	150°C	138°C

Table 33 : Preheat Temperatures from AWS D1.1-88 : Hydrogen Control Method

The carbon equivalent, $P_{\rm cm},$ is = 0.31 For the A514 Steel

Table	34:	HDM	for	Different	Consumables -	AWS	A4.	3-86	Method
1 4010		I JIVI	101	~ III VI VIIV	COND WILLWOLDD			~ ~ ~ ~	

Welding Condition	Expected H _{DM}	Measured H _{DM}	
FCAW - (E70T-1) (3/32" Wire) - 100% CO ₂	12	17.4	
Hydrogen Controlled FCAW - (E70T-1) (3/32" Wire) - 100% CO ₂	8	11.3	
Extra Low Hydrogen FCAW - (E70T-1) (3/32" Wire) - 100% CO ₂	< 2	2.6	
Extra Low Hydrogen Metal Cored - (E110 C-G) (1/16" Wire)- Ar-5% CO ₂	< 1	1.4	

Heat Input = 1.7 kJ/mm

CHAPTER 3

NUMERICAL SIMULATION OF HYDROGEN DIFFUSION

3.1 - BACKGROUND

INTRODUCTION

Hydrogen Assisted Cracking (HAC) in weld metal as well as heat affected zone (HAZ) occurs at temperatures lower than 150°C, hence, the residual diffusible hydrogen, H_r, correlates more closely²⁸ with the critical stress required for cracking than the initial hydrogen, H_o. The H_r remaining in a weld joint at a given instant is quantified by hydrogen outgassing relations. Hydrogen outgassing behavior for any weld joint is expressed via a relation between residual hydrogen fraction, $\frac{H_r}{H_o}$, and the outgassing heat treatment, which is expressed in terms of the thermal factor, D Δt . From these hydrogen outgassing relations; 1) instantaneous H_r content in a weld can be predicted, 2) thermal cycles, hence the welding conditions, to lower the H_r to a harmless level can be designed, and 3) hydrogen outgassing behavior of any groove/joint geometry. Therefore, $\frac{H_r}{H_o}$ ratio in a weldability test could be different than that of an actual weldment. As a result, the HAC susceptibility assessment of actual weld joints using weldability tests could be incorrect. It was, therefore, felt that hydrogen outgassing behavior of different joints needed to be analyzed in detail.

Hydrogen outgassing relations are generally determined; 1) experimentally, 2) using analytical equations or 3) numerically. The earlier efforts exclusively used analytical solutions to study hydrogen diffusion from simple geometries^{22, 95}. Later on, the emphasis shifted to experimental determination of hydrogen outgassing behavior^{11, 28, 96}. Application of numerical methods to hydrogen diffusion problems has been attempted by various researchers since mid-seventies and the complexity of the problems being tackled has increased with better understanding of numerical approaches in solving difficult problems.

ANALYTICAL AND EXPERIMENTAL APPROACHES

The diffusion of mass in a uniform medium is governed by Fick's second law :

$$\frac{\partial C}{\partial t} = D \frac{\partial}{\partial x} \frac{\partial C}{\partial x} + \frac{\partial}{\partial y} \frac{\partial C}{\partial y}$$
(9)

Where C = Concentration of a species, t = time, D = Diffusion Coefficient.

For simple initial and boundary conditions, analytical solutions exist that are used to describe hydrogen diffusion in steel welds. Coe⁹⁵ used an analytical solution to Fick's second law to calculate hydrogen removal heat treatments for simple specimen configurations. Coe and Chano²² solved one dimensional diffusion equation to obtain the hydrogen concentration at the fusion line and the residual hydrogen remaining in the weld. This equation can be applied for BOP type of weld but can not be accurately applied to evaluate the concentration at the root of the weld in a grooved specimen configuration. This is because the hydrogen concentration at the weld root must be considered as the boundary condition. Terasaki, et al⁹⁷, used the same concept to determine the optimum BOP specimen size using an analytical solution. These type of analytical solutions are available only for simple geometries. To characterize hydrogen outgassing behavior of complex weld geometries, different researchers resorted to experimental determination of hydrogen outgassing relations.

Mcparlan and Graville⁹⁶ experimentally obtained $\frac{H_r}{H_o} = A \exp(-75 D\Delta t)$ type equations to describe the mean hydrogen concentration remaining after time, Δt . Terasaki, et al²⁸, experimentally developed hydrogen outgassing relations to calculate H_r in Y-groove and BOP welds after a given heat treatment. Their experimental procedure involved H_{r200}, H_r at 200°C, measurements using collection over mercury. Matsada, et al¹¹, also developed hydrogen outgassing relations

for Longitudinal Bead - Tensile Restraint Cracking (LB-TRC)) test for different steels. Their experimental procedure involved H_{r100} , H_r at 100C, measurement in LB-TRC specimens using the gas chromatograph method. Hydrogen outgassing relations for various steel compositions were developed.

Experimental determination of the hydrogen outgassing relations is the best method of quantifying residual diffusible hydrogen content of a weld. The obvious drawback is that different specimen configurations demand separate determinations, which is difficult and is not always possible, because of the lack of instrumentation. Additionally, larger specimen sizes may require special instrumentation. These limitations can be overcome with the use numerical methods such as, the finite element method (FEM) or the finite difference method (FDM). The numerical methods also provide localized transient hydrogen concentration in any specimen configuration.

The analytical solutions to Fick's second law also have limited applicability to weld problems because:

- 1 Complex groove/weld geometries make definition of boundary conditions difficult. Analytical solutions for these geometries are rarely available.
- 2 The hydrogen diffusion coefficient, D_H, is a function of thermo-mechanical and metallurgical history of welded specimens. Most analytical solutions do not incorporate varying diffusivity.

The temperature distribution in a weldment after welding is non uniform. In addition, the different regions of a weldment, namely, weld metal, HAZ and base metal, experience different peak temperatures and cool at different rates, especially in the high temperature regime. Additional contribution to these differences arises from chemical composition differences between weld metal and HAZ. In general, the weld metal chemistry is different than the base metal to avoid HAC occurrence in the weld metal. Furthermore, different thermal histories experienced

by the HAZ induce microstructural gradients. Clearly, the D_H varies with location in the weldment. The analytical approaches can not incorporate varying diffusivities as the solutions to such problems do not exist. Numerical approach overcomes this difficulty to a certain extent.

NUMERICAL APPROACHES

Finite difference method and FEM are the commonly used numerical methods to solve diffusion problems. Due to the unavailability of D_H data for different weld regions, the numerical methods have the only advantage that hydrogen diffusion behavior in complex geometries can be simulated. From the literature review, it was observed that numerical approach to solve weld diffusion problems could be separated into two classifications. The simpler approach neglected stress induced hydrogen diffusion and the other approach incorporated hydrogen accumulation under the action of stress gradients.

Hydrogen Diffusion Modeling Neglecting Hydrogen Accumulation

Non-consideration of hydrogen accumulation due to stress driven diffusion is a more commonly used approach to quantify mean hydrogen content. This approach does not depend on the stress analysis results, and hence is much simpler. It consists of solving Fick's second law using the appropriate numerical scheme and initial and boundary conditions.

Terasaki, et al⁹⁸, determined H_r at the weld roots of different specimen geometries using FDM. The results showed that hydrogen evolution is more rapid from a pipe girth weld than from the root welds of Lehigh slot test and the Tekken test. This was attributed to the wider root surface open to the atmosphere in the pipe girth weld compared to the Lehigh slot test and the Tekken test. Terasaki concluded that local hydrogen related with HAC is determined not by the throat depth, but, by the root face, the root angle and the fusion line shape.

Takahashi et al⁹⁹, calculated hydrogen distribution in multi-pass welding using FDM. They compared the numerical results with that from gas chromatograph measurement on sectioned

specimens and found that peak level of hydrogen is close to that determined by the quenching method.

Following researchers also neglected stress induced hydrogen accumulation. Takahashi, et al¹⁰⁰, investigated post weld heat treatment (PWHT) for hydrogen removal using FDM. Kyte and Chew¹⁰¹ employed FEM to find PWHT for hydrogen removal in a double-V nozzle weld. Terasaki, et al¹⁰², analyzed hydrogen diffusion from multi-pass welds.

Dickehut and Ruge¹⁰³ applied FDM using explicit representation to calculate hydrogen distribution in a weld bead produced by SAW tandem technique. They used hydrogen diffusion coefficients from Sykes¹⁰⁴ and Johnson and Hill's¹⁰⁵ study. Uniform diffusivity in an homogeneous medium was assumed. The numerically calculated H_r values were compared with experimental results on sectioned specimens, and were found to be higher than the experimentally measured value.

Most of these researchers used 2 dimensional (2-D) FE mesh. This is a valid assumption, since hydrogen diffusion in the length direction is negligible, because of almost zero concentration gradient in this direction. With the availability of commercial FEA (Finite Element Analysis) softwares, numerical approach is finding more applications to welding related problems.

Hydrogen Diffusion Modeling Considering Hydrogen Accumulation

Localized hydrogen concentration is influenced by microstructural and geometrical features. For example, after austenitic transformation the HAZ is comprised mostly of hardened martensite with high dislocation density, where as, the weld metal is less hardened consisting mostly of ferritic or bainitic microstructures. Hydrogen tends to diffuse from a weld metal into HAZ to relax lattice expansion caused by highly dense dislocations¹⁰⁶.

Weld root with an acute notch is subjected to tri-axial stress in the as-welded condition; the tri-axial stresses also cause lattice expansion. Hydrogen diffusion is enhanced to the tri-axial

stress sites induced by localized stress and strain fields¹⁰⁶.

Stress induced hydrogen diffusion is important from HAC viewpoint, since delayed crack initiation suggests hydrogen enrichment of localized regions. Hydrogen diffusion under the action of hydrostatic stresses is well researched. To characterize the critical concentration of hydrogen responsible for cracking, stress induced hydrogen accumulation at microstructural and geometrical inhomogeneities should be considered.

Kikuta and Ochiai^{107, 108} used FEM to analyze the non steady state hydrogen accumulation at the notch of a circularly notched bar specimen under the action of applied stresses. Kikuta, et al¹⁰⁹, extended the same formulation and applied it to Y-groove welds. They observed hydrogen accumulation at plastically deformed root and martensitically transformed HAZ. The accumulation ranged from 2 to 7 times that of the initial hydrogen concentration and varied strongly as a function of plastic strain distribution.

Yurioka, et al¹¹⁰, observed that hydrogen accumulates at the HAZ of a weld root close to the open surface, although the problem was solved with the boundary condition that hydrogen concentration is nearly zero on the entire surface.

Andersson^{111, 112} formulated the hydrogen diffusion equation considering different types of hydrogen; i.e. 1) diffusible hydrogen, 2) hydrogen trapped in voids, and 3) permanently trapped or irreversible hydrogen. This approach was used to calculate hydrogen distribution in a bead-on-plate weld considering the distribution of plastic strain and elastic stresses in the weld. It was observed that diffusible hydrogen accumulated to some extent at the HAZ of the weld.

Sleptsov, et al¹¹³, solved hydrogen diffusion in a single-pass and a multi-pass weld by FDM, in which hydrogen concentration was directly obtained by varying the apparent D_H in their analysis depending on the local region of a weld. The introduction of varying D_H in different regions resulted in finding the occurrence of the hydrogen concentration peak in HAZ about two hours after welding. Makhnenko et al¹¹⁴, studied the transient hydrogen accumulation at crack

tips in the HAZ under the influence of residual stresses and found that hydrogen accumulation depended strongly on the magnitude of residual stresses in direction perpendicular to the crack tip. When these stresses exceeded 490 MPa, the hydrogen accumulation at the crack tip became 7 to 8 times higher than the surrounding region within 2-3 hours. Karakhin et al¹¹⁵, solved a diffusion equation incorporating hydrogen solubility as a function of microstructure using FEM. They found that the thicker the weld material, the higher the maximum concentration of hydrogen at 20°C and that preheating has a beneficial effect of lowering the hydrogen accumulation. In addition, ferritic and martensitic weld metal raised the hydrogen concentration in the HAZ, whereas, the austenitic weld metal lowered it since austenite has higher solubility for hydrogen.

One common factor in all these approaches is the assumption of hydrogen diffusivity and solubility dependence on stress distribution. This dependence, though well known, is not completely understood nor is the functionality describing this relationship. Hence, the results from these models are only qualitative in nature and can not be used to quantify critical hydrogen concentration.

The major drawback of all these approaches is the inaccuracy introduced by the stress analysis part. Most of these researchers have used simple 2-dimensional models for residual stress simulation. Whereas, in real welding situations, the stress distribution is three dimensional. Further complications arise from microstructure distribution in the weld metal and the HAZ. Hydrogen trapping and diffusivity is affected to different degrees by the type of microstructure, though exact relations are unknown²⁰. In addition, results from one study can not be extrapolated to other studies since the microstructure and stress distribution differ in these cases.

The effect of stress induced hydrogen accumulation can not be considered in the ANSYS FEA program. By using Sleptsov's approach, the stress induced hydrogen accumulation can be considered indirectly. This requires the weld metal, the HAZ and base metal hydrogen diffusivity data that are not available for all the steel compositions. But, a qualitative idea about localized hydrogen accumulation could potentially be gained from this approach. However, because of the

above mentioned difficulties, it was decided not to consider stress driven hydrogen accumulation. Therefore, all the subsequent hydrogen diffusion modeling efforts incorporated only hydrogen diffusion under concentration gradient.

FINITE ELEMENT ANALYSIS CONSIDERATIONS

Initial Condition

The correct choice of a suitable initial condition is a necessary condition for correct solution since the final results are dependent on the initial and boundary conditions. Various researchers have assumed that hydrogen is uniformly distributed in the weld metal after solidification. This simplifies the task at hand. Gedeon¹² proposed that hydrogen distribution in the molten metal is non uniform with hydrogen absorption occurring around the outer edge of the weld pool (the lower temperature region). But, this absorbed hydrogen will become evenly distributed throughout the pool due to vigorous convection within the pool¹². Gedeon also hypothesized that hydrogen will only absorb onto the top surface of the weld stagnant layer, and that there will be less hydrogen near the root of the weld. The above mentioned hypothesis suggests that initially absorbed hydrogen may not be uniformly distributed.

According to Chew and Wilgross¹¹⁶, hydrogen in the liquid weld pool will be evenly distributed due to the convective motions in the pool. During solidification the excess hydrogen rejected at the solidification front will pile up in the liquid. The diffusion distance calculated using Fick's first law was found to be equal to 0.1 mm. The hydrogen concentration gradient across the weld pool into the solid was assumed to have a form in which the concentration was highest in the "hot spot" and uniform in the rest of the weld pool except at the solidification front where it was assumed to be high due to rejection of hydrogen from solid.

The above discussion supports the hypothesis that hydrogen is not uniformly distributed in the weld. The exact distribution of hydrogen will no doubt provide a better understanding of subsequent diffusion in the solid state and also will help quantify diffusible hydrogen contents responsible for HAC more precisely. Unfortunately, the functionality for hydrogen distribution has not been defined in a suitable mathematical form. Thus representation of initial hydrogen distribution other than uniform distribution is not possible at present. Coe and Chano²², Yurioka¹¹⁰, Dickehut¹⁰³, assumed uniform hydrogen concentration in the weld metal and zero concentration in the HAZ and base metal. Andersson¹¹¹ prescribed a lattice concentration for bead-on-plate and a known concentration of 25 ppm in weld metal in Tekken specimens¹¹⁷. Because of the limited knowledge of hydrogen distribution, it was decided to assume an uniform initial hydrogen concentration in the weld metal.

Boundary Conditions

Two types of boundary condition (bc) can be used: 1) Essential or Drichlet boundary condition, in which the boundary is maintained at a fixed concentration through out the analysis, or, 2) Natural boundary condition, in which surface transfer coefficient or the flow rate is prescribed. As soon as hydrogen reaches the specimen surface, it escapes into the atmosphere and thus the concentration on the free surface is always zero. The surface transfer rate will depend on the concentration gradient across the specimen surface. Ideally, the surface transfer coefficient should be input as a function of hydrogen concentration. Again suitable data are not available, hence, zero hydrogen concentration at the surface of the specimen was assumed, i.e., by specifying essential or Drichlet boundary condition.

Coe and Chano^{22, 103} and Andersson¹¹⁷ assumed zero concentration on the boundary exposed to atmosphere. Yurioka¹¹⁰ and Andersson¹¹¹ used natural boundary conditions and specified surface transfer coefficients that were tantamount to a hydrogen concentration of "zero" at the boundaries.

THERMAL FACTOR DETERMINATION

Hydrogen outgassing behavior is expressed through a relation correlating the $\frac{H_r}{H_o}$ with the

thermal factor, $D\Delta t$. Where, $D = D_H$ for the material under consideration, t = time. This allows hydrogen removal efficiencies to be examined directly by a single parameter within which all variables are combined. $D\Delta t$ summarizes the hydrogen outgassing potential by incorporating the diffusivity at a given temperature and the time spent at that temperature. This concept can be applied to isothermal (PWHT) or continuous cooling (weld thermal cycles) conditions. This concept of thermal factor has been used extensively by different researchers, as discussed earlier.

For any cooling curve, the cumulative thermal factor can be defined as follows :

1) Divide the complete cooling curve into small temperature increments. 2) Select the time intervals, Δt , corresponding to these temperature intervals to calculate the thermal factor for each interval. 3) Calculate the cumulative thermal factor by summation of each of the thermal factor as follows.

$$\sum D\Delta t = D_1 \Delta t_1 + D_2 \Delta t_2 + \dots + D_n \Delta t_n$$
(10)

Since D_H is not known a priori, researchers have used diffusion coefficients for pure iron proposed by Sykes¹⁰⁴ and Johnson and Hill¹⁰⁵. This can lead to erroneous results if applied to any arbitrary steel composition, since, the diffusion coefficients for steels vary significantly from those for iron. Recently Terasaki, et al¹¹⁸, developed a relation to find the thermal factor at 100°C from cooling times for a CTS test configuration. These authors ignored the contribution of temperatures above 1200°C for calculation of the thermal factor. In the region above 1200°C, since iron is austenitic, the D_H is smaller than in ferrite, and furthermore the cooling time from solidification temperature to 1200°C is short. This is particularly useful when thermal cycle data above 1200°C is not available. In this study, the entire thermal cycle was used as this does not pose any additional difficulty.

WELD HYDROGEN DIFFUSIVITY DETERMINATIONS

HAC occurrence can be avoided by limiting the H_r below a critical hydrogen concentration. Quantification of the critical H_r for a given weldment is dependent on the hydrogen diffusivity, D_H , data for the steel weldment. Base metal D_H determination is generally carried out by using electrochemical permeation techniques. In electrochemical permeation, the base metal is hydrogen charged and then hydrogen evolution is monitored at selected temperatures. From the diffusivity data at different temperatures Arrhenius type of relation is established for the base metal. Some researchers¹¹⁹ have used strained and heat treated base metal for hydrogen charging to eventually determine the D_H under these base metal conditions. The D_H , thus determined, is applicable only for the base metal used.

As discussed earlier, during welding, the associated thermo-mechanical changes alter the microstructure and strain histories of the base metal. This structural change alters the D_H of the base metal, i.e. the HAZ. Additionally, the filler metal added during welding has different composition and hence diffusivity than the base metal. Because of these metallurgical and chemical differences, the diffusivity data which are available for a base metal can not be applied to a weldment made up of base metal, HAZ and the weld metal. In other words, weld metal/HAZ/BM have different diffusivities.

The need for exact knowledge of D_H in each of the weld regions can be circumvented, if a mean D_H value for the entire weld specimen is known. The same objective can be achieved by performing individual permeation studies on base metal, simulated HAZ and weld metal and combining the results. Alternatively, D_H measurement using an entire weld specimen can also be performed.

The unavailability of weld D_H data forced researchers to use either Johnson and Hill's data for pure iron or The Welding Institute (TWI) data on different base metals, to quantify hydrogen distribution in a weld. The use of these base metal diffusivities to characterize weld hydrogen content in different steels and in welds will induce substantial errors, since the composition and stress state in the weld deviate significantly from those in the base metal. One such approach by Coe and Moreton¹²⁰ involved determination of D_H for different steel/base metal compositions based on an analytical solution for residual hydrogen fraction, $\frac{H_r}{H_o}$, as a function of the thermal factor, D Δ t, for a simple geometry. The hydrogen evolution from a hydrogen charged cylindrical specimen was monitored as a function of time, and, using the analytical solution, hydrogen diffusivity was determined. Since hydrogen was charged at high temperatures, the diffusivity measurements were strongly influenced by the number of traps introduced during charging. Additionally, diffusivity data generated by this method is for base metals only, and hence can not be applied to welds. Evans¹²¹ used a similar technique for measuring D_H in weld specimens. He monitored $\frac{H_r}{H_o}$ as a function of time for simple weld geometries. The slope of the relation between log ($\frac{H_r}{H_o}$) and time provided the hydrogen diffusion coefficient.

Various attempts to numerically model hydrogen diffusion from different joint geometries by Terasaki⁹⁷, Taksahasi⁹⁹, Kyte and Chew¹⁰¹ and recently by Dickehut¹⁰³ have relied on Johnson and Hill's D_H data, and hence the results from these studies are only qualitative in nature. Clearly, various efforts to quantify hydrogen distribution have been seriously hampered by the unavailability of correct weld D_H data. Therefore, it is necessary to evaluate D_H for a weldment as simply and accurately as possible.

GROOVE CONFIGURATION EFFECT ON HYDROGEN OUTGASSING

It was observed by Terasaki et al⁹⁸, that hydrogen outgassing is strongly dependent on the groove configuration and location. These authors found that hydrogen outgassing is more rapid from a pipe girth weld than from the root pass welds of Lehigh slot test and the Tekken test. These authors concluded that local hydrogen related with HAC is determined not by the throat depth but by the root face, the root angle and the fusion line shape. These results suggest that hydrogen outgassing from different joint geometries can not be generalized. From the viewpoint of application of weldability test results to practical joints, these results are significant.

The widely used practical joints have X-groove, V-groove, K-groove or single bevel groove configurations. These groove configurations permit easier welding. The commonly used

weldability tests, such as, Lehigh and Tekken, have different groove configurations compared to these practical joints. These weldability tests are used to assess root-pass HAC susceptibility of different joints. If the weldability test results are to be directly applied to practical weld joints, it is imperative that hydrogen outgassing behavior of the practical joints be correctly simulated by the weldability tests. No such study exists on hydrogen outgassing behavior of different groove configuration. For a better correlation between weldability test results and practical welding joints, it is necessary to study this effect.

STUDY OF PLATE WIDTH EFFECT

In a generic welding thermal cycle, the weld cooling rate in the high temperature regime, i.e. above 300°C, is generally short and this cooling rate, especially the 800 to 500°C cooling time, $t_{8/5}$, is important from microstructure development point of view. Because of the shorter time span at higher temperatures and austenitic microstructure (which has higher solubility for hydrogen and also lower D_H), the contribution of the thermal cycle above 300°C to hydrogen diffusion is insignificant. The weld cooling rate slows down considerably at temperatures below 300°C. The low temperature cooling rate determines the amount of hydrogen outgassing. Therefore, two welding conditions that have different cooling rates in the low temperature regime will result in different amounts of hydrogen outgassing. In this respect, the effect of plate dimensions, especially width, on weld cooling rates could be significant.

From weldability testing viewpoint, this difference needs to be studied since the commonly used weldability test specimens, such as, Lehigh and Tekken, are smaller compared to the plates encountered in actual fabrication. Since the heat transfer after welding predominantly occurs in a direction perpendicular to the welding direction, the plate width is the chief dimension affecting these cooling rates significantly. The plate width effect on cooling rates in the high temperature regime can be neglected, as the heat transfer is limited to regions in close proximity of the weld. At temperatures lower than 150°C, the welding heat spreads to the regions farther away from the

weld. The wider the plate the more is the heat sink in the width direction, and the faster the cooling rate in the low temperature regime. Thus, it is to be expected that the wider plate will have higher hydrogen content as compared to narrower plate for identical welding conditions.

From the above discussion, it is clear that plate width effect needs to be quantified. This effect has been neglected by different researchers. These researchers have assumed that hydrogen diffusion below 100°C is negligible, based on Johnson and Hill's D_H values. According to recent D_H measurements D_H in steels is much higher than that from Johnson and Hill's data⁸⁴, and hence, hydrogen diffusion below 100°C can not be neglected. In the author's opinion, plate width effect needs to be studied for a closer assessment of HAC susceptibility of practical joints through weldability tests.

OBJECTIVES OF THE STUDY

The important issues in weld hydrogen diffusion studies, overlooked by previous researchers, were studied in detail. The specific objectives of this study were to :

- 1 Experimentally verify the numerically determined hydrogen outgassing relation. This study aimed at establishing confidence in the numerical procedure as applied to weld hydrogen diffusion.
- 2 Numerically develop the hydrogen outgassing relations for Tekken and Lehigh weldability test specimens. These relations were then used to compare the hydrogen outgassing behavior of these weldability test specimens.
- 3 Study the effect of different groove configurations, such as, X-groove, V-groove, Bead on Plate, on hydrogen outgassing behavior. The numerically determined hydrogen outgassing behavior was also compared with Tekken and Lehigh weldability test specimen behavior.
- 4 Study the effect of different thermal cycles on H_r content of a given weld joint. Specifically, study the effect of plate width on H_r content of Lehigh groove configuration.

3.2 - EXPERIMENTAL AND NUMERICAL PROCEDURE

INTRODUCTION

In this chapter, the experimental and numerical analysis procedures used to study hydrogen outgassing behavior are outlined. The H_r content was determined experimentally using the AWS weld hydrogen measurement specimens and the procedure described in Chapter 1. These H_r values were used for verification of the numerically obtained H_r values for the AWS specimens. After the verification of the numerical procedure, additional hydrogen outgassing studies were carried out with different specimen configurations using a similar numerical procedure.

EXPERIMENTAL DETERMINATION OF RESIDUAL HYDROGEN CONTENT

The AWS A4.3-86 diffusible weld hydrogen measurement procedure was employed to determine the H_r at different times. Four A36 specimen test assemblies were subjected to prior degassing at 650°C to remove any residual hydrogen left from the steel making practice. After the degassing treatment, the specimens were de-scaled using dry belt grinding. The center test specimens were weighed to the nearest 0.01 gms. The specimens were cleaned using ethanol and acetone and were hot-air blow dried.

The specimens were Gas Metal Arc welded with a filler wire conforming to ES-120-s1 and a shielding gas containing 0.27 % hydrogen with the welding parameters in Table 3. With these welding parameters, a spray type metal transfer and stable welding arc was obtained. The welded specimens were quenched in iced water within 5 seconds of extinguishing the arc to retain most of the initial hydrogen and agitated vigorously for 20 seconds and stored in liquid nitrogen for subsequent processing. Weld hydrogen measurements were performed according to the procedure described in Chapter 1. But in this case, H_r was determined instead of the initial hydrogen, H_o. H_r measurements were carried out by outgassing for shorter time intervals at room temperature and measuring the evolved hydrogen corresponding to these intervals. This procedure was repeated until no further evolution of hydrogen was observed. After the complete evolution of hydrogen, total or initial hydrogen, H_o, was calculated by adding up the evolved hydrogen at different time intervals. From the total hydrogen and evolved hydrogen at different time intervals, corresponding $\frac{H_r}{H_o}$ values were determined for each specimen.

NUMERICAL DETERMINATION OF HYDROGEN OUTGASSING BEHAVIOR

The hydrogen outgassing behavior for different specimens was numerically determined using a similar concept. The generic numerical procedure along with the assumptions, the initial and boundary conditions, the solution procedure and the residual hydrogen determination is discussed below. Following this, the FE mesh is described for each specimen.

The Finite Element Procedure

Hydrogen diffusion calculations were performed using a commercial FEA (finite element analysis) program, ANSYS. There is no direct procedure in the ANSYS program for solving diffusion problems, but the transient heat transfer modeling procedure can be modified by suitable variable substitution to solve diffusion problems. The heat transfer program has well documented and verified transient heat transfer modeling capabilities. Various benchmarks have established the accuracy of the numerical solutions, and, hence, its application to solve hydrogen diffusion problems is expected to yield definable results. The modification of the heat transfer equation for solving diffusion problems is explained below.

The governing equation for heat transfer analysis in 2-dimensions is:

$$\rho.C_{p} \frac{\partial T}{\partial t} = \frac{\partial}{\partial x} K_{x} \frac{\partial T}{\partial x} + \frac{\partial}{\partial y} K_{y} \frac{\partial T}{\partial y}$$
(11)

Where, K_x , K_y = thermal conductivity in x and y directions. C_p = specific heat, ρ = density, t = time, T = temperature. Substituting C_p = 1 and ρ = 1, equation 1 can be written as follows

$$\frac{\partial T}{\partial t} = \frac{\partial}{\partial x} K_x \frac{\partial T}{\partial x} + \frac{\partial}{\partial y} K_y \frac{\partial T}{\partial y}$$
(12)

If K = Diffusivity, and T = Concentration of a species, then equation 2 becomes 2-D diffusion equation, i.e., Fick's second law:

$$\frac{\partial C}{\partial t} = \frac{\partial}{\partial x} D_x \frac{\partial C}{\partial x} + \frac{\partial}{\partial y} D_y \frac{\partial C}{\partial y}$$
(13)

 K_x and K_y are equal for an isotropic medium. With this modification, diffusion problems can be solved by using the transient heat transfer modeling capabilities. In essence, diffusion problems are treated as transient heat transfer problems. The finite element formulation for solving a generic heat transfer equation is described in the ANSYS theoretical manual¹²².

Assumptions in the Numerical Model

Simplifying assumptions are necessary to obtain numerical solutions in most cases. These arise from the limited computer configuration, unavailability of material properties (D_H) and the lack of understanding of the phenomena. In this study, a 2-dimensional diffusion model was considered, as diffusion in the third dimension, i.e., the length direction, is negligible as the concentration gradient in the length direction is almost zero. In an actual weld hydrogen measurement specimen, Figure 1, the specimen faces on the specimen ends (lengthwise) may experience some diffusion as they are open to the atmosphere and will have a zero concentration on this surface. But, this hydrogen loss will be negligible compared to the diffusion taking across the thickness direction. In general, research efforts in numerical modeling of hydrogen diffusion have exclusively been two-dimensional.

The D_H for steels is known to be a function of material composition, but is generally unknown for specific compositions. This is particularly a problem where, as is the general case, weld metal and base metals have significantly different compositions. The Welding Institute (TWI) has compiled D_H data for different grades of steels (base metals). These D_H values along with Johnson and Hill's D_H data on pure iron were used in this analysis even though their use to describe localized hydrogen concentration will yield inaccurate results. But at a minimum, these D_H values can be used in the numerical models to establish outgassing relations.

Different D_H values used were assumed to be independent of hydrogen concentration and spatial location. A single D_H value was used for the entire weld specimen including the weld metal and the HAZ. Hydrogen diffusion was assumed to be independent of stress distribution and hydrogen trapping effect was neglected. These assumptions control the localized hydrogen concentration, but do not affect hydrogen outgassing behavior.

The experimental procedure adopted for $\frac{H_r}{H_o}$ determinations only measures the hydrogen that is mobile at RT. The hydrogen trapped at various defects will only affect the amount of hydrogen that is mobile, but will not affect the hydrogen diffusion process. Hence, the trapped hydrogen at defects does not exert any influence on the hydrogen diffusion process and this effect can be neglected in the numerical study without affecting the weld metal D_H calculations. In a similar manner, it can be assumed that the residual stress state does not affect the calculated apparent D_H.

The hydrogen concentration at the specimen surface was maintained at zero. This is a valid assumption as hydrogen escapes into the atmosphere (with a hydrogen partial pressure of about zero) as soon as it reaches the surface. Initial trials revealed that the FE mesh discretization was sufficiently fine to neglect the effect of mesh size on final results.

Initial Condition

An initial hydrogen concentration (temperature) of unity was prescribed in the weld metal in the present study. This was achieved by assigning nodal temperatures (concentrations) of unity for all the nodes inside the weld metal for a time period of 1 second. The prescribed initial hydrogen concentrations were inactivated after 1 second. The transient diffusion analysis was then initiated. Initial concentration in the base metal was assumed to be .0005.

Boundary Conditions

For the entire duration of hydrogen diffusion analysis, the concentration of hydrogen at the specimen boundary was maintained at 0.0 except for the plane of symmetry. The symmetry boundary condition was specified by maintaining the symmetry surface in adiabatic condition, i.e., no nodal temperatures and/or heat flows were prescribed on this surface. The boundary conditions were applied after 1.5 seconds time lapse.

SOLUTION PROCEDURE

After specifying the initial and boundary conditions, transient diffusion analysis was performed using a finite number of times steps (load steps). D_H values from different sources, Table 35, were input for analyzing hydrogen outgassing. A convergence criterion of 0.00001 was applied. Automatic optimized convergence criterion was used to control the number of iterations in a given time step while maintaining the desired accuracy. If the solution converged within a given number of iterations in a time step, then the next time step was input, and if it did not converge, then the run was terminated. The problem files were input and the solution was analyzed using the post-processor. The H_r in the specimen was determined as explained below.

RESIDUAL HYDROGEN CALCULATION

A program was written to calculate the instantaneous H_r content at different times for a given thermal cycle. Since nodal concentration is an average of surrounding elemental concentration, the summation of nodal concentrations (temperatures), for the entire specimen at different times during the outgassing cycle, provided the instantaneous H_r content in the specimen. The H_o was calculated by the summation of nodal concentrations in the weld metal only, after the application of boundary conditions, as this appropriately simulates the real welding situation. In real welding situation, no matter how short the time before quenching is, some hydrogen escapes

into the atmosphere from the weld surface. Therefore, for all numerical purposes, the H_o corresponds to the hydrogen remaining after the application of boundary conditions. From the numerically calculated H_o and H_r concentrations, instantaneous residual hydrogen fraction, $\frac{H_r}{H_o}$, was calculated for different isothermal outgassing conditions. Corresponding thermal factors were calculated using Coe and Chano's²² method, as described earlier.

THE NUMERICAL MODEL

The AWS Specimen

For the heat inputs used in the experimental part, the weld metal profile and the FEA mesh for the AWS specimen are shown in Figure 45(a),(b), respectively. Only half specimen geometry was used due to the symmetry to reduce the problem size. The symmetry boundary condition in ANSYS program is specified by making the boundary adiabatic, i.e. no nodal concentrations (temperatures) and heat flows are prescribed on this boundary. The initial and boundary conditions are also shown in Figure 45(c), (d).

Three isothermal outgassing treatments : 21 days at RT, 72 hours at 45°C and 6 hours at 150° C were used to determine the residual hydrogen. These outgassing treatments were simulated by specifying D_H values from TWI and Johnson and Hill corresponding to these temperatures, Table 35 and Table 36. The AWS A4.3-86 recommends the later two outgassing treatments only, but, all three outgassing treatments were considered, to study hydrogen outgassing at different temperatures in the AWS specimens.

The Lehigh Weldability Specimen

Initially, H_r determinations were performed using a hypothetical simple weld geometry derived from a submerged arc weld, Figure 46(a). The FE mesh is shown in Figure 46(b). Using this weld configuration and the mesh, the effect of different isothermal outgassing treatments on

 H_r content was analyzed. Three isothermal outgassing treatments, similar to those considered in the AWS specimen analysis were considered. In addition, different hypothetical weld thermal cycles were used to study if the hydrogen outgassing behavior in this case is similar to that from the isothermal outgassing treatments.

For the heat inputs used in the experimental part, the weld metal profile for GMAW weld and the FEA mesh for the Lehigh weldability tests specimen are shown in Figure 47(a), (b), respectively. To reduce the problem size, only half specimen geometry was used due to the symmetry. In addition, only a significant part of the specimen was used to reduce the problem size. The portion of the specimen that contained the weld metal, the HAZ and the base metal region that exceeded 100° C during welding was modeled. This was defined as the significant part. This covers regions 25 mm away from the weld centerline. The initial and boundary conditions are also shown in Figure 47(c), (d).

The Tekken Weldability Specimen

For the heat inputs used in the experimental part, the Tekken weld profile using GMAW and the FEA mesh are shown in Figure 48(a), (b), respectively. Because of the unsymmetrical nature of the Tekken specimen, both sides of the groove were used. In addition, only a significant part of the specimen was used to reduce the problem size. This covers regions 25 mm away on both sides of the weld centerline. The initial and boundary conditions are also shown in Figure 48(c), (d).

X-groove, V-groove and Bead on Plate Specimens

To study the hydrogen outgassing behavior from different groove configurations, additional numerical determinations were carried out using hypothetical X, V groove and BOP specimens. The weld profile was arbitrarily selected in X and V groove specimens, but was similar to the hypothetical Lehigh specimen. The BOP specimen weld profile was similar to that of the AWS specimen. The BOP specimen was 25 mm thick, instead of 12.5 mm thickness of the AWS

specimen. The weld profiles are shown in Figure 49(a),(b) and (c).

The objective of this study was to compare hydrogen outgassing dependence on the groove shape. The outgassing behavior of these specimens was then compared with Tekken and Lehigh weldability test specimens. Only half specimen geometry was used due to the symmetry to reduce the problem size. Similar to the Lehigh and Tekken specimen measurements, only near weld regions were used to reduce the problem size. The initial and boundary conditions were similar to that described for Lehigh, Tekken and AWS specimens.

Plate Width Effect Analysis

The plate width effect on H_r content was studied through a combined experimental and numerical analysis. The experimental part consisted of welding a standard Lehigh test specimen of 300 mm length x 200 mm width (12"x8") and a wider Lehigh specimen, 300 mm length x 400 mm width (12"x16"), Figure 50. The wider plate simulated a practical joint in terms of the plate dimensions. This plate is significantly smaller than most of the practical joints, but it was felt that this size plate still represents an "infinite" plate and, therefore, will simulate practical joints. Both plates were submerged arc welded at RT with a heat input of 1.4 kJ/mm using a robot to obtain reproducible welds. Weld thermal profiles were obtained by harpooning a Pt-Pt13%Rh thermocouple in the weld pool. Weld cooling times from 1500°C to 100°C, 45°C and RT were measured. The weld thermal cycles are shown in Figure 51.

In the numerical analysis part, these measured thermal cycles were input and H_r calculations were performed according to the procedure described earlier. The weld profile used approximately simulated the actual weld profile. Since the exact D_H was unknown, D_H values from TWI and Johnson-Hill were used. The H_r values for the standard and wider plates at 100, 45°C and RT were compared.

3.3 - RESULTS AND DISCUSSION

EXPERIMENTAL Hr DETERMINATION ON AWS SPECIMEN

The experimentally determined $\frac{H_r}{H_o}$ values at RT for different times are in Table 37 and are plotted in Figure 52. It is observed from Figure 52 that, $\frac{H_r}{H_o}$ values for all the four A36-AWS specimens exhibit negligible scatter. Therefore, these experimental results can be used to compare numerically determined results. It is also observed from Figure 52 that, complete hydrogen evolution occurs within 12 days and the amount of hydrogen evolution between 12 and 21 days is negligible.

NUMERICAL RELATION FOR HYDROGEN OUTGASSING IN AWS SPECIMENS

The $\frac{H_r}{H_o}$ variation as a function of the thermal factor, D Δt for different isothermal outgass-

ing treatments is shown in Figure 53 for the AWS specimen. It is observed that hydrogen outgassing is independent of diffusivity/outgassing temperature for a given specimen. For different outgassing treatments (diffusivities), the hydrogen outgassing curves exhibit same behavior, suggesting that hydrogen outgassing is controlled by the specimen geometry only. This suggests that H_r, after any isothermal outgassing treatment for any steel composition, can be calculated using the numerically determined hydrogen outgassing relationship, if the weld profile is kept constant.

Using a curve fitting program, hydrogen outgassing relations, i.e., $\frac{H_r}{H_o}$ variation with D Δt , for the AWS specimen were obtained. The polynomial relation for the AWS specimens is :

 $\frac{H_{r}}{H_{o}} = 0.9 - 18 \text{ } \text{D}\Delta t + 302 \text{ } \text{D}\Delta t^{2} - 3089 \text{ } \text{D}\Delta t^{3} + 18191 \text{ } \text{D}\Delta t^{4} - 59789 \text{ } \text{D}\Delta t^{5} + 101684 \text{ } \text{D}\Delta t^{6} - 69585 \text{ } \text{D}\Delta t^{7}$ (14) $\frac{H_{o}}{H_{o}}$

The regression coefficient for the above expression is 0.999. The exponential relation for the
AWS specimen is as follows:

$$\frac{H_{\rm r}}{H_{\rm o}} = .822 \exp(-8.02 \,\mathrm{D}\Delta t) \tag{15}$$

The regression coefficient for the exponential expression is 0.98. These equations will be slightly different for different bead surface areas, and hence can not be completely generalized. But, using a similar approach, hydrogen outgassing relations for different bead profiles can be obtained. From these relations, D Δt corresponding to a given $\frac{H_r}{H_o}$, or vice-versa can be obtained.

EXPERIMENTALLY AND NUMERICALLY DETERMINED $\frac{H_r}{H_o}$ COMPARISON

The numerically determined $\frac{H_r}{H_o}$ values were compared with the experimentally determined

values for the AWS specimens in the following manner. The numerically determined $\frac{H_r}{H_o}$ values are plotted AGAINST TIME in Figure 54, along with the experimentally determined values. The numerical values determined using Johnson and Hill and TWI maximum diffusivity are presented in Figure 54. From Figure 54, it is observed that the numerically determined hydrogen outgassing behavior closely follows the experimentally determined behavior. The difference in the numerical and experimental curves is because of the D_H differences. It is also observed that the experimental hydrogen outgassing behavior lies halfway between the numerically determined behavior suggesting that the D_H of the A36 AWS specimens falls between the TWI (maximum D_H) and Johnson and Hill's D_H values.

From Figure 54, the effect of D_H on hydrogen outgassing is clear. For example, a steel with D_H comparable to Johnson and Hill's value needs 21 days for 95% outgassing, whereas, a steel with D_H similar to TWI (maximum D_H) needs only 3 days for 100% hydrogen outgassing. Since D_H is steel composition and thermo-mechanical history dependent, complete outgassing is not

always possible using the hydrogen outgassing treatments recommended by the weld hydrogen measurement standards. For example, steels with D_H similar to TWI minimum will only experience 15% outgassing, i.e. $H_r = 85\%$, after 21 days at RT outgassing. Using similar analysis, the lower limits on D_H for 100% outgassing can be established for AWS or other standard weld hydrogen measurement specimen configurations.

NUMERICAL AND EXPERIMENTAL APPROACH TO WELD D_H MEASUREMENT

From the comparison between numerical and experimental $\frac{H_r}{H_o}$ data for the AWS specimen, D_H calculations for the entire A36 AWS specimens can be performed. Since the weld profiles in numerical and experimental analysis are almost same, the experimental $\frac{H_r}{H_o}$ data should fall on the numerically determined hydrogen outgassing relation shown in Figure 53. From this information, an apparent diffusion coefficient for the A36 composite weld specimen can be calculated as described below. It should be noted that this diffusivity value will be the combined value for composite weld metal plus HAZ and the base metal.

By taking log of both sides of the numerically determined exponential type outgassing relation for the AWS specimen determined earlier, eq. 6, we have :

$$\log \frac{H_r}{H_o} \frac{1}{8.22} = -8.02 \text{ D}\Delta t \tag{16}$$

Hydrogen diffusivity at RT then can be calculated easily by substituting the experimentally determined Δt and $\frac{H_r}{H_o}$ values for different outgassing times in the above equation. Since the regression coefficient is only 0.98, diffusivity estimates can not be obtained using the exponential equation. The polynomial expression can be used to get better estimates of the D value as the regression coefficient is 0.999, but, the seventh order polynomial makes it complicated to solve for D. Hydrogen diffusion coefficient estimation was, hence, performed graphically in this study. Using the numerical hydrogen outgassing curve for the AWS specimen, Figure 53, the D Δ t value corresponding to the experimentally determined $\frac{H_r}{H_o}$ was noted for different known time intervals, Δt . The Δt corresponding to the experimental $\frac{H_r}{H_o}$ was substituted and the apparent hydrogen diffusivity, D_{app} , value was calculated for each time interval. An average of all these apparent diffusivity values for the composite A36 specimen, estimated graphically, is 7.46 E-07.

To verify the estimated diffusivity value, additional numerical calculations were performed. The estimated D_{app} was used to calculate the experimental $D_{app}\Delta t$ by using the experimentally determined $\frac{H_r}{H_o}$ hydrogen fraction and Δt . The experimental $\frac{H_r}{H_o}$ and $D_{app}\Delta t$ for A36 specimen were plotted with the numerical data for the AWS specimen, Figure 55. The experimental data fall exactly on the numerically determined hydrogen outgassing curve for the AWS specimen for different $\frac{H_r}{H_o}$ and Δt , indicating that the estimated D_{app} is correct. If the determined apparent diffusivity value is incorrect, then the experimentally observed outgassing will not obey the numerically determined AWS specimen outgassing behavior.

Results in Figure 55 also suggests that the numerical procedure accurately describes the experimental outgassing behavior of the AWS specimen provided that the weld profile remains unchanged in both the cases. The procedure described above can be repeated to estimate apparent D_H for any given weld material. Diffusivity at 45°C can be similarly calculated. Calculation of diffusivity at temperatures higher than 150°C will pose some problems due to the cooling time required from higher temperature to RT during experimental hydrogen measurement. This cooling time influence on the thermal factor is difficult to determine, and hence, at temperatures above 150°C, the hydrogen diffusivity determinations will contain some inaccuracies. In fact, using a single weld specimen hydrogen diffusivity as a function of temperature can be calcu-

lated using the technique mentioned above. This procedure will involve welding the specimens and then monitoring hydrogen outgassing at different temperatures. For example, initially, hydrogen outgassing should be carried out at RT for a short time period and evolved hydrogen measurements should be carried out at the end of this outgassing treatment. This procedure should be continued with outgassing at successively higher temperatures followed by measurement of evolved hydrogen. After complete outgassing, the initial hydrogen, H_o, can be calculated by adding the total evolved hydrogen. $\frac{H_r}{H_o}$ corresponding to each outgassing relation for this specimen corresponding D Δt can be determined. From the knowledge of $\frac{H_r}{H_o}$ and Δt corresponding to each outgassing temperature, the D_H value at these temperatures can be determined.

The above D_H calculations suggest that the use of Johnson and Hill's D_H data for determining H_r content can lead to inaccurate results. In this regard, Dickehut and Ruge's¹⁰³ analysis is noteworthy. They applied FDM using explicit representation to calculate hydrogen distribution in a weld bead produced by SAW tandem technique. They used hydrogen diffusion coefficients from Sykes¹⁰⁴ and Johnson and Hill¹⁰⁵ studies. The numerically calculated H_r values were compared with experimental results on sectioned specimens and were found to be higher than the experimentally measured value. These authors attributed this difference to the hydrogen loss during specimen preparation. But, this difference could very well be due to the wrong assumption of D_H values. If they had used correct D_H values for these steels, which are higher than Sykes and Johnson and Hill's values, then a closer match between numerical and experimental results would be observed.

HYDROGEN OUTGASSING IN WELDABILITY SPECIMENS

The numerically derived hydrogen outgassing relation for the Lehigh specimens derived from SAW weld profile are shown in Figure 56. It is observed from Figure 56 that isothermal as well as continuous cooling type weld thermal cycles follow same outgassing behavior. This suggests that the $\frac{H_r}{H_o}$ vs. D Δt relations determined from isothermal outgassing treatment can be extended to weld cooling thermal cycles. All the subsequent numerical hydrogen outgassing relations were then determined using only RT outgassing.

The numerically determined hydrogen outgassing behavior for GMAW Lehigh specimen is shown in Figure 57. A curve fitting program was used to obtain the hydrogen outgassing relation and is as follows.

$$\frac{H_{\rm r}}{H_{\rm o}} = 0.79 {\rm e}^{(-7.67 {\rm D}\,\Delta\,t)} \tag{17}$$

The numerically determined hydrogen outgassing behavior for GMAW Tekken specimen is shown in Figure 58. A curve fitting program was used to obtain the hydrogen outgassing relation and is as follows.

$$\frac{H_{\rm r}}{H_{\rm o}} = 0.81 {\rm e}^{(-6.75{\rm D}\,\Delta\,{\rm I})} \tag{18}$$

The regression coefficient for both the specimens is 0.98. These equations are applicable to the weld configurations used in this study only, and hence can not be generalized. These equations suggest that hydrogen outgassing behavior in these two specimens is different.

Comparison between Outgassing in Lehigh and Tekken Specimens

The numerically determined hydrogen outgassing behavior of GMAW Lehigh and Tekken specimens was compared, Figure 59. The Lehigh specimen exhibits higher hydrogen outgassing compared to the Tekken specimen for a given thermal factor. Therefore, for an equivalent thermal factor, the Tekken specimen H_r content will be higher than that in the Lehigh specimens.

This effect is due to the groove shape difference. The bottom inverse-U shaped slot in the Lehigh specimen promotes hydrogen outgassing into the atmosphere as soon as the diffused hydrogen from the weld metal reaches the bottom slot. In comparison, in Tekken specimens hydrogen outgassing mainly occurs from the top bead surface and the weld root's contribution to hydrogen outgassing is small. Therefore, hydrogen distribution will also be different. The Tekken specimen will exhibit higher H_r concentration in the bottom region of the weld as compared to the Lehigh specimen.

After $\frac{H_r}{H_o}$ values equal to 0.20, the maximum hydrogen concentration region shifts inside the base metal and hydrogen outgassing is at its lowest. This causes almost equal hydrogen outgassing behavior in the later stages in these two specimens.

From the above discussion, the effect of groove shape on hydrogen outgassing is quite clear. This suggests that comparison between results from two different weldability tests is not valid since the H_r content and also the hydrogen distribution will be different in these specimens. For example, Controlled Thermal Severity (CTS) weldability test results were used by TWI and BS to propose a preheat prediction scheme for avoiding HAC in HAZ. The predictions from this scheme are, hence, strictly applicable to joints with similar weld configuration as the CTS test. The different hydrogen distribution and content will augment the effect of joint dependent residual stresses on HAC.

GROOVE SHAPE EFFECT ON HYDROGEN OUTGASSING BEHAVIOR

The hydrogen outgassing behavior for the root pass of V-groove and X-groove and BOP specimens was numerically determined and is shown in Figure 60 along with the results on Tekken and Lehigh weldability specimens. The numerical hydrogen outgassing equations for different groove specimens are listed below.

$$\frac{H_r}{H_o} = 0.79e^{(-13.31D \wedge t)} \text{ for V-groove specimen,}$$
(19)

$$\frac{H_r}{H_o} = 0.77e^{(-9.67D \Delta t)} \quad \text{for X-groove specimen,}$$
(20)

$$\frac{H_r}{H_o} = 0.82e^{(-5.46D \ \Delta t)} \quad \text{for BOP specimen,}$$
(21)

The regression coefficient for these expression is about 0.98. From Figure 60, the groove shape effect on hydrogen outgassing behavior is clear. Hydrogen outgassing decreases as groove shape becomes less and less open, i.e. in the order of V, X, Lehigh, Tekken and BOP. The V-groove shape promotes hydrogen outgassing, since hydrogen diffusion from the bottom plate surface occurs right in the early stages of hydrogen diffusion. Similar process allows more outgassing from the X-groove specimen. The BOP specimen is equivalent to the last pass in a multi-pass joint and this suggests that the last pass retains most of the initial hydrogen.

The above mentioned dependence of H_r content on weld/joint geometry has important implications. Weldability tests are used to assess HAC susceptibility of practical joints. Since outgassing is dependent on the weld bead profile and the groove geometry, the results from weldability tests can not be directly applied to practical welding joints with different joint geometries, e.g., V-groove, X-groove and BOP specimen configurations. This is because the H_r content and hydrogen distribution will be different for different joint geometries for a given thermal cycle. This prevents direct correlation between the weldability test results and actual welding joints. Since, the Lehigh and Tekken groove geometry is less open to the atmosphere and Vgroove, X-groove root weld geometry more open, the latter will have less H_r content for same weld thermal cycles. Therefore, the welding conditions/preheating temperatures predicted from the Lehigh/Tekken weldability test results will be too conservative. Alternatively, some other groove configurations will allow less hydrogen outgassing compared to these weldability tests, and in this case, the welding conditions/preheating temperatures from these weldability tests may not avoid HAC occurrence.

The above discussion also implies that most of the weldability tests are improperly designed for assessment of HAC susceptibility of practical joints. In fact, suitable weldability tests should be designed which will exhibit similar hydrogen outgassing behavior to that of the practical weld joint being considered. This can be achieved by keeping the weldability test groove configuration similar to the weld joint. Using a similar approach, weld joints promoting hydrogen outgassing can be designed so that H_r content can be reduced. This type of application was also proposed by Sleptsov, et al¹¹⁰.

With the help of equations (8) and (9), H_R at 100°C, 45°C, R.T. or at the crack initiation temperature can be calculated for any given thermal cycle, if D_H and the crack initiation temperature is known. Hence, preheat temperatures/thermal cycles can be specified to lower the H_R to a harmless level. It is proposed that these equations could assist assist in development of a new preheat prediction scheme incorporating H_r rather than the H_o . This type of preheat prediction scheme will be more fundamental as H_R correlates with the cracking tendency more closely than the H_o^{28} .

PLATE WIDTH EFFECT

The plate width did not significantly affect $H_{r_{100^{\circ}C}}$, Table 38. This was true for different D_{H} values used in this study. The plate width effect became significant when H_{r} at 45°C and at 25°C was analyzed and is strongly D_{H} dependent. For steels with D_{H} similar to that from Johnson and Hill's data, the plate width effect was insignificant. Whereas, for steels with D_{H} similar to TWI maximum D_{H} , the H_{r} content at 45 and 25°C was lower for standard Lehigh specimens compared to the wider Lehigh specimens. This suggests that the plate width effect is D_{H} , and, hence, composition dependent.

The above result is important for successful application of preheating temperature from weldability tests to practical joints involving larger plates. If HAC is not observed in a weldability test, e.g. Tekken and Lehigh test, with the use of certain welding conditions, then application of these welding conditions to practical weld joints may not always avoid HAC in these joints. In such cases, the larger plate dimensions of a practical joint will promote more H_r retention and this will enhance HAC susceptibility of the joint.

This plate width effect on H_r at 100°C will not be observed in preheated specimens if 1500 to 100°C is similar in both cases. But, if weld cooling rates to RT are considered, then practical

joint with localized preheating will cool at a faster rate than an uniformly preheated weldability test specimens and will have a higher H_r at RT. This effect has been neglected since most researchers^{11, 28} assumed that hydrogen diffusion below 100°C is almost negligible based on the Johnson and Hill's D_H data. But, according to recent studies⁸⁴, the D_H values in steels are much higher than those proposed by Johnson and Hill's study, and in these steels, the H_r content at 45° C/RT/cracking is a better measure of HAC susceptibility. The plate width effect will be more pronounced when applying Tekken weldability test results to practical welding conditions because Tekken test dimensions are much smaller compared to the Lehigh specimen. Therefore, the plate width effect on H_r content at 45° C and RT will be substantial. On many occasions, weldability tests are selected based on the material availability without any consideration to the differences in cooling rates. The plate width effect results suggest that this practice is wrong. In fact, the weldability tests should be designed such that weld cooling rate in practical joints are closely simulated in these tests.

3.4 - CONCLUSIONS

- The modified numerical procedure accurately simulates hydrogen outgassing behavior of the weld specimens as verified by the experimental measurements.
- 2) A simple numerical and experimental method was proposed for average weld hydrogen diffusivity determinations using AWS specimen geometries. Using this method, mean weld hydrogen diffusivity as a function of temperature can be measured with a single specimen.
- 3) From the numerical results on AWS specimens, it can be concluded that complete outgassing is not always possible using the hydrogen outgassing treatments recommended by the weld hydrogen measurement standards. Complete outgassing is a strong function of hydrogen diffusivity, which is steel composition and thermo-mechanical history dependent. Therefore, calculations should be performed to ensure complete outgassing for steels with lower hydrogen diffusivity.
- 4) The groove shape significantly affects hydrogen outgassing behavior. It was observed that hydrogen outgassing in Lehigh specimens was more pronounced compared to the Tekken specimens.
- 5) Hydrogen outgassing behavior in Lehigh and Tekken weldability tests differed significantly from the root pass welds of commonly used practical weld joints. Weldability specimens exhibited less pronounced hydrogen outgassing behavior compared to the practical welding joints suggesting that the weldability test results are conservative.
- 6) The weldability test welds cool at a slower rate, due to their smaller size, thus allowing more hydrogen outgassing compared to the practical weld joints which cool at a faster rate. In this respect, the H_r content, and therefore, HAC susceptibility of practical joints is not accurately assessed by weldability tests.







Figure 46. (a) Lehigh (SAW) Test Weld Profile, (b) FE Mesh, (c) Initial Condition,(d) Boundary Condition



Figure 47. (a) Lehigh (GMAW) Test Weld Profile, (b) FE Mesh, (c) Initial Condition,(d) Boundary Condition



Figure 48. (a) Tekken (GMAW) Test Weld Profile, (b) FE Mesh, (c) Initial Condition,(d) Boundary Condition





Figure 49. Weld Profile, (a) X-groove Specimen, (b) V-groove Specimen, (c) BOP Specimen









Figure 51 Experimentally Measured Thermal Cycles for Standard and Wider Lehigh Test Specimens



Figure 52. Experimental Hydrogen Outgassing Behaviour for AWS Specimen at RT







Figure 54. Comparison between Numerical and Experimental Hydrogen Outgassing Behavior for AWS Specimen at RT



Figure 55. Comparison between A36-Experimental and Numerical Hydrogen Outgassing Behavior for AWS Specimens



Figure 56. Numerically Determined Hydrogen Outgassing Behavior

-Lehigh Specimen Welded Using SAW

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Figure 57. Numerically Determined Hydrogen Outgassing Behavior -Lehigh Specimen Welded Using GMAW







Figure 59. Comparison between Numerically Determined Outgassing Behavior of Weldability Tests Welded Using GMAW



Figure 60. Comparison between Numerically Determined Outgassing Behavior of Different Specimen Configurations

Table 35 : Hydrogen Diffusion Coefficients at 25°C, 45°C, 150°C

Temperature	The V	Velding Institute	Johnson & Hill's data	
	Maximum	Mean	Minimum	
	cm ² /sec	cm ² /sec	cm ² /sec	cm ² /sec
25°C	2.591 e-06	1.194 e-06	3.191 e-09	2.191 e-07
45°C	4.989 e-06	2.179 e-06	1.792 e-08	5.031 e-07
150°C	3.103 e-05	2.016 e-05	1.061 e-05	1.088 e-05

** Johnson & Hill's data was extrapolated to lower temperatures to obtain diffusivities at 25° C and 45° C

Table 36 : TW	l, Sykes and	Johnson	Hill's	Hydrogen	Diffusivity	Data
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Temperature Range	TWI Maximum Diffusivity, cm ² /sec
$\theta \ge 500^{\circ}C$	$D = 1.06X10^{-2} \exp(-13136/RT)$
$500^{\circ}C \ge \theta \ge 90^{\circ}C$	$D = 1.4X10^{-3} \exp(-3200/RT)$
$90^{\circ}C \ge \theta > 20^{\circ}C$	$D = 8.66X10^{-3} \exp(-6165/RT)$
	TWI Minimum Diffusivity, cm ² /sec
$\theta \ge 300^{\circ} C$	$D = 1.06X10^{-2} \exp(-13136/RT)$
$300^{\circ}C \ge \theta \ge 150^{\circ}C$	$D = 9.8X10^{-4} \exp(-3847/RT)$
$150^{\circ}C \ge \theta > 20^{\circ}C$	$D = 2637 \exp(-16240/RT)$
	TWI Mean Diffusivity, cm ² /sec
$\theta \ge 400^{\circ} \text{C}$	$D = 1.06X10^{-2} \exp(-13136/RT)$
$400^{\circ}C \ge \theta \ge 150^{\circ}C$	$D = 1.5 X 10^{-3} \exp(-3629/RT)$
$150^{\circ}C \ge \theta > 20^{\circ}C$	$D = 1.7X10^{-2} \exp(-16240/RT)$
	Diffusivity, cm ² /sec - From Sykes et al.
$\theta \ge 500^{\circ} C$	$D = 1.51X10^{-2} \exp(-11970/RT)$
	Diffusivity, cm ² /sec - From Johnson and Hill.
$500^{\circ}C \ge \theta > 200^{\circ}C$	$D = 1.4X10^{-3} \exp(-3200/RT)$
$200^{\circ}C \ge \theta \ge 20^{\circ}C$	$D = 1.2X10^{-1} \exp(-7820/RT)$

** Johnson & Hill's equation was extrapolated to lower temperatures even though the equation is valid for temperatures $>100^{\circ}C$

Time	Specimen 1		Specimen 2		Specimen 3		Specimen 4		Average
Seconds	H _R	$\frac{H_R}{H_o}$	$\frac{H_R}{H_o}$						
72000	.68	.49	.63	.48	.69	.51	.67	.49	.49
234000	.30	.22	.29	.22	.33	.24	.31	.23	.23
320400	.21	.15	.20	.15	.23	.17	.22	.16	.16
410400	.15	.11	.15	.11	.17	.12	.16	.1	.11
496800	.11	.08	.11	.08	.12	.09	.12	.08	.08
586800	.08	.06	.09	.07	.1	.07	.1	.07	.07
684000	.07	.05	.08	.06	.08	.06	.08	.06	.06
856800	.06	.04	.06	.05	.07	.05	.06	.05	.05
961200	.05	.03	.055	.04	.06	.04	.05	.04	.04
1717200	.04	.028	.05	.03	.05	.03	.04	.03	.03

Table 37 : Experimentally Determined $\frac{H_r}{H_o}$ for A36 AWS Specimens - RT Evolution

* $H_o = 1.394, 1.328, 1.361, 1.371$ for specimen nos. 1,2,3,4 respectively and was determined after evolution at 25,45 and 150°C

Table 38 : Plate Width Effect on $\frac{H_r}{H_o}$ at Different Temperatures

Diffusivity	Sta 30	ndard Leh 00x200 mi	igh n	Wide Lehigh 300x400 mm			
Diritasivity	100°C	45°C	25°C	100°C	45°C	25°C	
Johnson & Hill	0.99	.986	0.977	0.99	0.99	0.99	
TWI (Mean)	0.99	0.93	0.90	0.99	0.99	0.99	
TWI (Maximum)	0.98	0.87	0.80	0.99	0.98	0.97	

CHAPTER 4

NUMERICAL SIMULATION OF THERMAL AND STRESS ANALYSIS

4.1 - BACKGROUND

INTRODUCTION

Initiation condition for hydrogen assisted cracking in a weldment can be described accurately by local stress and strain history along with hydrogen content and microstructures. Finite element analysis allows one to evaluate transient as well as residual local stress and strain development during welding as a function of different variables. In general, experimental measurements of stress distribution in weldments do not provide information in the areas of interest, such as, stress concentrations, interfaces in weld zones or other embedded locations. With the help of FEA, this difficulty can be overcome, and therefore, the influence of joint geometries, dimensions and configurations can be studied. Therefore, differences between weldability tests and actual weldments can be studied in detail using FEA. In this study, important issues related to the heat transfer and residual stress development in weldability tests are analyzed and the differences with actual weldments are qualitatively discussed. In the following section, residual stress formation mechanism is discussed.

RESIDUAL STRESSES

Residual stresses are internal forces without the action of external forces. As constraint sources they are in equilibrium only with themselves. The stresses from external load can augment or decrease the total residual stresses. Residual stresses exist in domains ranging from atomic to macroscopic levels. From a practical engineering viewpoint, only global level residual stresses are more relevant¹²³.

Mechanism of Residual Stress Generation

Residual stresses are the result of inhomogeneous permanent deformation which is made up of ¹²³:

- Volumetric (or dilatoric) strain as a result of thermal expansion, chemical conversion, microstructural transformation or change in state;

- Distortional (or deviatoric) strain as a result of (time-independent) plastic, and also (timedependent) viscoplastic deformation.

Formation of residual stresses in a weldment can be attributed mainly to two sources :

 The non-uniform heating and cooling of the part during welding. This non-uniform heat flow process causes non-uniform expansion/shrinkage and produces thermally induced strains.

2) The volumetric changes during metallurgical phase transformations.

A distinction exists between the transient residual stresses during welding and the permanent welding residual stresses following complete temperature equalization¹²³. The former determine the weldability (hot cracking tendency); the latter are of interest with respect to the strength of the structure, HAC, stress corrosion cracking and fatigue¹²³.

Welding Residual Stress Fields

Both longitudinal and transverse residual stresses occur as a result of similar mechanism in weldments. Longitudinal stresses are generated due to the resistance to the longitudinal shrinkage of the weld. The tensile stresses are limited to a narrow area close to the weld, their maximum value being at or above the yield limit. Lower compressive stresses exist in the surrounding region, dropping off rapidly the further away they are from the weld.

Transverse stresses in the plane of the plate are generated due to the resistance to the transverse shrinkage of the weld scam, especially when the plate is restrained. They are not res-

tricted to a narrow area close to the weld, but also comprise the surrounding region. Weldtransverse stresses in the plate thickness direction may be generated if the plate thickness is sufficiently large and give rise to unsafe tri-axial tensile stress state.

It is clear from the mechanisms stated above, that the variation of concentrated heat in space and time is responsible for residual stresses and distortion in welding. The extreme heterogeneous and transient temperature fields, coupled with physical and microstructural changes in the material, determine the residual stresses. The temperature distribution influences residual stresses directly through the thermal strains and indirectly through the transformation strains which accompany the changes in state and microstructure.

To study the distribution and the magnitude of residual stresses and strains, it is necessary to analyze the thermal history. In the following section, various aspects of welding heat transfer are discussed. Analytical as well as numerical approaches for understanding the thermal behavior are then studied in detail. Because of its suitability for residual stress analysis, the numerical approach used in this research study is detailed.

HEAT TRANSFER FUNDAMENTALS

From the principle of conservation of energy for any homogeneous and isotropic continuum, the field equation for three dimensional (3-D) heat transfer with temperature independent material properties is :

$$\frac{\partial}{\partial x} K_x \frac{\partial T}{\partial x} + \frac{\partial}{\partial y} K_y \frac{\partial T}{\partial y} + \frac{\partial}{\partial z} K_z \frac{\partial T}{\partial z} + \partial \frac{Q_v}{\partial t} = \rho.C \frac{\partial T}{\partial t}$$
(22)

Where K_x , K_y , K_z = thermal conductivity in X, Y and Z directions, respectively. C = specific heat, ρ = density, t = time, T = temperature and Q_v is the heat energy consumed or released per unit volume.

Rosenthal¹²⁴ developed analytical solutions for a moving heat source problem assuming

quasi-steady state heat transfer for both 2-dimensional (2-D) and 3-dimensional (3-D) heat flow conditions. He also assumed a point heat source, no melting and negligible heat of fusion, constant thermal properties, no surface heat loss and infinite plate. The 2-D heat flow solution for thin plates was obtained assuming line heat source, and considering radiation and convection losses. Modified solutions based on the same theme have been proposed by Ryaklin¹²⁵ Okerblom¹²⁶, Naka & Masubuchi¹²⁷, Wells¹²⁸, Nippes¹²⁹, Christensen¹³⁰, Adams¹³¹, Myers¹³², etc. Recently, Kasuya, et al¹³³. have used analytical solutions to predict cooling rates in finite, infinite plates, and even in Electroslag welding. According to Myers¹³², the application limiting assumptions of most of the analytical methods are :

1) Idealization of a distributed heat source as a point or line source

- 2) Inaccurate representation of material properties
- 3) Quasi-steady state heat flow
- 4) Convection and radiation surface losses are not considered in most of the analytical models
- 5) Two-dimensional approach and infinite plate size assumption
- 6) No consideration of the Latent heat
- 7) Weld pool convection effects, hydrodynamics, surface tension effects are not considered

These analytical models provide limited accuracy because of gross simplification of the welding process and the results are accurate only in the regions away from the heat source. Nevertheless, these solutions simplistically provide a general idea about the thermal behavior during the welding process.

Kasuya and Yurioka¹³³ have extensively applied these analytical solutions with distributed heat sources to predict preheating conditions for avoiding HAC for different plate configurations and preheating methods. This type of application is still beyond the reaches of the numerical methods due to the computational costs, time and effort involved. Therefore, for a generalized idea about heat flow in welding, the analytical solutions are still relevant in an engineering sense.

Moore, et al¹³⁴. compared the solutions from point source models of different researchers with numerical and experimental results, and found that point source analytical solutions give a good estimate of 800–500°C cooling time, which determines the transformation products, when a thermal conductivity value of 25 W/m°C is used for both thick and thin plates. However, the 1500–100°C cooling times from the analytical solutions using point heat source, which determine hydrogen outgassing potential of a weld joint, did not agree well with the measured and numerically determined cooling rates. The authors attributed this to the use of only one temperature independent thermal conductivity value.

As discussed above, the analytical approach can not be used for accurate description of localized temperature fields during a welding operation. In addition, generalized temperature distributions obtained from an analytical approach are only good enough for calculating global displacements and distortion. In contrast, HAC initiation can be more realistically defined by local stress and strains, hydrogen concentration and microstructure. All these variables are dependent on localized temperature distribution which can be determined accurately by using numerical approaches, such as, FEM and FDM. Localized transient temperature distribution is the starting point for welding residual stress calculation using numerical method. Finite element analysis is now used extensively to study the development of residual stresses during welding. The numerical approach and various issues relevant to FEA are discussed in the following sections.

NUMERICAL APPROACH

Numerical modeling of thermal response of a welding process is a complex phenomenon because of the following considerations¹²³:

- Electrode melting and fusion of the base metal which is dependent on the metal deposition rate, shape and size of the weld groove and the welding process itself

- Weld pool convection in the molten state

- Latent heat effects during melting and phase transformations

- Non-linear nature of the process and temperature dependent material properties

- Radiation heat transfer during welding and convective heat transfer during cool-down
- Energy density distribution in the arc and shape of the arc column

Efforts are being made elsewhere to understand the arc physics and fluid dynamics in the pool. At present, only a limited number of models exist that define each of these phenomena on an individual basis. Because of the lack of understanding of the underlying fundamental processes, most of these phenomena are simplified for ease of model implementation. Various simplifications commonly adopted in FEA of welding processes are discussed below.

MODEL SIMPLIFICATIONS

Finite element solutions, in principle, permit nearly any complexity to be taken into account although, in practical terms, economic constraints limit it¹²³. Common to essentially all models is the fact that the complex processes in the weld pool are ignored, in particular, the melting and solidification process with melting movement and heat transfer by means of convection and radiation¹²³. A relatively coarse simulation of the thermal process, which is matched to the resolution of the subsequent continuum-mechanical model, is sufficient for welding residual stresses and welding deformations caused directly by the temperature field through thermal expansion¹²³. Detailed mechanical property data at elevated temperatures play an insignificant role in on residual stress development. As temperatures rise to near melting point, these properties vanish and the material does not support any internal or external load. Detailed modeling is required, howeever, as cooling and transformation stresses at lower temperature play a significant role¹²³.

Spatial Simplifications of the Heat Source

The concentrated heat input is simplified for the heat transfer analysis. The arc energy density distribution influences temperature field at distances of equal magnitude to the dimensions of the heat source. Farther away, this influence decreases even if the distributed heat source is replaced by a concentrated source in the center of its area or its volume¹²³. In other words, the temperature near the source is determined by the energy density distribution, and farther away from the heat source, the geometric condition determines the temperature distribution¹²³.

Heat Source Simplification

The heat input flux distribution can be simulated by assuming point, line or distributed source. For correct simulation of temperature fields, distributed heat sources have been used. The energy distribution of choice is gaussian. Pavelic et al¹³⁵, formulated the radially symmetric circular gaussian distribution. A more general ellipsoidal heat distribution model was proposed by Goldak, et al¹³⁶ Their double ellipsoidal model provides more control and flexibility to the analyst to predict correct dimensions of the weld pool. Smartt, et al¹³⁷. have experimentally found that the heat flux distribution matches a circular Gaussian distribution.

Time Simplification of Heat Source.

The heat transfer from the arc is further simplified by specifying the duration for which the heat source is active. A momentarily active source is characterized by the heat input acting at that moment. The movement of the heat source depends on type of the welding speed, such as, stationary, moving and rapidly moving, and is assumed as linear and uniform¹²³. If a higher welding speed is employed, then the heat propagation in the heat source direction is negligible and this makes the problem amenable for 2-D analysis. For high heat input and lower welding speed processes, this assumption is not representative and accurate results require a full 3-D analysis.

Geometrical Considerations

The heat flow is 1,2 or 3-dimensional based on the component geometry, welding process parameters and the welding process itself. For a semi-infinite solid, the heat transfer is three dimensional. For thin plates, the heat transfer is two dimensional¹²³. Because of cost considerations, 3-D model is generally avoided and a 2-D cross sectional plane is assumed. The choice is

primarily based on the desired accuracy and the computational resources at hand.

Radaj¹²³ has mentioned that the simplification to 1-D or 2-D instead of 3-D is all the more justified, the larger the dimension of the component in the appropriate direction, the shorter the period of heat propagation under study, the greater the thermal diffusivity, the less distant from the heat source the area under study, and greater the coefficient of heat transfer. For accurate results, three dimensional analysis is generally required¹³⁸. With the advent of faster and cheaper computers this problem is being tackled easily.

Heat Losses

The arc energy input to the weld pool is primarily transferred to the base metal by conduction and convection within the weld pool¹³⁹. The heat input in welding is localized near the weld and the heat is dissipated to the other areas of the weldment through conduction in metals. Heat loss through thermal radiation is dominant near the weld pool due to the large temperature difference between the weldment and the environment¹³⁹. Far from the heat source, heat loss can be convection dominated if the workpiece and jigging are not so large that the temperature rise in the workpiece is negligible¹³⁶. In field welding, the air drafts can affect the convection losses, here the actual values of surface heat transfer coefficient, if available, should be used to simulate the correct response. Additionally, in SAW, where the weld is covered with an insulating slag layer, the heat loss due to convection and radiation from the molten weld pool is negligible. Convective heat losses do not affect the metallurgically important weld cooling rates significantly. For GMAW, GTAW and SMAW processes, radiation losses contribute significantly to the net heat losses when the arc is on. Tekriwal^{140, 141} has considered these losses while simulating GMAW with filler metal addition.

Other researchers have considered heat losses during welding by using the heat efficiency term. Various heat efficiency values for different welding processes are available in the literature.

Initial and Boundary Conditions

The heat transfer field equation is solved under the given initial and boundary conditions. Initial or preheating temperature of the component and interpass temperature during multiple pass welding is prescribed through the initial boundary condition. Convective heat losses occurring through the component surfaces, effect of air drafts during field welding and the effect of insulating slag layer can be incorporated by specifying the appropriate boundary conditions.

Three types of boundary conditions can be specified.

- Uniform or non-uniform temperatures can be prescribed. This is also known as an essential boundary condition
- The heat transfer through the surrounding medium can be specified through the heat transfer coefficients, the natural boundary condition
- Adiabatic boundary conditions, (also used to prescribe the symmetry surfaces in the ANSYS program), on insulated surfaces.

Spatial Discretization

To get a balance between computing cost and desired accuracy, the finite element mesh needs to be optimum. Finer mesh will help getting desired results but will be cost-prohibitive. Whereas, the use of a coarse mesh comes at the expense of accurate results. A carefully graded mesh is the solution but is not necessarily easier to obtain. Some guidelines must be met to obtain better results. The mesh should be sufficiently fine to model the heat source accurately. Therefore, weld metal and HAZ regions should be discretized with smaller elements than in regions far removed from the heat source. Furthermore, the choice of linear or quadratic elements will also affect the mesh coarseness. Pammer¹⁴² has determined the criterion for the element size near the hot boundary to achieve accuracy in the analysis. Tekriwal¹⁴¹ recommends that at least 5 to 6 nodes should be taken under the arc beam radius for steel materials to track peak temperatures accurately. Finer mesh should be used in conjunction with fine time

increments to get accurate results. Also, Thomas et al¹⁴³ have shown that every given mesh has an inherent limit of achievable accuracy, therefore continuous refinement of time increments will not continuously improve the accuracy.

Time Discretization

Goldak, et al¹³⁶. have suggested a use of 10-20 time steps for simulating the passage of ellipsoidal heat source to cross the reference plane in the cross-sectional model. Whereas, in the in plane and 3-D models the heat source can move approximately one half of a weld pool length in one time step. The effect of the interaction between the integration scheme, [θ ; explicit (θ =0.0), Crank-Nicholson (θ =0.5), Galerkin (θ =2/3) and implicit (θ =1.0)] and any prescribed thermal loads, fluxes or power densities is to convert these thermal loads to an equivalent nodal thermal load specified at the beginning of each time step and another at the end of each time step. The time integration scheme linearly interpolates these to the θ point¹³⁶.

In transient heat transfer analysis, faster convergence and reasonable accuracy can be achieved by an appropriate time-stepping scheme. Due to the sharp temperature rise in the initial heating period, short time increments and a relatively large temperature convergence tolerance should be used while longer time increments and smaller tolerances should be used during cool-ing¹³⁹.

NON-LINEAR MATERIAL PROPERTIES

The FE transient heat transfer analysis requires accurate values of thermal conductivity, specific heat, material density and latent heat of fusion and phase transformations up to the melting point¹³⁹ The biggest obstacle is the lack of high temperature data for most of the materials. Generally, the low temperature data is extrapolated to generate the high temperature data. In general, properties depend on chemical composition, microstructure, and dislocation density¹³⁶. The complex thermo-mechanical history affects these properties through the microstructural effects such as precipitation, transformation, annealing, recovery etc. Properties of the base metal and the weld metal available in the literature can not be used to simulate the structural response to welding, as these properties vary with temperature, time and strain histories. The temperature dependent momentary properties are suitable for FEA, and the temperature dependent mean values are more suited for linearized analytical solutions¹²³.

Density

Density is measured at room temperature (RT) and converted through the mean, linear thermal expansion coefficient at higher temperatures¹²³. Density drops with increasing temperature, and at the transformation temperature there is a discontinuity due to denser atom packing of the face-centered austenite compared to the body centered ferrite. The material density has a minor effect on heat transfer, and hence can be assumed to be constant. The uncoupled thermomechanical analysis field equations consider constant volume, and hence, material density is generally kept constant.

Specific Heat

The temperature dependence of specific heat is usually incorporated in a FEA program by evaluating the capacitance matrix at the given temperature. This value of specific heat then applies for the entire duration of the time increment. If the specific heat value undergoes a drastic change in the phase transformation range, then a large time step will introduce a significant error in the results. So fine time-steps should be used in transient heat transfer analysis. Tekriwal¹⁴¹ assumed that the specific heat remains constant above 800°K for mild steel. He used latent heat of fusion equal to 273790 J/kg with solidus and liquidus temperatures as 1700°K and 1755°K, respectively. The latent heat effects during melting and phase transformations can be considered through specific heat values. Goldak¹³⁶ used specific heat values considering latent heat of transformation. Latent heat release and absorption primarily affects the predicted weld pool shapes. Latent heat inclusion results in smaller fusion zone but longer weld pools and lower peak temperatures in the weld pool¹³⁹. This effect is accounted for by incorporating an artificial

increase in the value of the specific heat over the melting temperature range. The FEM strategy involves fine time stepping over this range so that these latent heat spikes are not missed. The Latent heat associated with phase transformations is similarly incorporated. This effect is small and can be ignored in the heat transfer analysis¹³⁹.

Thermal Conductivity

Thermal conductivity and diffusivity are based on the mobility of free electrons in the metallic lattice and decrease as the temperature increases. They decrease at a given temperature as the alloying content increases¹²³. Tekriwal¹⁴¹ used a continuously decreasing thermal conductivity value from R.T. to 1950°K. For simulating the weld pool stirring, he artificially hiked thermal conductivity values for temperatures above 2000°K which were twice the value at sub-liquidus temperature (1950°K). Goldak et al¹³⁶, used thermal conductivity values which were 4 times the value at sub-liquidus temperature to simulate the weld pool stirring effect.
SIMULATION OF WELDING RESIDUAL STRESSES

Welding residual stress analysis using numerical approach involves difficult and immense efforts to get a detailed solution for the following reasons¹²³:

- To allow for the different cooling conditions in the interior and on the surface, the model should be 3-D, at least in the weld area

- The process to be modeled is transient to a high degree, with extremely different field gradients dependent on position and time because of the rapid heating-up and cooling-down cycles

- The thermo-mechanical material behavior is highly non-linear

- The instantaneous, local material behavior is dependent on the local thermal and mechanical stress and strain history

- Addition of filler metal is difficult to model

- Simulation of state and microstructure changes

- Presence of defects and cracks makes the continuum concept doubtful

This extremely complex problem is further complicated by the following facts¹²³:

- Solution to this problem requires powerful computers, excellent solution algorithms and ideally, self-adaptive meshes and procedures. The convergence proofs and error management pose great difficulties.

- Unavailability of temperature dependent material property data for most of the industrially used materials.

Therefore, the problem is consequently reduced to its central issue, and only influencing parameters dominant in the respective case are presented in the finite element model¹²³

THEORY

The accuracy of a finite element stress solution depends on the constitutive model which describes the material behavior in the elastic-plastic regime¹³⁹. The FE simulation of the welding problem is undertaken in a series of small time increments from the start to the finish of the welding process¹³⁹. In each time increment, Δt , at each point in the material the temperature increment, ΔT , is computed first. Then the total strain increment, $d\epsilon_{ij}$ is predicted. It is composed of an elastic strain component, $d\epsilon_{eij}$, a thermal strain increment $d\epsilon_{Tij}$, a plastic strain increment $d\epsilon_{pij}$, a plastic strain from transformation plasticity $d\epsilon_{tpij}$ and transformation strain $d\epsilon_{trij}^{123}$.

$$d\varepsilon_{ij} = d\varepsilon_{eij} + d\varepsilon_{pij} + d\varepsilon_{tpij} + d\varepsilon_{Tij} + d\varepsilon_{trij}$$
(23)

During the first iteration of each time increment, the material stiffness is calculated based on an elastic assumption regardless of whether yielding has occurred in the last iteration or not. This allows the incremental solution to be obtained in one iteration if elastic unloading takes place in that increment. If yielding occurs, then subsequent iterations use an elastic-plastic stiffness matrix. Hooke's law, which is subdivided into deviatoric ε_{dij} and dilatoric or volumetric ε_{vij} portion, applies to elastic strain.

$$\varepsilon_{dij} = \frac{1}{2G} \sigma_{dij}, \quad \varepsilon_{vij} = \frac{1}{3K} \sigma_{vij}$$
 (24)

Shear modulus G and compression modulus K can be expressed by the elastic modulus E and Poisson's ratio v:

G =
$$\frac{E}{2(1+\nu)}$$
, and K = $\frac{E}{3(1-2\nu)}$ (25)

Next the associated stress increment $d\sigma_{ij}$ is predicted. The stress increment must satisfy the yield criterion. The combination of yield condition, yield law and hardening law applies to the plastic strain. The yield condition designates the initiation of yield in the multi-axial stress state¹²³. The yield law correlates the plastic strain increments with the momentary stress state and with stress increments. The hardening law states how the yield limit is changed by yielding.

Most metals obey von Mises yield criteria. It states that attainable stress states are con-

strained to lie within a cylinder in stress space or, equivalently, a sphere in deviatoric stress space¹³⁶. The yield condition is represented as¹²³:

$$0.5\sigma_{\rm dij} \ \sigma_{\rm dij} - \frac{1}{3} \ \sigma_{\rm ys}^2 = 0 \tag{26}$$

The stress increment must satisfy the yield criterion. This is called the consistency condition¹³⁶. If plastic deformation occurs, the plastic strain increment $d\epsilon_{pij}$ must be calculated. It is not always recognized that, in tri-axial stress states, individual stress components such as σ_{xx} can be far greater than the uniaxial yield stress without causing plastic deformation. The obvious example is that no matter how great the hydrostatic pressure is, no plastic strain will occur in metals. The yield law states that the plastic strain increment is coaxial and proportional to the deviatoric stress :

$$d \epsilon_{pij} = d\lambda \sigma_{dij}$$
 (27)

Strain hardening that changes the radius of the sphere while holding its center fixed is called isotropic hardening. Strain hardening that maintains the radius fixed but moves the center of the sphere is called kinematic hardening. Isotropic-Kinematic hardening permits both the position of the center and the radius to change with strain hardening. The strain-hardening (contained in $d\lambda$) is represented by the Tangent (plastic) modulus, E_T , and the effective plastic strain increment $d\epsilon_p^*$:

$$E_{\rm T} = \frac{d \sigma_{\rm ys}}{d\epsilon_{\rm p}^*} \qquad (28)$$

$$d \varepsilon_p^* = \frac{\sigma_{dij} d \varepsilon_{pij}}{\left(1.5\sigma_{dij} \sigma_{dij}\right)^{\frac{1}{2}}}$$
(29)

In addition to strain hardening, the yield function is a function of temperature. Usually the

radius decreases at higher temperatures. However, metallurgical phenomena such as precipitation hardening can cause hardening as temperature increases. In passing through a phase change such as the austenite-ferrite transformation in steel, the yield strength may be very low¹²³.

Thermal-Stress Contribution

Thermal stresses are calculated based on thermal strains. The stress is related to the strains by :

$$\{\sigma_{ij}\} = [D](\{\varepsilon_{ij}\} - \{\varepsilon_{Tii}\})$$
(30)

Where $\{\sigma\}$ = stress vector,

[D] = elasticity matrix, is presumed to be a symmetric matrix and is positive definite

 $\{ \varepsilon_{ij} \}$ = strain vector, and $\{ \varepsilon_{Tii} \}$ = thermal strain vector.

The above equation can be rewritten to yield :

$$\{ \epsilon_{ij} \} = \{ \epsilon_{Tii} \} + [D]^{-1} \{ \sigma_{ij} \}$$
 (31)

The thermal strain ε_T is given by the dilatoric strain components :

$$d \epsilon_{Tii} = 3 d \epsilon_T = 3 \alpha \Delta T$$
 (32)

For the 3-dimensional case, the thermal strain vector is :

$$\left\{ \begin{array}{c} \varepsilon_{\text{Tii}} \end{array} \right\} = \begin{cases} \alpha_x & \Delta T \\ \alpha_y & \Delta T \\ \alpha_z & \Delta T \end{cases}$$

Where $\alpha_x \alpha_y \alpha_z$ are average coefficients of thermal expansion in the x, y and z direction, respectively. $\Delta T = T - T_{REF}$, T = Current temperature at the point in question. T_{REF} - reference (strain-free) temperature.

Transformation Strain

The dilatoric transformation strain, ε_{tr} , depends on the defining quantities of the transformation process¹²³. The transformation-plastic (deviatoric) strain ε_{tr} is set proportional to the stress σ (assuming uniaxial low stress):

$$\varepsilon_{\rm tp} = K_{\rm t} \sigma$$
 (33)

The factor K_t is found proportional to the transformation related volume change and has been determined experimentally and by microstructural calculations¹²³. Within a simplified algorithmic procedure, transformation plasticity is described by a severe reduction of yield stress in the transformation temperature range setting strain hardening to zero in this range.

From the above formulation, the constitutive equation of the thermo-elastic - plastic continuum can be written in matrix notation as :

$$\{ d\sigma \} = [D] \{ d\varepsilon \} + dC^* \{ \sigma_d \} - M^* \{ d\varepsilon_T \}$$
(34)

The six components of the respective tensor are summarized in the column vectors { d σ }, { d ϵ }, { σ_d } and { $d\epsilon_T$ }. The momentary thermo-elastic-plastic stress-strain matrix [D] and the coefficients dC^{*} and M^{*} depend in a complicated manner on the temperature-dependent material characteristic values G, K, E_T , σ_{ys} , α and also on the equivalent stress^{144, 145}. The cumulative up plastic strains are reset to zero during fusion and added up again following solidification. The system matrix equations and FE formulation are described in ANSYS theoretical manual¹²².

MODEL SIMPLIFICATIONS

The FE solution to the complex welding stress analysis can be applied easily with some simplifications. The simplifications, for example can be¹²³:

- Reduction of the three-dimensional mechanical model to a two- dimensional or even to a onedimensional mathematical model. - Simplification of geometry, support and load conditions.

- Taking advantage of symmetry.

- Reduction of the non-linear thermoelastic-viscoplastic model to a linear thermoelastic model.

- Decoupling of the thermal and structural processes.

- Ignoring defect or crack formation.

- Omitting the fusion and solidification phase as well as the transformation processes which take place at elevated temperature and consequently at low yield criteria.

- Registering the transformation at lower temperature only globally in the specific heat and thermal expansion coefficient.

- Ignoring creep and hardening as well as introducing simplifications in yield laws.

- Simplification of groove shape and layer structure.

- Replacement of heat source movement by momentary application of total heat quantity or by rapid heat source movement, ignoring heat propagation in the direction of movement.

- Replacement of the temperature-dependent material characteristic values by temperatureconstant averaged values in the decisive temperature range.

- Modeling of the formation of residual stresses as a pure cooling process.

As the melting point is approached, Young's modulus and the yield strength approach zero. Since a small force will then produce large displacements, the stress analysis will fail at some temperature. The technique to overcome this is simply to set temperature that are above some cut-off temperature¹³⁶. Cut-off temperature can be defined as the temperature above which the stress history is almost meaning less as it does not have any effect on the developing residual stresses. Patel used a cut-off of 1200°C. Tekriwal¹⁴¹ studied the effect of different cut-off temperatures and found that it had a negligible effect on the residual stress state. Use of cut-off temperature appears to be adequate for predicting residual stress and strain¹³⁶.

Since the molten metal solidifies as a damage-free material, any deformations occurring

above the melting point will not affect the final residual stress¹³⁹. Therefore, the heating part can be neglected and stress analysis can be carried out only during the cooling part. Ueda⁴⁵ discarded all of the temperature-time curve except the cooling part of the thermal cycle below 800°C.

Dimensionality of the Problem

The transient and the residual stress distribution in a welding operation is threedimensional¹³⁹. But to conduct a fully 3-D analysis is prohibitively expensive. The problem can be reduced to 2-D by analyzing the whole plate if the weld bead length is long ¹⁴⁶ or by analyzing the whole plate if it is thin. In the 2-D analysis, the restraint in the third dimensions must be specified¹³⁹.

For a simply supported plate, out-of-plane deformation is possible and is modeled by using generalized plane-strain elements¹³⁹. For a plate clamped on either side of the weld, rotation out of plane of the analysis is prevented by the clamp but the plate is free to slide under the clamp. For a plate fully fixed along all its edges (similar to a weld in the middle of an infinite plate) all transverse and longitudinal shrinkage is prevented. Thus the actual restraint condition will decide the dimensionality of a given problem.

Leung et al¹³⁹. observed that the longitudinal stress and the longitudinal tensile stress zone changed negligibly when zero plane strain elements or generalized plane strain elements were used . This suggests that, with the exception of the start and end of the weld, a 2-D analysis is sufficient for most purposes. A generalized plane-strain analysis indicates rotations out of the plane of analysis and, therefore, can be used to give an estimate of plate bowing and uneven long-itudinal shrinkage with little increase in computational cost¹³⁹.

Restraint Conditions

Various restraint conditions in the plane of the analysis substantially alter the transverse residual stress distribution and to a lesser extent the longitudinal residual stress distribution.

Prevention of shrinkage and bending tends to increase the tensile stresses in the transverse direction and by Poisson's effect in the longitudinal direction¹³⁶. For a simply supported plate, the transverse stresses developed are small with the maximum tensile stresses located just inside the fusion zone¹³⁹.

Mahin et al¹³⁸ performed 2-D plane stress numerical analysis welding of a 4.7 mm thick plate and compared the results with experimentally determined residual stresses using neutron diffraction. The computed residual stresses were in reasonable agreement with neutron diffraction measurements except near the weld zone. From these results, these authors point out that the 2-D plane stress assumptions are not valid near the weld for either the heat transfer or the stress analysis. With the advent of faster computers and better algorithms, a full 3-D analysis in reasonable time is possible.

Selection of Proper Elements

Patel¹⁴⁷ used 20 noded three dimensional brick elements. Tekriwal used 3-D eight-noded brick elements to discretize the mesh for heat transfer as well as stress analysis. In 3-D analysis, brick elements are preferred because they tend to be more accurate and easier to use and interpret than tetrahedral elements¹³⁶.

Time Stepping

Most FEA analysis performed to date have used the same mesh and the same time stepping scheme for both the heat transfer and the stress analysis, because the programming is simpler¹³⁶. The stability and accuracy characteristics of the heat transfer and stress-strain analysis have fundamental differences¹³⁶ Therefore, the computational efficiency can be improved by using different time steps in the two analyses. In broad terms, the stress analysis requires shorter time steps during heating and permits longer time steps during the later part of the cooling phase. Use of this strategy requires that temperature-time data be interpolated or mapped onto the time stations in stress-strain analysis¹³⁶. Since the heating part can be neglected, the problem size can be

reduced. Finer time-stepping should also be considered in the phase transformation temperature range.

Karlsson and Josefson¹⁴⁸ used 240 steps of 0.23 seconds for the welding process and 50 steps of increasing length for cool down to R.T., while simulating the stress in girth weld. Lindgren and Karlsson¹⁴⁹ used 89 steps for 89 seconds of welding time and 25 steps for cool-down to R.T. in 14000 seconds for computing transient and residual stress for a complete girth weld. The choice of time-stepping depends on the computational algorithm and available computer configuration.

THERMO-MECHANICAL PROPERTIES

For residual stress analysis, the following temperature dependent thermomechanical material properties are required: Thermal expansion coefficient, α , Elastic modulus, E; Poisson's ratio, ν ; Yield strength, σ_{ys} ; Tangent or plastic modulus, E_T; Density, ρ . In addition, α and σ_{ys} values at transformation temperatures are required.

Thermal Expansion Coefficient

The momentary ("differential") thermal expansion coefficient, α , measured in the dilatometer rises with the temperature and is higher for high-alloy steels than for mild and low alloy steels¹²³. At the transformation temperature, A_{c1}, an abrupt change to negative values occurs, caused by the transformation strain acting in the opposite direction to the thermal strain. At the melting temperature, a considerable increase in volume occurs and solidification has an opposite effect. This is generally of no consequence for the formation of welding residual stresses, however, because the yield stress drops towards zero at melting or solidification temperature.

Tekriwal¹⁴¹ used α values which increased with temperature up to 700°C and then decreased due phase transformation from ferrite to austenite between 700°C and 900°C and again increased above 900°C. Patel¹⁴⁷ used gradually increasing α values until 700°C and then maintained it at a constant level for higher temperature.

Elastic Modulus

The modulus of elasticity, E, for metals decreases with increasing temperature, and above a certain temperature decreases rapidly. The elastic modulus drops rapidly near the transformation temperature A_{c1} while Poisson's ratio rises sharply (towards 0.5)¹²³. As noted by Michel¹⁵⁰, this point of departure corresponds approximately to one half of the melting point on the absolute temperature scale. In addition to the temperature dependence, E also depends on strain rate and thermal history, especially in the high temperature range. The deformation in high temperature range is elasto-plastic due to the low E_T . Since the stiffness matrix is formed initially based on E, which is higher than E_T , it takes more iterations and consequently more time for convergence. For this reason some researchers have adopted lower than actual values for the elastic modulus in the high temperature range^{139, 147}.

Yield Stress

In general, yield strength decreases as temperature increases. The yield strength is greatly reduced in the temperature range of transformation compared to the lower yield stress of the two phases concerned. In the case of microstructural transformation in the presence of stress, a plastic deviatoric transformation strain occurs in addition to the volumetric transformation strain. This plastic anomaly, termed "transformation plasticity", is caused by second order residual stresses resulting from the discontinuous distribution of transformation progress in neighboring crystal-lites¹²³.

Yielding is defined by von Mises criterion. Once yielding has occurred an associated flow rule is used to calculate the incremental plastic strains. A rate dependent plasticity model with kinematic hardening or isotropic hardening is generally adopted to characterize the material behavior during welding. The kinematic theory simplistically models reverse plasticity and Bauschinger effect that is expected during welding¹³⁹. The plastic behavior is assumed through a constant work-hardening slope. A more accurate model would ideally be the combined

kinematic-isotropic hardening model.

As solid material looses its strength completely at liquidus temperature, it is reasonable to assume zero yield strength at the liquidus temperature. But, specifying zero yield strength results in division by zero in analysis, and therefore, many researchers have used a low value of yield strength approaching but not equal to zero. Tekriwal¹⁴¹ used yield strength value of 1 MPa for mild steel at the liquidus temperature which was approximately 0.4% of the value at RT.

Plastic Modulus

In a kinematic hardening model a zero E_T must be specified at temperatures above melting so that the molten material can resolidify as damage-free material. The yield strength at melting should be very low but not zero as this may create division problems in computation. Leung¹³⁹ found that a value of about 0.5% of the R.T. value for temperature above melting would give reasonable results. The temperature dependent E_T can be approximated based on the assumption that the E_T / E ratio remains constant up to the melting temperature¹⁵¹. Tekriwal ¹⁴¹ used a very low value of $E_T = 10^{-6}$ MPa (1.0 N/m²) which will introduce insignificant error. Patel¹⁴⁷ assumed that E_T is 11.1% of E at all temperatures for a 0.23% steel and this generated higher E_T values.

Phase Transformations

Allotropic phase transformations occur in welding of ferritic steels. These transformations are accompanied by specific volume changes which can alter residual stresses. These effects can be neglected during heating because the accumulated damage will be relieved at high welding temperatures. During cooling, the material expands as it transforms from austenite to different transformation products depending on the cooling rate and composition. This expansion opposes the thermal contraction, thus the magnitude of residual stresses may be reduced¹³⁹. Mok and Pick¹⁵² have found that this transformation effect can be neglected unless the transformation occurs at very low temperatures, the volumetric strain is comparable to the thermal contraction

strain or there is rapid cooling of the weld pool. Jones and Alberry¹⁵³ observed directly that, during an austenite-bainite transformation, the developing residual stress falls to zero. During an austenite-martensite transformation, the stress becomes compressive. In part, this is because the volume change to martensite is greater.

Poisson's Ratio

The Poisson's ratio, v, generally increases with temperature, but data at temperatures above 1000°C are not available. Patel¹⁴⁷ used v varying from 0.29 at R.T. to 0.48 at 1000°C and above for a 0.23% carbon steel. Tekriwal¹⁴¹ studied the effect of three different values of v as a function of temperature on residual stress distribution and found that v had no significant effect on final residual stress distribution. At higher temperatures, where the material behaves in a plastic manner, the Elastic Modulus and Poisson's ratio have little influence on the calculated stresses. A Poisson's ratio approaching 0.5 at temperatures above the melting point is not necessary because the material is assumed to behave in a fully plastic manner ($E_T = 0$) so that the incompressible condition is satisfied. In addition, specifying a Poisson's ratio approaching 0.5 is computationally unattractive, since during the first iteration in each increment of the solution the material is assumed to behave elastically. This can result in a large initial material stiffness matrix, thus causing the solution to diverge or at best converge slowly¹³⁹.

Density

Tekriwal¹⁴¹ assumed a constant density (ρ) of 7870 kg/m³. At higher temperatures metal usually expands resulting in lower density. However, the uncoupled heat conduction analysis assumes that the material occupies the same volume throughout the analysis. Therefore a constant density needs to be used due to the model limitations. Patel¹⁴⁷ used decreasing density as a function of temperature.

FEA OF WELDABILITY SPECIMENS

Earliest thermal elastic-plastic analyses using FEA of Tekken tests were performed by Ueda⁴⁵. Ueda⁴⁵ assumed idealized cross section of the specimen, so that the analysis was reduced to 2-D and the effect of groove shape on local stresses was neglected. Uncoupled thermal and stress analysis was carried out. Separate thermo-mechanical properties were used for both the base metal and weld metal. Thermal analysis was conducted by using the FDM. Stress analysis was performed using the FEM. Filler metal addition was incorporated in this analysis.

Additional studies on Tekken test welding stress relieving by annealing using FEM were performed by Ueda and Fukuda¹⁵⁴. They found that there is essentially no difference in the resulting residual stresses produced by a moving heat source and by an instantaneous heat source. The problem was further simplified using an instantaneous heat source and by assuming that the residual stress distribution is symmetric with respect to the two orthogonal axes. Additionally, they used simple groove geometry so that effect on local stresses was neglected. They observed that the magnitude of residual stresses decreased after stress relieving heat treatment. Ueda et al¹⁵⁵ also analyzed effective restraint intensities as a function of groove shape and bead eccentricity in a Tekken test specimen by a 3-D elastic analysis using the FEM.

Finite element analysis of Tekken specimens was also attempted by Andersson^{112, 117}. The temperature distribution was calculated using two different 2-D numerical models, Figure 61. The thermo-elasto-plastic stress analysis in the HAZ was modeled by using two 2-D models. Model A was used to determine the transverse mean stress in the weld metal under assumed plane stress conditions in the XY-plane. Model B was used to determine stresses in the HAZ. The mean transverse stress calculated (at point a) with model A was applied in model B. The longitudinal strain $\varepsilon_x(y,t)$ as obtained from model A was prescribed in Model B assuming a uniform strain ε_x through the plate thickness.

It is clear from the literature survey that not many researchers have attempted thermo-

mechanical behavior simulation. In fact, there is no research work detailing thermo-elasto-plastic behavior of Lehigh weldability test specimens. The earlier efforts by Ueda had inherent assumptions which might be all right for global stress distribution understanding. Since HAC is a localized phenomenon, it is imperative to use as realistic groove shapes and welding conditions as possible. Andersson's attempt to include different material properties for base metal, HAZ and weld metal along with phase transformation effects is noteworthy. In spite of this, the use of two 2-D models does not represent the actual Tekken test weld response. Additionally, the stress distribution at the crack initiation site (double bevel weld root notch) was not addressed in Ueda's as well as Andersson's analysis. As discussed earlier, for realistic simulation of localized thermal and stress behavior, full 3-D analysis is necessary. Therefore, in this study 3-D thermo-elasto-plastic simulation of Tekken weldability test was attempted.

OBJECTIVES OF THE STUDY

The objective of this study was to get an idea of the overall thermo-mechanical response of weldability specimen. The overall aim was to study how well suited are the Tekken and Lehigh weldability tests for assessing the HAC response of actual weldments. The study specifically focussed on:

- Transient thermal field developments and distribution in Tekken and Lehigh weldability test specimens
- Residual stress development and its localized distribution along the weld zone in Tekken weldability test specimens.
- 3) The effect of preheating temperature on residual stress distribution and cooling behavior.
- 4) Detailed analysis of the residual stress distribution at crack initiation location.

4.2 - NUMERICAL PROCEDURE

4.2.1 - THERMAL ANALYSIS

Introduction

The objective of this thermal-stress analysis exercise was to get an idea of the overall thermo-mechanical response of weldability specimens. The analysis was simplified by making certain assumptions about different phenomena which will not have any adverse effect on the temperature and stress distribution. Most of these assumptions were forced from the limited amount of available computational configuration and material property data.

ASSUMPTIONS

The heat source was simplified by assuming a distributed temperature distribution over the weld metal volume at a particular instant of time. Heating of the weld metal beneath the heat source to temperatures above the melting point was carried out in a single load (time) step. The nodal temperatures were prescribed instead of using a distributed heat flux as it allowed melting in fewer time steps, where as, with nodal heat flow rates, melting required many load steps which increased the problem size dramatically. In addition, a good control over the molten pool size could not be exercised when the heat flux boundary condition was used. Although, prescribing nodal temperature distribution is a coarse representation of the heat source, the temperature fields will not be significantly altered, since the weld pool size determines the temperature distribution.

Weld pool convection was not modeled in the present analysis. Convective heat transfer in the molten weld pool was considered by artificially hiking the thermal conductivity. Latent heat of fusion and austenite-ferrite transformation during cooling was considered. The values were obtained from Patel¹⁴⁷. This was considered by increasing the specific heat in the liquidus and phase transformation temperature range.

Instead of individually specifying the convective and radiative losses during welding (arc on), arc efficiency was used to control the amount of effective heat input. In this analysis, a arc efficiency, η , = 0.8 was used to consider the effect of hydrogen addition to the shielding gas. In the literature, a value of η = 0.75 is more common³⁴.

The surface heat transfer coefficient, h_f , is a function of temperature, but in the present analysis, a constant value of h_f was assumed. This is a reasonable approximation as the temperature rise in the air surrounding the specimen surface is negligible, the steel surface is oxidized (heavy mill scale) and the residence time at high temperatures is significantly low to not have any substantial effect on convective heat transfer.

Another assumption arose from the limitation of the FEA program. The present version of ANSYS used in the program, ANSYS 4.4A, does not have the capability of dynamically adding the filler metal during the transient analysis. Therefore, filler metal was assumed to be preexisting before welding. The weld metal deposition was carried out by melting this pre-existing weld metal volume directly under the heat source. This simplification does not pose any problem in the thermal analysis as the heat transfer in longitudinal direction in front of welding is negligible for the welding speeds used in this study.

FINITE ELEMENT MESH

Tekken Specimens

To reduce the computational cost and time, only symmetric half part of the weldability specimens was simulated. The Tekken weldability specimen is not symmetric along the weld centerline because of the oblique-y groove. Even then, the Tekken thermal analysis was carried out assuming symmetry along the weld centerline since the other half does not experience significantly different thermal field. In the subsequent structural analysis, only the left half of the specimens was used. For Tekken specimens, the double bevel side is only important as cracking generally starts from the double bevel side side of the groove, Figures 19, 33-35, 44. The 3-D FE

mesh for the Tekken specimen is shown in Figure 62(a) and (b). The dimensions of the Tekken specimen were as shown in Figure 15. The specimen thickness was 25.4 mm (1").

Eight noded isoparametric thermal brick elements were used. Since the regions near the weld experience large thermal gradients, the mesh was finer in this region. Farther away from the weld region, the thermal gradients are less steep and hence coarser mesh was used. To optimize the computational costs and the desired accuracy the mesh was graded Figure 62(a) and (b). Nodal connectivity was ensured throughout the domain. The temperature gradients are steeper in the initial part of the weld thermal cycle and are shallower later in the cooling cycle. Therefore, the time steps were much shorter in the beginning part of the analysis and longer in the later part of the analysis.

Lehigh Specimens

Using the discretization concepts discussed in the previous section, a FE mesh was developed for the Lehigh specimen, Figure 63(a) and (b). Unlike the Tekken Specimen, the Lehigh specimen is symmetrical along the weld center line. Due to the meshing complexities, the 12.5 mm (1/2") diameter holes at test groove ends were assumed to square gaps of smaller dimensions, Figure 63(a). This will unrealistically augment the residual stress state at these locations. The specimen thickness was 25.4 mm (1").

HEAT SOURCE MODEL

The weld pool shape used in the numerical analysis is slightly different than actual weld pool shape observed in the experimental weldability test specimens. Actual weld metal shape is un-symmetrical, Figure 64. The exact representation of the actual weld shape in the numerical scheme will require a much finer mesh. In addition, the weld metal shape varied along the weld length in experimental specimens. Weldability specimens welded with different shielding gases exhibited wide ranging weld pool shapes. It was, therefore, impractical to simulate the exact weld metal shapes in the numerical simulation procedure. Since the final objective was to gain a better understanding of the temperature distribution rather than specific temperature values, uniform weld metal shape over the entire weld length was selected, Figure 62(b).

The heat source was modeled, by prescribing a gaussian temperature distribution to the nodes under the heat source in the weld metal. This was achieved by specifying the highest temperature for the node at the center of the heat source, and by decreasing the temperature exponentially with node location away from the center both laterally and vertically, Figure 65. The nodes on the weld pool boundary were maintained at 1490°C. This is similar to the radial temperature distribution in the molten weld pool. The region immediately below the arc experiences the highest temperature and it dropped to the melting temperature (liquidus) as the weld pool boundary is approached. This was the most appropriate method of controlling the weld pool size for the finite element mesh used in this study. The effect of different peak temperatures on total cooling behavior was initially studied. The most appropriate peak temperature distribution was then selected.

INITIAL AND BOUNDARY CONDITION

The weldability specimens were welded at either at RT or at a T_{ph} depending on the actual test conditions. The initial temperature as well as any selected preheating temperatures can be prescribed by using a parameter TUNIF in the ANSYS program. This was the initial condition in the analysis.

Only convective heat losses were assumed in this analysis. The convective heat transfer during and after welding can be prescribed by specifying the convective heat transfer coefficient at the surfaces, i.e., natural boundary condition. A temperature independent heat transfer coefficient of 15 W/m°C was used.

The symmetry boundary condition can be simulated by assuming that the boundary is adiabatic. Adiabatic boundary is specified in the ANSYS program by not specifying any heat flow rates, nodal temperatures or surface heat transfer coefficients on that boundary.

MATERIAL PROPERTIES

Temperature dependent material properties were used in this analysis. Thermo-physical properties, such as, specific heat and thermal conductivity could not be obtained for the A514 base metal used in this study. Therefore, material properties for AISI 1020 steel were obtained from Patel, Figure 66 (a),(b)¹⁴⁷. Temperature dependent mechanical properties were obtained from Department of Defense Handbook¹⁵⁶. These properties were extrapolated to higher temperatures, since the database was limited to temperatures below 800°C. A constant temperature independent density value of = 7860 kg/m³ was used.

SOLUTION PROCEDURE

A transient thermal analysis was performed using temperature dependent material properties. The initial temperature or the preheating temperature was specified by the TUNIF command in the ANSYS program.

Convective heat transfer was activated on all the surfaces except the symmetry surface as soon as welding was initiated. Welding was initiated by selecting nodes inside the expected weld pool and prescribing gaussian temperature distribution. The heat source movement was simulated by selecting next set of nodes under the instantaneous position of the arc. The heat source was moved piecewise by prescribing and deleting the temperatures in incremental fashion for given time intervals. In each time step, the heat source moved a quarter weld pool length or one element length. This was performed until the heat source reached the weld end. The welding time used in this procedure corresponded with the experimental welding time. The heat source movement was carried out by writing a sub-routine. At the end of welding, the specimens were allowed to cool on their own. Only convective heat transfer was allowed in this time period. The thermal analysis was carried out until the specimens reached room temperature.

An automatic convergence criterion was used through out the thermal analysis to optimize the solution times. During the welding part, a convergence criterion of 1°C was used, since steeper temperature gradients are encountered. As soon as welding was over, the convergence criterion was relaxed to 2°C as the temperature gradients are less steeper.

THERMAL ANALYSIS OF TEKKEN AND LEHIGH TEST SPECIMENS

The thermal response of Tekken weldability tests was analyzed for one heat input condition and at three preheating temperatures of RT (22°C), 130 and 175°C. Uniform preheating temperature was assumed. These conditions correspond to those used in the experimental Tekken test study. Effect of these preheating temperatures on 1500 to 100°C and 800 to 500°C cooling rate was studied. Additionally, thermal cycles at the weld start, weld mid-length and weld end were analyzed to characterize the effect of dimensionality changes and weld start and weld end effects. Instantaneous temperature contours at different times were analyzed to study temperature distribution development. For Lehigh test specimens, thermal analysis was performed only at RT.

4.2.2 - NUMERICAL PROCEDURE - STRESS ANALYSIS

INTRODUCTION

A half symmetry 3-D model was used, since the temperature as well as the residual stress distribution in weldability test specimens is 3-D due to the slot nature of the test specimen. Because of the complexities of underlying phenomena, unavailability of material property data base and limited computer configuration, simplifying assumptions are necessary. These assumptions are discussed in the following section. It should be noted that these simplifications only affect the magnitude of residual stresses but not the residual stress distribution.

ASSUMPTIONS

Most of the assumptions were due to the unavailability of temperature dependent mechanical properties for different weldment regions. Therefore, same material properties were used for base metal, weld metal and HAZ. This will change the magnitude of the resulting residual stresses. But, the overall residual stress distribution will be unchanged. Since the objective of this research work was to study the residual stress distribution instead of specific stress values, the final conclusions will not be affected dramatically with this assumption.

The weldability test specimen geometry was simplified as meshing became quite complicated with the original geometry. This was especially true for Lehigh test specimens. These specimens have a bottom - inverted U slot and two 12.5 mm (1/2") diameter holes at the weld start and weld end. It is quite difficult to get rounded shapes with brick elements unless the mesh density is quite high. With tetrahedral elements this difficulty can be circumvented. Since it was decided to use brick elements because of their accuracy, ease of interpretation and lower number of elements, the specimen geometry was simplified. This resulted in sharp corners and simplification of curved shapes, but it does not influence residual stress distribution in the regions of interest, i.e. weld metal and HAZ regions. Symmetrical model was used to decrease computational time in the thermal analysis. This may not always be realistic but is required because of computation configurational limitations. The Lehigh weldability specimen is symmetric along the weld centerline and symmetrization is possible. The Tekken specimen is not symmetric along the weld centerline axis because of the oblique y-groove. A full FE Tekken model was not possible because of limited computer configuration, hence, it was decided to only model the double bevel side of the Tekken specimen. The rationale behind this was the observed fact that in Tekken specimens, HAC generally occurs from the weld root notch on the double bevel side. Because of this assumption, the model in reality represents a X-groove weld joint. In addition, a little extra material on the single bevel side will not affect restraint and heat transfer significantly so that symmetrization should not affect final residual stress distribution. The symmetry boundary condition was specified in the structural analysis by selecting the symmetry surface and using an ANSYS command.

The 3-D FE mesh was developed by adding the expected weld filler metal and closing the welded section of the groove prior to the transient thermal analysis. A more representative model would be the one that allows the addition of filler metal dynamically. This dynamic addition of filler metal (also known as element birth and death capability) is not possible in the current version of the ANSYS program. This pre-existing filler metal increases the restraint acting on the cooling weld metal, but, the net effect of this assumption on the final residual stress distribution will not be substantial, because, at higher temperatures, the weld metal has very high plasticity, and hence will not support any load. In other words, the transient stress build-up will be negligible and will not affect the final residual stresses.

The phase transformation expansion effect was considered to be similar during heating as well as during cooling because of the available thermal expansion coefficient values. The effect of transformation plasticity was neglected. A more representative treatment will involve incorporating the transformation plasticity effect for weld metal as well as different HAZ regions. A linear elastic-plastic stress-strain relation was assumed. The elastic and plastic regions were defined by elastic modulus and plastic modulus, respectively. The yield stress defined the separation between these two regions. Von-Mises yield criterion and associated flow rule with kinematic hardening were assumed to consider the Bauschinger effect. Since welding is short-duration at high temperature phenomenon, creep effects were neglected.

BOUNDARY CONDITIONS

The weldability specimens were welded without any external clamping. The residual stresses develop under self-restraint only. Therefore, no external restraints were applied in the analysis. Only symmetry boundary condition was used as mentioned earlier. To prevent rigid body motion during the finite element analysis, only two nodes on the specimen edge were fully restrained. This suggests that the weldability specimens are not restrained but are pivoted around these restrained nodes and the restraint free condition is realized. The symmetry boundary condition are shown in Figure 67.

SOLUTION PROCEDURE

A transient thermal analysis was first performed to obtain nodal temperature solution as discussed in a previous section. The nodal thermal history was saved to be input as the thermal load in the structural analysis. The same geometry and FE mesh was employed in both the thermal and structural analysis. For stress analysis, these nodal temperatures were mapped directly in the structural analysis part. In this study, the time stepping scheme was unchanged in heat transfer and stress analysis. Entire welding thermal history, i.e. heating and cooling parts of the thermal cycle, were used. In the ANSYS program, change of analysis type from thermal to structural is accomplished by specifying a parameter that reassigns the element type and corresponding attributes.

Material properties needed for the structural analysis, such as, Coefficient of thermal expansion, Elastic modulus, Yield stress and the tangent (plastic) modulus are shown in Figure 68 (a),(b),(c) and (d).

A stringent convergence criteria of 1% plasticity ratio was specified and time-step optimization was employed to automatically determine convergence checking and optimize on the number of iterations. Different load steps (thermal loads) were defined by reading nodal temperatures corresponding to certain time intervals. This load stepping scheme was essentially the same as the temperature history from the thermal analysis, i.e., the time stepping was much finer in initial part of the thermal cycle and gradually coarsened in the later part of the weld cooling.

The residual stresses were analyzed to study variation across the weld length, across the weld cross-section and in the transverse and longitudinal direction. The effect of preheating temperature on residual stress magnitude was also studied in case of Tekken specimens.

4.3 - RESULTS AND DISCUSSION

4.3.1 - THERMAL ANALYSIS

The Tekken test numerical results were compared with those from the 3-D analytical solutions, Table 39, because of the unavailability of Tekken test experimental heat transfer data. These analytical solutions were specifically developed for Tekken test specimens using 3-D heat transfer equations by Kasuya⁷⁶. It was observed that numerically determined 800 to 500°C, $t_{8/5}$, cooling rate differs significantly along the weld length. The numerically determined $t_{8/5}$ cooling time at weld start location was faster than the analytical cooling times. The numerical $t_{8/5}$ weld cooling times match well with those from the 3-D analytical equation for the location quarter weld length from the weld start. The numerical $t_{8/5}$ weld cooling times for the weld mid-length location were higher than those from the analytical equation. Numerically determined 1500 to 100°C cooling times were significantly higher than those predicted by the 3-D analytical equation. This trend was also observed with $t_{8/5}$ for higher preheating temperatures, Table 39. Whereas, $t_{15/1}$ did not show any difference along the weld length for higher T_{ph} .

The slower overall weld cooling rates from the FEA could be due to many reasons. Some of the more prominent ones are discussed below. The main objective of this study was to understand the temperature field development and distribution in Tekken weldability test and not to exactly quantify the thermal histories. It should be noted that the ensuing discussion will still hold even if more accurate numerical simulation was attempted.

The biggest contribution to deviation from actual thermal behavior was from the use of thermo-physical properties of AISI 1020 steel for the ASTM A514 steel used in this study. Specific heat and thermal conductivity values affect heat transfer immensely. Therefore, the use of inappropriate material properties will lead to an inaccurate simulation of the heat transfer process due to welding. The weld thermal cycles in Figure 69 (a) and (b) indicate a significant slow-

ing down of the weld cooling rates in 800 to 500°C range due to the release of latent heat of transformation after austenitic decomposition, especially for the weld locations at quarter weld lengths from the weld start and in the middle. This type of pronounced decrease in weld cooling rates was not observed in actual experimental weld thermal cycles, Figure 70. In this Figure, the weld cooling rates of A514 AWS size grooved specimen (25W x 12.5T x 80L mm), Figure 3, welded under identical welding conditions to that used in this study are presented. This experimental thermal cycle does not exhibit any latent heat of transformation release. In contrast, the amount of latent heat release in the present numerical simulation was significant, and thus reduced the overall weld cooling rates. Therefore, the use of AISI 1020 steel properties in the numerical simulation scheme resulted in slower overall cooling rates.

Another influential factor could be the thermal conductivity, K, values. Even a small deviation from the actual values over the entire temperature range will significantly affect the heat transfer process. In addition, the K data at temperatures above 800°C is extrapolated.

The use of a generic heat efficiency value from the literature can also contribute to the differences in weld cooling rates. In the actual experiments, shielding gases containing different concentrations of hydrogen were used. Addition of hydrogen makes the arc hotter and stiffer and these type of shielding gases are used to attain better penetration¹⁵⁷. The arc efficiency, η , values listed in the literature are for generic classification of welding processes. For example, the GMAW standard process employs inert gases or CO₂-inert gas mixtures. The η for the commonly used welding processes are well established. In this study, a value of 0.80 was used instead the more common value of 0.75 to account for the addition of hydrogen. In reality, the true heat efficiency value will vary between 0.75 to 0.80, thereby, affecting the actual heat input. Therefore, exact matching of numerical weld cooling rates with analytical cooling rates is not possible.

In addition, the limitation of the ANSYS program precluded the use of heat flux distribution and limited computer configuration prevented the use of internal heat generation rates (the mesh being too coarse). These quantities can be related more accurately to the actual heat input. Whereas, the nodal temperature distribution used in this study can not be precisely related to the actual heat input as the exact temperature distribution in actual weld pools is not well known.

The simplied symmetrical weld pool shape in the numerical model resulted in a slight increase of the overall weld volume. This will also significantly decrease the actual weld cooling rates. Additionally, in the numerical model, the weld pool shape was maintained at a constant level as approximated from the cross sections of the welded specimens. In reality, the weld metal shape is not constant at all along the weld length as discussed in a previous section on weldability testing. Due to the welding instabilities in the beginning the weld metal area is smallest. As the weld progresses and attains stability, the weld metal area becomes increasingly larger. As a result, the weld metal area is never constant along the weld length. Therefore, the use of a constant weld metal area throughout the weld length in the numerical model also adds to the overestimation of weld cooling rates.

All these factors combine together to yield somewhat distorted values of weld cooling rates. Hence, the numerical study focussed more on transient temperature distribution in the Tekken and Lehigh test specimens than on closely quantifying the thermal histories.

THERMAL FIELD DEVELOPMENT IN TEKKEN SPECIMENS

Figures 71-74 show temperature contour development at different locations along the weld length for Tekken specimens welded at RT. In Figure 71, the instantaneous temperature distribution at welding start is shown. In addition, the lowest temperature is -59° C, which is incorrect. This type of temperature oscillation is observed in the first time step only. Even with increasing the number of iterations in the first time step, it could not be avoided. In the next time step, the temperature again becomes positive. The reason behind this temperature oscillation could be the discontinuity near the weld start to simulate the 2 mm gap in actual welding. It is also observed from Figures 71-74 that molten pool shape as defined by contour H is not smooth. This trend was observed over the entire weld length as the weld pool moved. The elongated tear drop weld shape is also coarsely exhibited all along the weld length. This is because of the relatively coarse FE mesh. With an increase in mesh density, i.e. by adding more nodes inside the weld pool volume, the high temperature contours can be made smoother. Coarser isotherms were particularly observed for high temperature regimes, i.e., 1200 and 1500°C. Low temperature isotherms are smoother compared to the high temperature contours. The isotherms are well developed by the time the weld is half complete.

Another interesting feature of the heat source is the heating of pre-existing weld metal due to the heat conduction process in the welding direction. It is observed from Figures 71-74 that about 5 mm of the pre-existing weld metal in front of the heat source experiences temperatures in excess of 200°C. The presence of this preheated pre-existing narrow band of weld metal will decrease the overall weld cooling rates compared to the actual Tekken test weld. In real welds, the heat transfer ahead of the molten weld pool is absent due to the lack of any pre-existing filler metal.

DIMENSIONALITY OF HEAT TRANSFER IN TEKKEN TEST SPECIMENS

The specimen exhibits non-uniform temperature distribution very clearly. For example, when the heat source is at center of the specimen, the weld start location has already cooled down to 600°C, Figure 73. By the time the welding is complete, Figure 74, the welding start location is at about 325°C. This suggests highly heterogeneous temperature distribution during the welding of Tekken test specimens.

This type of temperature distribution affects the transient stress development during welding and will be discussed in detail in the stress analysis section. In addition, the non-uniform temperature distribution also affects the diffusible hydrogen content. For example, when the heat source is at the end of groove the weld start location has already cooled down to 325°C. This results in hydrogen diffusion out of the weld metal and into a thin HAZ layer surrounding the fusion line. Therefore, the coarse grained HAZ is already getting susceptible to HAC compared to the weld end location at this instant. This type of stress development and hydrogen diffusion is less detrimental as one moves towards the weld end location because of the higher weld metal temperatures.

Examination of lower temperature isotherms indicates that the heat transfer is more pronounced in the transverse direction during and after welding. By the end of welding, the low temperature isotherms, particularly 75 and 300°C suggest that heat transfer is mostly in the direction perpendicular to the welding direction. This is especially clear in Figure 74, where it is observed that the 75°C isotherm is more spread out in the transverse direction. In fact, this isotherm is almost symmetrical with respect to weld centerline. This trend becomes more pronounced in latter cooling stages, Figure 75 and 76. In these Figures, it is clearly observed that weld start and end locations cool at higher rates compared to the middle weld region.

The above observation suggests that heat transfer in Tekken test specimen is highly dimensional in nature. The faster cooling rates at the weld start and weld end locations are due to the additional heat transfer in the longitudinal directions. This occurs because of the extra material at the start and end locations due to the slot nature of the Tekken test. This additional material carrics the heat away, and hence, these locations experience faster cooling rates. In this sense, the heat transfer at these locations is truly 3-D. Whereas, in the center of the weld, the heat transfer pre-dominantly occurs along the direction perpendicular to the weld length. Therefore, the heat transfer for this region is mostly 2-D, the other heat conduction dimension being in the thickness direction. Heat transfer in most of the practical joints will be 2-D as weld end heat transfer effects will be absent in actual weld joints, since the welds run from plate edge to edge length wise and usually use run-on and run-off tabs as well. The Tekken test heat transfer is more typical of slot welds and repair welds in this sense and does not represent an actual weld cooling behavior.

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CONTROLLING WELD METAL SHAPE

The instantaneous near weld metal isotherm development at different locations of the heat source is shown in Figures 77-79. It is observed from these Figures that at the welding start, the isotherms are concentrated near the weld pool, Figure 77. With further progression of the heat source, the isotherms start to spread out in the transverse direction. After load step 6, the isotherms are stabilized, Figure 78. With further progress in welding, the individual isotherm location from the fusion line remains constant. This indicates that stabilization in temperature fields occurs only after the heat source has traveled about a quarter of the groove length. The test specimen is at RT and has not experienced any welding heat, therefore, the cooling rates are higher in the beginning. This results in faster heat dissipation in the early stages of welding. As the welding progresses, the test specimen experiences welding heat and after a certain distance the heat input and heat dissipation are in equilibrium. This results in stabilization of isotherm locations, Figures 78 and 79. In actual welding, the weld metal area is smallest at the weld start and steadily increases as the welding progresses similar to that observed for Lehigh specimens, Figure 40(a), (b). This is because of the instability in the welding process itself in the beginning combined with faster heat dissipation.

The isotherms in Figures 77-79 clearly show that the weld metal area can be most effectively controlled using nodal temperature distribution. The obvious problem with prescribing nodal temperature distribution is the fact that exact temperature distribution is unknown. Therefore, the peak temperature and the temperature distribution were chosen to produce a reasonable weld pool temperature distribution. Different peak temperatures and distribution had no significant effect on weld cooling rates.

WELD THERMAL CYCLES

The thermal cycles at the weld notch root on the double bevel side were compared in Figure 69 (a) and (b). It is observed from these Figures that the overall cooling rates are faster in the

weld start and end locations compared to those in the middle of the groove. If 1500 - 100°C cooling rates are compared, then the weld start location exhibits the fastest weld cooling rate of them all. This is especially true for no preheat (RT) case. It is also observed that the weld cooling rates decrease for the mid-length locations in the 800 to 600°C range, Figure 69 (a) and (b). This decrease in weld cooling rate is due to the release of latent heat of transformation. As discussed earlier, slower weld cooling rates due to the release of latent heat of transformation were not observed in the experimental study. This is because of the inflated latent heat of transformation were thermal cycles will better represent actual behavior. It is also interesting to note that the latent heat of transformation release is not pronounced at the weld start and end locations. This is because of the 3-D heat dissipation at these locations, whereas, in the middle weld regions it is 2-D.

The overall temperature distribution and weld thermal cycles indicate that thermal field development in Tekken specimens is highly dimensional and far from achieving a quasi-state condition. This is especially true in the beginning of the test weld. The variation of $t_{15/1}$ along the weld length indicates that use of a single $t_{15/1}$ to characterize residual hydrogen content at 100°C can be misleading. This is especially true for specimens welded at RT. For specimens welded at higher T_{ph} , the effect is less pronounced due to very large cooling times, Table 39.

EFFECT OF PREHEATING TEMPERATURE

As expected, with increasing preheating temperature, T_{ph} , the overall weld cooling rates decrease but the temperature distribution characteristics are unchanged compared to RT, i.e. no preheat, case. The 800 to 500°C cooling rate decreases significantly with an increase in the T_{ph} , Table 39. The weld cooling rates are considerably lower in the latter stages of the weld thermal cycle, i.e. below 300°C. This drastically increases the 1500 to 100°C cooling rate.

The observed decrease in weld cooling rates with an increase in T_{ph} is beneficial from HAC

susceptibility viewpoint in two ways. With an increase in 800 to 500° C cooling time, the formation of softer (less susceptible) microstructures is preferred in the HAZ. As a result, the resistance to HAC increases depending on the steel composition. Secondly, by increasing the 1500 to 100° C cooling times an increase in T_{ph} promotes hydrogen outgassing out of the weld zone. This is clear from the substantially increased weld cooling times with an increase in T_{ph}. In fact, if weld cooling times to RT are compared, then the beneficial effect of T_{ph} is even more pronouned. As a result, lower diffusible hydrogen content will be present in the weld zone and this will reduce HAC susceptibility of the weld joint. These beneficial effects are in addition to the reduction in residual stress which will be discussed in the section on stress analysis.

HEAT TRANSFER IN LEHIGH SPECIMENS

Similar temperature distribution characteristics were also observed for Lehigh specimens. Most of the discussion on Tekken test heat transfer results also holds for the Lehigh specimens, and, therefore, is not discussed here for brevity. Because of the longer test weld length in Lehigh specimens, the temperature distribution varies more prominently than in Tekken specimens, Figures 80 and 81. It is clear from these figures that the weld start location is at about 200°C when the heat source is at the weld end. Therefore, at this instant of time the weld start location is more susceptible to HAC compared to the weld end location. This is in agreement with the observed crack propagation from weld start to weld end location in experimental Lehigh test specimens.

The non uniform temperature distribution along the weld length is also observed in Lehigh specimens. The weld ends cool at a much faster rate compared to the middle region of the plate, Figure 81. As a result, the HAC susceptibility varies along the weld length in Lehigh specimens as observed for the Tekken specimens. Other features of heat transfer are essentially similar to those of the Tekken specimens.

4.3.2 - STRESS ANALYSIS

VALIDATION OF NUMERICAL RESULTS

Since no experimental residual stress measurements were performed, the numerical stress distribution results were compared with the experimental and FEA results by Ueda^{45, 154, 158, 159}. Ueda performed extensive 2-D, 3-D FEA as well as experimental residual stress measurements on Tekken specimens. This comparison with Ueda's results will be qualitative, since the welding conditions, model simplifications and assumptions in Ueda's study are different than those considered in the present study. The problem with experimental measurements is that residual stress measurement can not be performed near the weld zone due to the slot nature of the specimen and the test weld being embedded under the plate surface. Because of these limitations, Ueda^{45, 154, 158} performed residual stress measurements away from the weld metal and HAZ region and on the top plate surface.

The thermo-elasto-plastic FEA of Tekken test welds in Ueda's study was conducted with a number of assumptions. Some of them were discussed in the background section. Ueda's results that will be discussed in this section are from an experimental and numerical study conducted to analyze the effect of phase transformation on restraint stresses¹⁵⁸. In this study, Ueda performed thermo-elasto-plastic finite element analysis using various idealized mechanical properties in the region of phase transformation. The phase transformation temperature and strain at phase transformation as a function of various heating and cooling rates were experimentally determined. Instantaneous thermal expansion, yield stress and elastic modulus, E, were also experimentally determined as a function of temperature for a high strength steel, HT-80. The mechanical properties in the phase transformation range are difficult to determine, and hence were idealized but using different models M1 to M5, Table 40:

In model M1, the measured values of E from rigidity recovery temperature, T_m , to transformation start temperature, T_{cs} , were idealized by a straight line. The E values were same during heating as well as cooling stage even during the phase transformation temperature range. Yield stress hysteresis was assumed in this model.

- In model M2, the E values were linearly decreased from T_m to T_{cs} start temperature and followed the same path during heating and cooling. But at the T_{cs} , E value was dropped to zero and was linearly increased to the measured value at the transformation end temperature, T_{cf} . This is an idealized model for transformation superplasticity phenomenon. The yield stress was assumed to fall to zero at T_m and was interpolated to T_{cs} by a straight line. Yield stress hysteresis was also assumed in this model. In a similar fashion, these properties were varied for models M3 to M5 as shown in Table 40.
- 3 The thermal expansion coefficient, α , was varied for models M1-M4 as discussed below. From T_m (700°C), to the T_{cs}, the measured values were idealized by a straight line, and the value at T_m was used for temperatures higher than T_m. For phase transformation region, T_{cs} and T_{cf} were connected for idealization and α was determined. This α was applied to the transformation region as constant, Table 40. For model M5, α varies similarly until T_{cs} as in models M1-M5, but below T_{cs} it agreed with those in the heating stage, Table 40.

With these idealizations and different mechanical properties in the phase transformation region, Ucda¹⁵⁸ conducted thermo-elasto-plastic analysis on the slit weld specimen, Figure 82. It should be noted that this slit weld specimen configuration is almost identical to that of the Tekken test specimen except for the groove shape. The slit weld specimen used by Ueda has symmetrical Y-groove compared to the oblique y-groove in the Tekken specimen. This difference will only influence the localized stresses near the weld notch root but the overall residual stress distribution will be almost identical since these specimens have similar restraint intensity. The exact details of the numerical procedure adopted were not described in detail in Ueda's paper. The thermal analysis was conducted using the FDM and the elasto-plastic stress analysis was conducted using the FEM.

In addition, experimental measurements were conducted using the experimental set-up shown in Figure 82. After the completion of welding, contact balls were attached on both sides along the weld line at the gage length of 20 mm on the top and bottom surfaces as shown in Figure 82. Restraint stresses were released by cutting the weld metal along the weld line, and the displacement produced along the weld line was measured. The average displacement values measured on both surfaces are indicated by the bullet for one side of the weld along with the FEA results from different models in Figure 83.

Additionally, two SR-4 strain gages were attached at right angle on the top and bottom surfaces of the plate as shown in Figure 82, such that residual stresses could be measured as average values on the two surfaces by the stress relaxation method. Measured residual stresses along the line (y = 13.5 mm), parallel to the weld line longitudinal stresses, (σ_x), and those perpendicular to the weld line transverse stresses, (σ_y), are indicated in Figure 84 along with the numerical results. Similarly, residual stresses along the y-axis (transverse direction) at x = 5 mm, σ_x and σ_y , are shown in Figure 85. The transverse residual stress σ_y distribution in the weld metal along the weld length using the numerical analysis (using model M3) and an analytical method are compared in Figure 86.

It is clear from these Figures that models M1, M2 and M3 closely reproduce the experimental results. Specifically model M2 (incorporating yield stress hysteresis, transformation expansion and transformation plasticity) and its simplified version agree well with the experimental results. It was observed in Ueda's¹⁵⁸ study on rigid restraint cracking (RRC) test that neglecting these effects resulted in somewhat higher residual stresses but the overall stress distribution was similar.

In the present study, the effect transformation plasticity was neglected. This suggests that the residual stresses will be somewhat higher than those expected. But it has also been observed by other researchers^{160, 161} that neglecting transformation effects leads to an increase in the magnitude of residual stresses especially if the phase transformation is occurring at low temperatures. The phase transformation temperature for the A514 steel used is about 600–500°C, therefore, the effect of neglecting the phase transformation effect will be less severe. Hence, a comparison with Ueda's experimental results will yield information about the validity of numerical results in this study.

Transverse Displacement

The transverse displacement, UX, along the weld length at the plate top surface groove edge is shown in Figure 87. It is observed that the Tekken test specimen undergoes maximum displacement near the middle portion of the groove. The displacement decreases parabolically as the groove ends are approached. The UX distribution is exactly opposite to the restraint intensity variation. When the transverse displacement is compared with that determined by Ueda in Figure 83, it is observed that the overall distribution is almost identical.

The individual values are not similar since the assumptions and welding conditions are different. The transverse displacement distribution suggests that the middle region of the weld is more compliant (less stiff or restrained), and therefore, offers less resistance to the shrinkage of the weld metal. Whereas, at the weld ends, the transverse displacement is smallest since the restraint is maximum here. Therefore, opposition to the weld shrinkage is maximum and this results in higher transverse stresses at the weld ends. The higher resistance to transverse displacement at weld ends is because of the additional mass in the longitudinal direction at the groove ends. This additional mass of material resists the transverse displacement and contributes to non-uniform restraint as well as stress distribution along the weld length.

The overall residual plate deformation is shown in Figure 88. It is clear from this Figure that the Tekken test undergoes bending type of deformation which also affects the ensuing residual stress deformation. It should be noted that overall deformation is magnified by a factor of about 160 for clarity. The plate deformation is pivoted around one restrained edge. This restrained edge was used to prevent rigid body motion during the numerical simulation. In reality,
the specimen is allowed to displace freely. Therefore, the displacement pivoting around the restrained end will not be observed in Tekken test experiments. It should be noted that the use of restrained edge does not affect the stress development and final residual stress value near the weld region as the restrained edge is far away from the weld.

Stress Distribution Perpendicular to the Weld Length

The residual stress distribution in the direction transverse to the weld centerline at different locations is shown in Figures 89 and 90. At the mid-thickness of the specimen, Y=.0125 m, Figure 89, it is observed that the longitudinal stress is lower at the weld centerline but increases and has a peak distribution at 3 mm away from the centerline which corresponds to the weld metal and HAZ interface. The longitudinal stress then steadily decreases as one moves away from the HAZ and then becomes compressive and is constant. The equivalent stress shows a similar distribution but is always tensile in nature. The transverse stress (SX) first increases and the decreases rapidly as the HAZ is approached. Farther away from the HAZ and base metal interface, the SX distribution is constant and tensile.

The residual stress distribution in transverse direction at the plate top surface is shown in Figure 90 and is significantly different than that at the plate mid-thickness, Figure 89. The overall stress distribution magnitude is considerably smaller than at the plate mid-thickness. The stress distribution in the transverse direction was compared with Ueda's results, Figure 85. It is observed from Figures 89 and 90 that the stress distribution in the present study is considerably smaller than Ueda's result. Figure 90 shows the stress distribution on the top surface of the plate, and therefore, does not include the stress distribution in the immediate vicinity of the weld because of the 30° groove angle. Whereas, Ueda's results include stress distribution in the weld region as seen from the transverse dimensions in Figure 85, even though, residual stress measurements were carried out on the plate top surface. If Ueda's results in Figure 85 are compared with the transverse stress distribution at the mid-thickness of the plate, Figure 89, then it is observed

that the stress distribution is similar except for the longitudinal stress near the weld centerline. This difference near the weld centerline is because the phase transformation effects, such as, transformation plasticity, were neglected in the present study. This was also observed by Goldak et al¹⁶², for a 522 mm long bead on plate. They observed that completely neglecting phase transformation effects led to yield-point magnitude, tensile, longitudinal stresses near the weld, Figure 91. Inclusion of only transformation volume change created large compressive stresses near the weld, Figure 91. Inclusion of transformation plasticity as well as the transformation volume change resulted in longitudinal stresses near the weld of low magnitude but tensile in nature, with maximum longitudinal stress located just outside the HAZ, Figure 91. This last stress distribution in fact matches with that found in the present study, Figures 89 and 91. The phase transformation effects are important in Ueda's study because the transformation start temperature is about 425°C and the transformation finish temperature is about 250°C. Whereas in the present study, the transformation start temperature is about 500 to 600°C depending on the weld cooling rate and transformation finish temperature is at about 300 - 350°C. Therefore, the phase transformation effects are less significant for the steel used in this study. Nevertheless, more accurate residual stress distribution can only be obtained using correct physical as well as mechanical properties.

Stress Distribution Along the Weld Length

The residual stress distribution in the longitudinal direction (along the weld length) at the plate top surface is shown in Figure 93, 94. In Figure 93, the stress variation along the top groove edge indicates marked variation in stress distribution as well as the magnitude along the weld length. The transverse stress (SX) and equivalent stress (SIGE) are maximum near the groove ends. In contrast, the longitudinal stress is lowest and compressive at the groove ends and steadily increases with a peak value at the plate mid-length. Farther away from the top groove edge, the overall stress magnitude decreases and also changes sign, Figure 94. When compared

with Ueda's results, Figure 84, it is clear that the variation in transverse and longitudinal stress distribution along the weld length matches quite well, Figure 93. The stress magnitude is some-what lower compared to Ueda's results. The longitudinal stress (SZ) in this study is compressive at the groove ends and steadily increased to tensile in the middle region of the weld, Figure 93. Whereas, the transverse stress (SX) is tensile along the weld length in contrast to Ueda's results. This difference in longitudinal and transverse stress distribution at the groove ends is probably because of the weld configuration differences in the present study and Ueda's model. In the present study, the total groove length is 80 mm and is similar to that used by Ueda, but, the total weld length is 76 mm to simulate the 2 mm gap at the weld ends as per the original configuration, Figure 95. In contrast, in Ueda's study, the weld length was 80 mm. This 2 mm gap at the weld ends allows free shrinkage at least in the longitudinal direction. This is particularly true for the weld ends. As a result, the restraint on the groove ends is lower and may lead to compressive stresses in the present study. But the overall stress distribution along the weld length matches quite well with that in Ueda's model.

Additionally, the welding transverse residual stress along the weld length was also estimated by Ueda for the weld metal region, Figure 86, using an analytical as well as FEA approach. Similar transverse stress distribution in the weld metal at the maximum stress location obtained in the present study is shown in Figure 92. It is clear from these Figures that the stress distribution is quite similar to that by Ueda's FEA. This also gives additional confidence in the numerical stress analysis results of the present study.

It is clear from the above discussion that the stress distribution varies significantly through the plate thickness. Surface residual stress measurement, therefore, is not a true indicator of internal residual stresses. The stress distribution comparison with Ueda's results indicates that the numerical model used in the present study simulates the stress distribution fairly accurately. The stress magnitude is different because of the material as well as phase transformation considerations. The phase transformation effects will be less pronounced in this study because of the higher phase transformation temperature, and hence the results in the present study can be used to analyze the transient as well as the residual stress distribution in the Tekken test weld zone.

LOCALIZED STRESS DISTRIBUTION

The residual stress distribution in the longitudinal as well as transverse direction was analyzed in detail at various locations of interest in the weld region. The iso-stress contours describing the residual stress distribution along the weld length are shown in Figures 96-98. Transverse stress (SX) distribution shows a maximum in the center of the weld metal instead of the weld root notch location, Figure 96. In fact, at the weld root notch along the weld length, transverse residual stresses, SX, are compressive near the groove ends and change to tensile after about 15 mm from the groove ends, Figure 99. The compressive SX distribution at the weld root notch is some what unexpected. It suggests that, near the weld end locations crack initiation is next to impossible as transverse stresses are compressive. Whereas, tensile transverse stresses are necessary to initiate cracking⁴⁸. This type of SX distribution is only typical of the test weld configuration used in the experimental study. The influence of test weld configuration on residual stress distribution will be analyzed next.

The JIS Z-3118 standard recommends two test weld configurations depending on the welding process employed for test welding, Figure 95. For the automatic welding used in this study, the test weld configuration was chosen such that there was a 2 mm gap on the weld ends. To simulate this test weld configuration, the FE model used in this study also incorporated this gap. For ease of simulation and meshing, the actual gap in the numerical model was about 2.5 mm. Because of this 2.5 mm gap at the weld ends, the restraint on the groove ends is greatly reduced. This is because the molten weld pool at the weld ends is free to shrink in the longitudinal direction. As a result, the transverse stresses near weld ends are compressive. If a different test weld configuration, Figure 95 (a), is selected such that there is no gap at the test weld ends, then the transverse stress distribution will be tensile because of the restraint from the groove ends. This is is an important result as the type of test weld configuration will affect the transverse stress distribution at the HAC initiation location, and therefore, the cracking or no cracking condition. The test weld configuration with a 2.5 mm gap at the weld ends will show lesser tendency to cracking at the groove ends because of the compressive transverse stresses at the weld root notch. Whereas, in the test weld configuration without the gap at the weld ends the transverse stress distribution will be tensile at the weld notch root (HAC initiation) location. Therefore, caution needs to be exercised when comparing Tekken test results with different test weld configurations.

The test weld stress distribution effect will also be in addition to the extra heat input in the configuration without the gap at the weld ends, Figure 95 (a). To avoid the crater crack formation in manual welding, the weld is started on the top of the plated surface and then is moved laterally into the test weld location. At the groove ends, the weld is terminated by laterally moving the arc to the plate top surface. This practice results in additional overall heat input which lowers the weld cooling rates thereby promoting less susceptible microstructure formation and hydrogen outgassing out of the specimen. The overall effect of this practice is then to lower the susceptibility to HAC for a given welding condition. Similar observations hold true for the Tekken test weld practice employed by a few other researchers¹⁶³. In this practice, the restraint welds in the vicinity of the test groove in the Tekken test are subjected to grinding. The restraint welds near the groove ends are ground to a certain depth. This ground restraint weld is then used to start and end the test weld to avoid crater formation. As a result, the test weld length is longer than 80 mm. As discussed in the previous section, this practice results in additional heat input which alters the cooling characteristics of the specimen particularly the groove ends. The transient as well as the residual stress development is also influenced by this type of temperature distribution. The above discussion suggests the influence of test welding practice on stress, temperature distribution and hence on HAC. Therefore, good correlation among different test weld practices will be difficult to obtain.

The Transverse (SX) distribution at the weld root notch suggests that the transverse tensile stresses do not play a significant part in the crack initiation process at the weld root notch in Tekken specimens, since the magnitude of tensile stresses is 20 MPa at the maximum SX location, Figure 99 and is lower than at 20 MPa at other locations.

The SX-stress distribution at the maximum SX location along the weld length is shown in Figure 92. It is observed from Figure 92 that SX is maximum at about 15 mm from the weld start and end locations and decreases in the middle length of the specimen. The maximum SX distribution along the weld length increases continually and peaks at about 2.5 mm away from the weld end. This type of stress distribution is similar to the restraint intensity distribution in Tekken test specimen^{158, 159} and arises primarily from two factors. The restraint is maximum at the weld ends and gradually decreases in a parabolic manner in the middle region. As a result, transverse shrinkage of the weld is not uniform along the weld length. The resistance to transverse shrinkage being the maximum at the weld ends and least in the middle. The second reason which augments this type of SX distribution is the non-uniform cooling experienced by the specimen due to the slot nature of the test. Because of the dimensionality differences along the weld length, the specimen weld ends cool at a faster rate than in the middle, Figure 69 (a) and (b). As the weld temperature decreases the yield strength starts increasing. This suggests that when the yield strength starts to recover with a decrease in temperature the weld ends already regain their strength. Whereas, the weld middle region is at higher temperatures, and therefore, has a lower yield strength compared to the weld ends. As a result, the developing transient stresses are absorbed by plastic straining in the middle region, since the shrinkage stresses probably exceed the yield stress at these temperatures. In other words, the middle region is less restrained and readily absorbs the transient stresses by yielding thus avoiding higher stress build up in this region. The latter explanation suggests that middle region should exhibit higher plastic strains than the weld ends. This was also observed by Ueda45, 154 after conducting FEA on Tekken test stress distribution after welding. Whereas in this study, the equivalent residual stress

distribution is highest at the weld ends, suggesting that maximum plastic strains are at the weld ends. However, it should be realized that these are the residual stresses, and therefore, the transient stresses need to be examined. This involves inherent errors as the material properties in these temperature ranges are not accurately known.

The longitudinal stress (SZ) distribution shows that a maximum occurs at the interface between the weld pool and HAZ, Figures 89 and 97. This is similar to the trend reported in the literature^{123, 162}. The SZ-stress distribution at different locations is almost uniform along the weld length, except that the SZ-stress magnitude is somewhat lower in the mid-length region compared to weld start and end locations, Figure 93,94 and 99. It is also observed that SZ-stress distribution near the fusion line is tensile and is almost equal to the yield stress of the base metal at RT, Figure 99. The longitudinal residual stress in a weld is always as high as the yield strength level, independent of restraint intensity⁴⁸. Therefore, the observed SZ distribution matches to that reported in the literature. The longitudinal residual stress does not influence HAC. In contrast, it has been reported by Yurioka and Suzuki⁴⁸ that the transverse stress causes root and toe cracking and is dependent on the transverse restraint intensity.

The longitudinal stress distribution in the transverse direction at mid-plate thickness exhibits a maximum at the fusion line, Figure 89. In comparison, the longitudinal stress distribution at the plate top surface rapidly changes from tensile at the groove edge to compressive as one moves away from the groove edge, Figure 90. In fact, at 10 mm from the weld centerline the SZ stress is compressive at the plate top surface, Figure 90, whereas, at the mid-thickness it is tensile, Figure 89. This indicates marked variation in residual stress distribution across the specimen thickness. Hence, residual stress measurement at the plate surface may not be a good indicator of internal residual stress distribution.

The von-Mises equivalent (SIGE) stress distribution at the weld notch root on the double bevel side along the weld length is shown in Figure 99. It is observed that the residual equivalent stress exceeds the yield strength of the material at RT near the weld start and end locations. This suggests that the weld notch root region location undergoes plastic straining. The SIGE stress distribution is slightly lower than the yield strength in the middle of the weld indicating that plastic straining does not occur in this region. The SIGE iso-stress contours are shown in Figure 98. Maximum SIGE stress distribution is observed around the weld toe and root notches. It is observed from Figure 98 and 99 that a small area around the weld notch root location undergoes plastic straining near the weld ends. This plastic region also includes localized region of the HAZ. This agrees well with the experimental Tekken specimens. It was observed after fractography on cracked Tekken specimens that at the weld root notch (crack initiation) location, the fracture mode was micro-void coalescence indicating occurrence of plastic straining. This additionally provides further validation of the numerical results.

In the Tekken specimens used in the present study, MVC was observed along the entire weld length. Whereas, according to the numerical results plastic strains are observed only 12 mm into the weld from the weld ends. There could be several reasons to this anomaly. The plastic strains observed in the actual Tekken specimen could have developed at high temperatures. Whereas, only residual stress distribution from numerical simulation is being analyzed here. In other words, at higher temperatures localized yielding will occur if the instantaneous equivalent stress exceed the instantaneous yield strength, but, as the weld cools this no longer may be the case. Hence, a definite statement about the high temperature strains and stresses can not be made from RT residual stress distribution.

This problem is again compounded by the fact that material properties are not accurately defined at higher temperatures. In addition, it was assumed that the weld geometry is uniform along the weld length, whereas in actual specimens the weld root notch has significant variation due to varying depth of penetration. This results in varying stress concentration which affects the plastic strains at these locations. This will give rise to the observed trend of yielding along most of the weld length at these locations. In experimental Tekken specimens, the weld root notch is much sharper than that used in the numerical study. Therefore, plastic straining in actual

specimen will be observed to a greater extent than in the numerical study. Nevertheless, it is observed from this numerical study that definite statements about the stress distribution can be made using a simplified numerical procedure.

STRESS DISTRIBUTION AT CRACK INITIATION LOCATION

From the equivalent stress distribution and the transverse stress distribution, the crack initiation in Tekken specimens can be analyzed in detail. The presence of hydrogen facilitates the failure process by promoting whatever failure mechanism is operative as per Beachem's hydrogen embrittlement theory⁹³. At the end of welding, depending on the hydrogen concentration, HAC initiates from the weld root notch location due to the plastic zone and presence of hydrogen. As soon as a micro-crack is initiated, the residual stress is redistributed. The plastic region size in the actual Tekken specimens is quite small, i.e. ranges from 25 to 130 microns. Whereas, in the numerical solution the plastic region is substantially larger, especially near the weld ends. This difference could be because of two reasons. First the material properties used in the numerical simulation are somewhat different than the actual material properties. In the numerical model, the same material properties were used for HAZ, weld metal as well as the base metal. In reality, the coarse grained HAZ is much harder and has higher yield strength than the base metal as well as the weld metal. For example, the CGHAZ hardness was around 420 VHN which translates to tensile strength of 192 ksi (1323 MPa) or a yield strength of approximately 175 ksi (1200 MPa). If we consider this value of yield strength then the plastic zone size near the weld notch root becomes very small in the numerical model, especially at the weld ends. Similarly, the plastic zone in middle region becomes almost negligible. With different material properties in the weld zone, the residual stress development will be also slightly different compared to the case of uniform properties in the weld zone. In addition, as soon as the developed plastic zone becomes of a critical size crack initiation occurs in actual specimens. This prevents stress build up in actual experiments. In comparison, in a numerical scheme the residual stress build up will continue to

its maximum value at RT.

The equivalent stress distribution from the numerical simulation is quite peculiar in the sense that most of the plastic or maximum SIGE region is located on the HAZ side of the fusion line at the weld notch root location, Figure 98. The microcrack then should preferably occur in the HAZ region. This type of SIGE stress distribution can be attributed to the oblique groove shape. Since the HAZ region is more susceptible to HAC compared to the weld metal, crack propagation along the fusion line in the HAZ is preferred. Further crack propagation occurs in the HAZ before either turning into the weld metal or propagating all the way in the HAZ depending on the local HAC conditions and stress intensity factors as discussed in the section related to weldability testing.

Presence of both near yield equivalent and transverse tensile stresses is required for HAC. This suggests, according to the Tekken test numerical results, that HAC may not occur near the weld ends because the transverse tensile stress is compressive. Whereas, in the weld middle region the equivalent and transverse stress combination is just right, and hence, cracking should be preferably begin in the middle part of the weld. This mechanism is well suited for the test weld configuration used in the present study. This mechanism is somewhat different than that proposed by Ueda et al^{45, 159}. According to them, extent of HAC is higher in the middle region of the Tekken test specimen because of the higher plastic strains as observed from the numerical results. It should be noted that these authors did not mention the Tekken exact weld location where they observed these plastic strains. In fact, most of the results were quoted from the plate top surface. In addition, the groove geometry was assumed to be square with the end result that localized stress distribution near the crack initiation location was not analyzed in detail. Therefore, it is difficult to correlate their surface stress results with the observed trend of higher cracking extent in the middle region of the Tekken test.

EFFECT OF PREHEATING TEMPERATURE

With an increase in T_{ph} the overall stress magnitude decreased, Figures 99, 100. This decrease was not substantial when compared with the increase in the magnitude of T_{ph} . Figure 99 and 100 show that plastically strained region decreases with an increase in T_{ph} . The overall nature of residual stress distribution is unchanged with an increase in T_{ph} .

This suggests that preheating decreases the HAC susceptibility of a given material because of a decrease in the transient and residual stress state as well as a pronounced decrease in weld cooling rates. The decrease in stress state occurs due to a decrease in the thermal stress caused by the differences in the base metal temperature and the molten metal. As per the mechanism of thermal stress development described in the background section, the temperature difference between the initial base metal temperature (strain free temperature) and the molten metal (current temperature) defines the thermal strain. With higher initial T_{ph}, the thermal strain is lower, and thereby the overall residual stress magnitude is reduced. This decrease in residual stress magnitude with an increase in temperature is not significant enough to reduce the HAC susceptibility drastically, Figures 99 and 100. But the beneficial effect of preheating is more pronounced in 1500 to 100°C weld cooling rates, Table 39. This suggests that reduction in the residual diffusible hydrogen at lower temperature is the predominantly beneficial effect of preheating. An increase in T_{ph} also decreases the 800 to 500°c cooling rate, Table 39. This effect is more pronounced in the present study because of the excessively high latent heat of transformation used. As is clear from the 3-D analytical equation results, Table 39, the effect of T_{ph} on 800 to 500°C cooling time is still significant even if the latent heat of transformation is neglected. Therefore, the beneficial effect of T_{ph} on HAC susceptibility can be summed up as follows in order of expected effect on decreasing HAC.

 Pronounced decrease in 1500 to 100°C cooling rates which reduces the residual diffusible hydrogen content.

- Decrease in 800 to 500°C weld cooling rates which promotes the formation of less susceptible HAZ microstructures.
- 3 Decreased transients as well as residual stress magnitude.

RESIDUAL STRESS DISTRIBUTION IN LEHIGH SPECIMENS

The overall residual stress distribution trend was in accordance to that observed for the Tekken specimen. There were important differences in the residual stress state near the weld zone compared to the Tekken tests. Some the most significant differences are discussed in this section.

The residual stress distribution in the weld zone is shown in Figures 101, 102 and 103. The residual transverse stress (SX) distribution in Figure 101 shows compressive stresses at the weld root notch, which is also the crack initiation site in Lehigh specimens, Figure 32. With this type of compressive transverse residual stress distribution, crack initiation at this location not possible. In addition, the maximum SX location is on the groove face which is an anomalous result. The presence of compressive SX stresses can be explained on the basis of plate bending as a function of weld bead eccentricity from the neutral axis of the plate⁸⁶. According to the FEM results by Suzuki, Figure 43, as the weld bead moves away from the neutral axis of the plate in the downward direction, the transverse stresses at the weld root notch become compressive. The results in this study are then in agreement with those by Suzuki⁸⁶. This result is very important for understanding the weldability test methodology and its application to practical joints. The Lehigh test groove is shaped such that the weld root notch will always be below the mid-thickness (neutral axis) of the test specimen. Depending on the degree of this eccentricity, the transverse stresses at the weld root notch will vary from tensile to compressive. The exact location of this transition from tensile to compressive residual transverse stress distribution is unknown, but can be determined. This type of residual stress variation due to the weld bead eccentricity contributes to the experimental scatter between results from different sources.

In this context, it is interesting to note that the Lehigh specimens in the experimental study exhibited MVC fracture mode at the weld root notch. The equivalent stress (SIGE) distribution in Figure 103 does not exhibit any peak distribution at the weld root notch, instead a peak SIGE distribution is observed at the weld toe. The SIGE value at the weld root notch is substantially lower than the yield strength to cause any yielding.

This discrepancy in the experimental and numerical results can be explained on the basis of degree of root penetration in actual specimens. It was observed in Figures 24(a), (b) and 40(a), (b) that the root penetration is Lehigh specimens was non-uniform along the weld length. Near the weld start location, the root penetration is lowest, and therefore, the weld bead shows least eccentricity compared to the weld end locations which show higher eccentricity. As a result, tensile transverse residual stresses will exist at the weld root notch locations near the weld start location. In addition, the equivalent stress will also exhibit a peak distribution at these locations. In contrast, the weld root notch locations near the weld ends will exhibit compressive stresses which also lower the SIGE peak distribution. Therefore, HAC initiation tendency will vary along the weld length, being highest at the weld start and decreasing steadily towards the weld ends. This effect will be augmented due to faster weld cooling rates at the weld start locations, as it affects microstructure and H_r content, and by the stress concentration variation at the weld root notch due to the degree of penetration. This argument agrees well with the observed physical crack propagation in Lehigh specimens which was always from the weld start to weld end location. In the numerical simulation, only a single uniform value of root penetration was used along the weld length. The extent of root penetration in the numerical scheme was almost close to full penetration, Figure 63(b), and hence, resulted in compressive SX distribution at the weld root notch location. It should also be noted that weld bead eccentricity in Tekken specimens was lower than the Lehigh specimens.

It was also observed that the transverse residual stress, SX, magnitude in Lehigh specimen weld zone is almost one half of that in the Tekken Specimens, Figures 96 and 101. This directly

follows from the fact that Tekken test is a very severe restraint test and has almost twice the restraint intensity as compared to the Lehigh specimen, Table 20. The mean stress acting on the weld metal is a strong function of R_F , in accordance with equation 1. This also suggests that Tekken specimen is much more severe in assessing HAC susceptibility compared to the Lehigh specimen.

APPLICABILITY OF WELDABILITY TESTS FOR HAC

SUSCEPTIBILITY ASSESSMENT

The highly dimensional nature of heat transfer along the weld length suggests that HAC susceptibility will also vary along the weld length. For example, The weld start and weld end experience lowest $t_{8/5}$ compared to the middle region. As a result, weld end region HAZ will exhibit more susceptible microstructure compared to the middle region HAZ. In addition to the microstructure variation along the weld length, H_{r100} content will also vary along the weld length due to faster $t_{15/1}$ at the weld ends compared to the weld middle region. This is only true for weldability test specimens welded at RT, because with increasing T_{ph} , $t_{15/1}$ is almost constant along the weld length.

The residual stress distribution along the weld length suggests that higher plastic straining is observed at weld ends compared to the weld middle region. This suggests that weld ends should exhibit highest HAC susceptibility as plastic strains at the weld start facilitate hydrogen trapping, and hence, the crack initiation process. All these factors discussed above suggest that weld ends should exhibit higher HAC cracking tendency than the middle region. However, practical experience suggests that cracking extent is higher in the middle regions compared to the weld ends^{67, 68, 159, 164}. There could be various explanations for this type of behavior. For the test weld configuration with 2 mm gap at the weld ends, it was observed that compressive transverse residual stress exist at the crack initiation location, i.e., weld root notch on the double side, near the weld ends. Another reason could be the non-uniform hydrogen distribution during welding.

Evans³² observed that because of the weld end perturbations, initial hydrogen content at the weld ends will be significantly lower than the middle region. Therefore, this region will exhibit a lower HAC susceptibility than the middle region. A third reason has been suggested by Ueda et al¹⁶⁵. These authors observed from FEA of Tekken test that plastic strains were higher in the middle portion and vanished near the weld ends. The plastic strain distribution is similar to the cracking tendency. The exact weld location where they analyzed plastic strains is not clear from their study. This observation, however, does not agree with residual stress distribution observed in the present numerical study.

Nevertheless, it is clear from the above discussion that initial as well as residual hydrogen, microstructure and residual distribution varies substantially along the weld length. This suggests that the HAC susceptibility will also vary along the weld length in the weldability tests. Therefore, the slot weldability tests are not suitable for assessing HAC susceptibility of steels in service. The severe restraint nature of the Tekken test along with weld end effects suggest that applicability of Tekken test results to practical weld joints is doubtful as Tekken weldability test results are highly conservative. Same observations apply to the Lehigh weldability test, except for the fact that this is less stringent test. The above observations from this study suggest that slot weldability tests should be redesigned to determine safe welding conditions for weld joints used in practice.

4.4 - CONCLUSIONS

- The simplified thermal analysis procedure can be successfully used used to understand the heat transfer in weldability specimens.
- 2) The heat distribution in Tekken and Lehigh specimens was highly dimensional due to the slot nature of the test weld. As a result, the weld ends cool at a much faster rate compared to the middle region.
- 3) The cooling rate variation along the weld length will affect microstructure development and the $\frac{H_r}{H_o}$ distribution. Therefore, the HAC susceptibility of a weldability test will vary along the weld length.
- Increasing T_{ph} reduces the HAC susceptibility primarily through the reduction in H_r content and by promoting the softer microstructure formation, and secondarily by reducing the residual stress level.
- The stress analysis procedure provided results that were in good qualitative agreement with published results.
- 6) The stress distribution in Tekken specimens varied strongly along the weld length due to the slot nature of the test and non uniform restraint conditions along the weld length.
- The stress distribution in Tekken specimens agreed well with experimentally observed fracture modes.
- 8) The Lehigh test transverse residual stress level was significantly lower than that in Tekken specimens. This is in agreement with the fact that the restraint intensity in Lehigh specimens is almost half that of the Tekken test.
- 9) The test bead eccentricity influences plate bending, and hence, the stress state at the weld root notch. This strongly affects HAC susceptibility due to development of compressive stresses.



Figure 61.

- (a) Dimensions of Tekken Specimen in mm,
 (b) Finite Element Model A for Thermal and Stress Analysis,
 (c) Finite Element Model B for Thermal and Stress Analysis, From Andersson (115)



Figure 62 (a) Finite Element Mesh for the Tekken Specimen













(b) Finite Element Mesh for the Lehigh Specimen - Cross Section







Figure 65 Gaussian Temperature Distribution in the Tekken Test Weld Pool



Figure 66 Temperature Dependent Thermo-physical Properties used in this Study from Patel (147)





Restraint and Boundary Conditions for Stress Analysis of Tekken Specimen



Figure 68 Temperature Dependent Mechanical Properties used in this Study from Department of Defense, (156)



Figure 68 Continued



(a)



Figure 69 : Weld Thermal Cycles for the Tekken Specimen (a) Room Temperature, (b) 175 C



Figure 70 Weld Thermal Cycle for AWS A4.3 size Grooved Specimen at RT - 1.55 kJ/mm



Figure 71 Temperature Distribution at Time = 0.533 Seconds, Time Step 1 Tekken Test at RT







Figure 73 Temperature Distribution at Time = 7.995 Seconds, Time Step 15 Tekken Test at RT



Figure 74 Temperature Distribution at Time = 14.7 Seconds, Time Step 27 Tekken Test at RT











Figure 77 Temperature Isotherms at Time = 0.533 Seconds, Time Step 1 Tekken Test at RT, 5 mm from Weld Start



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Figure 79 Temperature Isotherms at Time = 14.7 Seconds, Time Step 27 Tekken Test at RT, 5 mm from Weld End


Figure 80 Temperature Distribution at Time = 24 Seconds, Time Step 49 Lehigh Test at RT







Figure 82 Slit Weld Specimen Used by Ueda (158)



Figure 83 Inherent Displacement along the Weld Length, From Ueda (158)



Figure 84 Longitudinal and Transverse Residual Stresses along the Weld Length of the Slit Specimen, From Ueda (158)



Figure 85 Longitudinal and Transverse Residual Stresses in the Transverse Direction of the Slit Specimen, From Ueda (158)



Figure 86 Transverse Residual Stress in the Weld Metal along the Weld Length, From Ueda (158)



Figure 87 : Transverse Displacement (UX) Distribution along the Weld Length (Top Surface Groove Edge) - Tekken Specimen Welded at RT



Figure 88 Residual Deformation of the Tekken Test Welded at RT, Magnification = x160



Figure 89 : Residual Stress Distribution along the Plate Width from Weld Centerline at mid-thickness, Tekken Specimen Welded at RT



Figure 90 Residual Stress Distribution in Transverse Direction, from the Groove Edge at the Top Surface of the Tekken Specimen Welded at RT (mid-length)



Figure 91 Residual Longitudinal Stress on the Top Surface vs. Transverse Distance from the Weld Centerline, from Goldak (162)



Figure 92 Residual Stress Distribution along the Weld Length at the Maximum SX Location, Tekken Specimen Welded at RT (Mid-thickness)



Figure 93 Residual Stress Distribution along the Weld Length from the Top Surface Groove Edge, Tekken Specimen at RT



Figure 94 Residual Stress Distribution along the Weld Length, 5 mm from the Top Surface Groove Edge, Tekken Specimen Welded at RT



Fig. 3

Unit: mm



Figure 95 Tekken Test Weld Configurations - JIS 3158

(a) Manual Welding, (b) Mechanical Welding



Figure 96 Residual Transverse Stress (SX) Distribution at the Plate mid-length Tekken Specimen Welded at RT - Close-Up



Figure 97 Residual Longitudinal Stress (SZ) Distribution at the Plate mid-length Tekken Specimen Welded at RT - Close-Up



Figure 98 Residual Equivalent Stress (SIGE) Distribution at the Plate mid-length Tekken Specimen Welded at RT - Close-Up



Figure 99 Residual Stress Distribution along the Weld Length, at the Weld Root Notch, Tekken Specimen Welded at RT



Figure 100 : Residual Stress Distribution along the Weld Length, at the Weld Root Notch, Tekken Specimen Welded at 175 C



Figure 101 Residual Transverse Stress (SX) Distribution at the Plate mid-length Lehigh Specimen Welded at RT - Close-Up



Figure 102 Residual Longitudinal Stress (SZ) Distribution at the Plate mid-length Lehigh Specimen Welded at RT - Close-Up



Figure 103 Residual Equivalent Stress (SIGE) Distribution at the Plate mid-length Lehigh Specimen Welded at RT - Close-Up

Preheat Temperature	Analytical		Numerical					
	t _{8/5} Seconds	t _{15/1} Seconds	t _{8/5} Seconds			t _{15/1} Seconds		
			Weld Start	Quarter Length	Middle Region	Weld Start	Quarter Length	Middle Region
RT	7	70	4	7.5	11	97	120	129
130°C	11.5	1300	7	11	15	1800	1800	1800
175°C	15	1800	8.5	14	18	2350	2350	2350

Table 39 : Comparison between Numerical and Analytical Weld Cooling Times

Table 40.Idealization of Material Properties of HT-80 and Modelsfor Finite Element Analysis, From Ueda (158)



CHAPTER 5

SUMMARY

A comprehensive study of the existing weldability testing methodology for HAC susceptibility assessment was performed. Each stage of weldability testing procedure was analyzed in detail using experimental and numerical approaches.

In chapter 1, the existing AWS A4.3-86 weld hydrogen measurement procedure was analyzed from weldability assessment viewpoint. The results demonstrated that weld hydrogen levels in actual weldability tests will be significantly different than that determined from the existing weld hydrogen measurement standard, such as, AWS A4.3-86. This is because the standard weld hydrogen measurement technique was developed for classification of consumables, and, hence, does not consider the effect of prior degassing treatment and base metal composition on final weld hydrogen content. It was also observed that groove shape and weld length differences contribute to a high degree of scatter in weld hydrogen levels. This suggests that weld hydrogen content in weldability tests as well as actual weldments will be different than that determined by weld hydrogen measurement standards. It is therefore suggested that the existing weld hydrogen measurement should be modified for use in weldability testing procedure.

The experimental results on Lehigh and Tekken tests suggested these tests can be successfully used to assess weld metal and HAZ susceptibility provided that welding process variation does not alter the stress concentration at the weld root notch. The root penetration varied significantly along the weld length, and hence influenced affects residual stress level as well as crack propagation. Different T_{ph} prediction methods were analyzed and it was observed that none of them provides safe and optimum T_{ph} at all the weld hydrogen levels. Yurioka's method is the most comprehensive of all these methods but generally predicts too conservative preheating temperatures. The applicability of weldability test results to practical weldments is still not well understood, but, Yurioka's method is the right step in this direction. The numerical simulation of hydrogen diffusion behavior indicated that the residual diffusible hydrogen content of weldability tests will be significantly different than that in the commonly used practical joints. The influence of groove shape and thermal cycle on residual diffusible hydrogen content again supports the view that Lehigh and Tekken weldability tests were not designed to simulate HAC response of actual weldments.

The numerical heat transfer analysis indicated that the heat transfer was highly dimensional and the weld ends cooled at a much higher rate compared to the middle region in both Lehigh and Tekken weldability tests. It was observed that the residual stress magnitude varied substantially along the weld length due to the slot nature of the tests and non uniform restraint conditions along the weld lengths. This results in HAC susceptibility variation along the weld length. The stress distribution at the crack initiation location, i.e., the weld root notch, agreed well with experimentally observed fracture modes. The groove shape affected the stress distribution at this location, and therefore, the HAC tendency. These results reiterate the hypotheses that weldability test results can not be applied to many practical weldments.

It is concluded that the existing weldability testing methodology needs to be changed for better applicability to industrial fabrication practice since it is not properly designed to address the effect of different variables. Some of these variables can not be avoided because the existing testing techniques were not specifically developed for HAC susceptibility assessment, but were adopted. The weldability tests should, therefore, be redesigned to simulate weld hydrogen levels, welding process variables, hydrogen diffusion, heat transfer and residual stress levels of actual weldments. In this author's opinion, incorporating these suggestions will lead to a better control over HAC occurrence and prevention in industrial fabrication.

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