STUDIES ON HYDROGEN ASSISTED CRACKING SUSCEPTIBILITY OF 2.25Cr-1Mo STEEL WELDMENTS

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ABSTRACT

Hydrogen Assisted Cracking Susceptibility of 2.25Cr-1Mo steel was studied using UT-Modified Hydrogen Sensitivity Test (UT-Mod. HST). Welding was carried out using GTAW process and hydrogen was introduced into the weld metal by mixing it with argon shielding gas. Autogenous bead-on-plates were made on specimens fixed on a copper fixture and the specimens were strained for 4% strain immediately after welding for 24 hrs and checked for cracks. Preheating during welding was achieved by heating the copper fixture and the minimum preheat temperature required to prevent cracking for known vol.% of hydrogen in the shielding gas found out. Diffusible hydrogen content (H_D) in the specimen prepared under identical conditions employed for testing was measured using Gas Chromatograph. Weld cooling curves were obtained by plunging a thermocouple into the molten weld pool and cooling time ($t_{8/5}$) was estimated from the curves. Variation in weld metal hardness and microstructure was also studied. The results showed that both minimum preheat temperature and (H_D) content increased with hydrogen in the shielding gas. Preheating results in reduction in (H_D) content but does not affect the hardness indicating that cracking is prevented due to reduction in (H_D) content in the weld prepared with preheating. The results show that it is possible to choose an optimum preheat temperature to prevent cracking if (H_D) level in the weld is known.

INTRODUCTION

Cr-Mo steels are known for their high temperature strength and oxidation resistance and are widely used in power and petrochemical industries for the fabrication of steam generators and heat exchangers. The high alloy content of the steels makes them susceptible to hydrogen assisted cracking (HAC) during welding, and a suitable preheat temperature is recommended for welding depending on the alloy content and thickness of the plate (1). However, information available on the HAC susceptibility of these steels is rather limited, especially in the case of alloys containing high amount of Cr. In these alloys both weld metal and HAZ are reported to be susceptible to HAC (2) and the susceptibility of weld metal increases with hardness (3). Most of the test results available on these steels are obtained from various cracking tests like, implant test (4), Y-groove test (5), G-BOP (Gap bead-on-plate) test (3), University of Tennessee-Modified Hydrogen Sensitivity Test (UT-Mod. HST) (2), etc.

In the present study, HAC susceptibility of 2.25 Cr-1Mo steel was examined using UT-Mod. HST. Hydrogen content in the weld was varied by varying the volume % of hydrogen in the shielding gas and was measured using Gas Chromatograph. The critical preheat temperature required to prevent cracking up to a maximum strain of 4% was determined for different levels of diffusible hydrogen content in the weld.

EXPERIMENTATION

The material used in this study was 2.25 Cr-1Mo steel with a composition of C=0.118, Cr=2.184, Mo=0.998 and Ni=0.161 wt% in the normalized and tempered condition. The specimens were machined out for testing from 12 mm thick plates.

Cracking Tests

Susceptibility of this steel to HAC was studied using UT-Modified Hydrogen Sensitivity Test (UT - Modified HST). This test was originally developed at Rensselaer Polytechnic Institute

(RPI) (6-8). It was further modified at the University of Tennessee to enhance its applicability (2). In this test, a specimen for cracking test was prepared by making a bead - on - plate weld on a specimen blank of dimensions 40x15x3 mm kept on a copper fixture, using autogenous GTAW process. It was possible to preheat the copper fixture so that welds could be prepared at different preheat temperatures. Schematic sketch of the copper fixture is shown in **Fig. 1**. Welding parameters were chosen such that the melting was confined to half the thickness of the specimen. Welding was carried out at a speed of 125 mm/min. The current and voltage were maintained at 90 A and 12V respectively. Further details of the test procedures are given elsewhere (9).

After welding the specimen was removed from the copper fixture and allowed to cool in air up to a temperature of 50°C and then loaded to a pre-determined strain level using a loading fixture, as shown in **Fig. 2.** In the case of a specimen prepared with preheat, the specimen was allowed to cool in the copper fixture up to the preheat temperature and then only unloaded from the fixture. The nominal augmented strain on the surface of the specimen after loading is given by the formula

$$\varepsilon = t/2R$$

where ε = nominal augmented strain, t = specimen thickness (3mm), and R = bending radius. By varying the diameter of the die block, it was possible to obtain different augmented strains on the specimens. The specimen was loaded for a minimum of 24 hrs and checked for cracks. The minimum temperature at which no cracks were obtained at two consecutive tests was taken as the critical preheat temperature. Critical preheat temperatures were determined for different vol.% of hydrogen in the argon shielding gas.

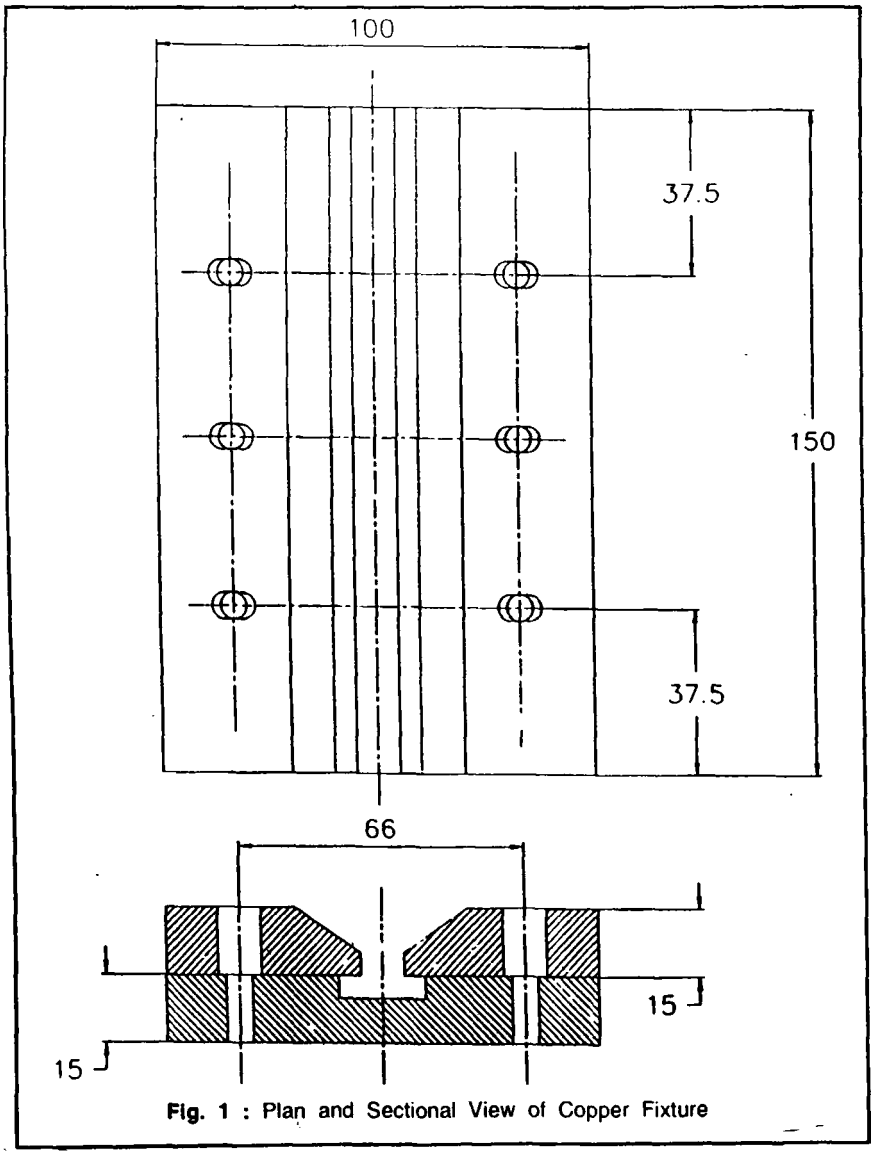

Fig. 1 : Plan and Sectional View of Copper Fixture

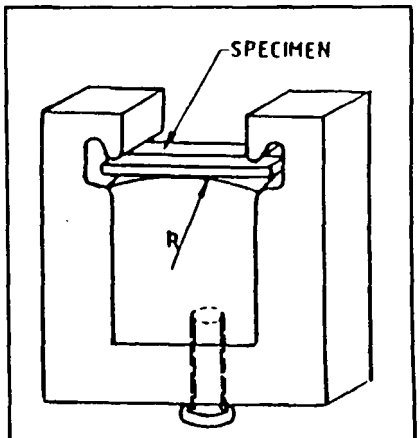

Fig. 2 : Loading Jig for UT-Modified Hydrogen Sensitivity Test

Determination of Cooling Time ($t_{8/5}$)

The time taken by the welds to cool from 800 to 500°C was determined from the weld cooling curve. Cooling curves for welds made at different preheat temperatures were obtained by plunging one end of W-Rh thermocouple into the weld metal as welding was in progress. The other end of the thermocouple was connected to an x-t recorder which plotted the variation in thermo emf with time and $t_{8/5}$ values for welds prepared at different preheat temperatures were estimated from these plots. The welds were prepared exactly in the same way as prepared for the cracking test.

Measurement of Diffusible Hydrogen from Welds

Collection of Hydrogen Evolved from the Weld

The hydrogen evolved from the weld was collected in a specimen chamber. A schematic sketch of the chamber is shown in **Fig. 3**. The specimen chamber is made of stainless steel and consists of an inlet and an outlet connected to needle valves. The chamber can be opened or closed using a plug and the leak tightness of the plug is ensured with the help of an O-ring. The chamber was subjected to He leak testing in vacuum before use.

For measuring diffusible hydrogen, specimens were prepared exactly in the same way as prepared for UT-modified hydrogen sensitivity test. After welding the specimen was removed from the copper fixture and put inside the specimen chamber and closed. The chamber was then closed, flushed and then filled with argon gas at a pressure of 2Kg/cm². The chamber was kept in an oven at 45±2°C for 72 hrs for the complete evolution of diffusible hydrogen from the specimen. The concentration of hydrogen in the gas mixture present in the chamber was then measured using Gas Chromatograph.

Measurement of Hydrogen Concentration and Estimation of (H_D) Content

The measurement set-up consisted of specimen chambers containing the weld specimen, six port gas sampling valve and a gas chromatograph with molecular sieve as the column material and thermal conductivity detector as the detector. Argon gas was used as the carrier gas. This is a slightly modified version of that reported by M.A. Quintana and

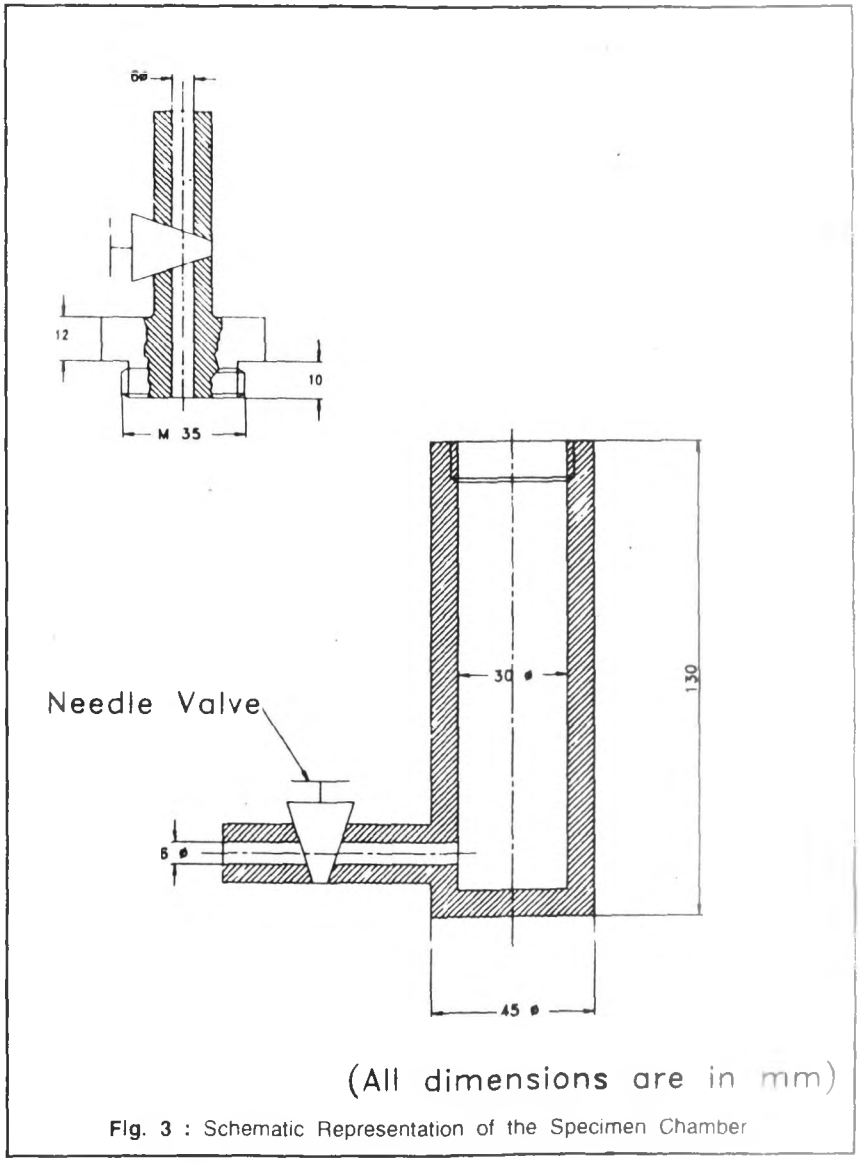

Fig. 3 : Schematic Representation of the Specimen Chamber

J.D. Dannecker (10). A schematic of the set-up is shown in **Fig. 4**. The specimen chamber is connected to GC through a six port gas sampling valve provided with a sample loop of known volume. Sampling valve collects a known volume of the gas from the chamber and injects into GC which gives the concentration of hydrogen in the chamber. Before every set of measurements, GC was calibrated using argon-hydrogen mixtures of known concentrations and measurement was carried out using exactly the same operating conditions as were used for calibration.

From the concentration of hydrogen in the chamber, the total volume of hydrogen at STP was calculated based on the volume of the chamber and pressure of gas inside it. Length and cross sectional area of each bead was separately measured to calculate the (H_D) content per 100 cc of weld metal.

Metallography and Hardness Measurements

Metallography examination was carried out on the top surface of the weld specimen. It was etched with 2% Nital to reveal the weld metal, HAZ and fusion line. For hardness measurement specimens which were not subjected to cracking tests were used and hardness variation as a function of preheat temperature was studied.

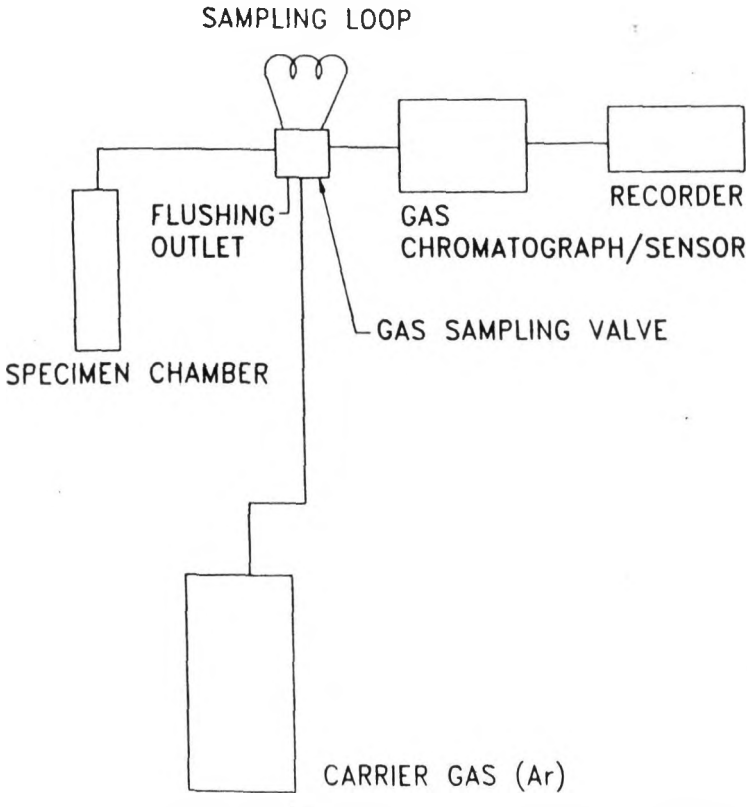

Fig. 4 : Line Diagram Showing Diffusible Hydrogen Measurement Set-up

RESULTS

Fig. 5 shows the variation of $t_{8/5}$ with preheat temperature. The values shown are the average values of the $t_{8/5}$ estimated from three different cooling curves obtained for same preheat temperature. It may be noticed that $t_{8/5}$ varied from 2.6 to 12s as the preheat temperature increased from 24 to 200°C. Also shown in the figure is the variation of hardness of the weld metal with preheat temperature. There is no substantial variation in hardness with preheat temperature and the value is above 400 VHN. Microstructural examination revealed that the structure remained predominantly martensitic both in the weld metal and coarse grained HAZ for all the preheat temperatures studied.

Cracking tests revealed that both weld metal and heat affected zone are susceptible to cracking. A typical micrograph showing a crack extending to HAZ is shown in **Fig. 6.** It may be noted that the propagation of the crack is arrested in the fine grained heat affected zone. Results of the cracking tests at different preheat temperatures are shown in **Fig. 7**. The minimum preheat temperature to prevent cracking increased from 100 to 175°C with increase in hydrogen content in the shielding gas from 1 to 5%. **Fig. 8** shows the variation in the diffusible hydrogen content (H_D) in the weld metal with increase in hydrogen in the shielding gas. As the weld beads are autog-

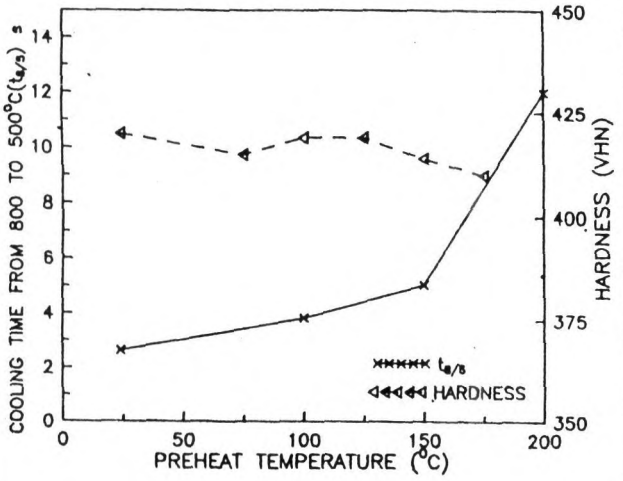

Fig. 5 : Variation of Cooling Time ($t_{8/5}$) and Hardness with Preheat Temperature

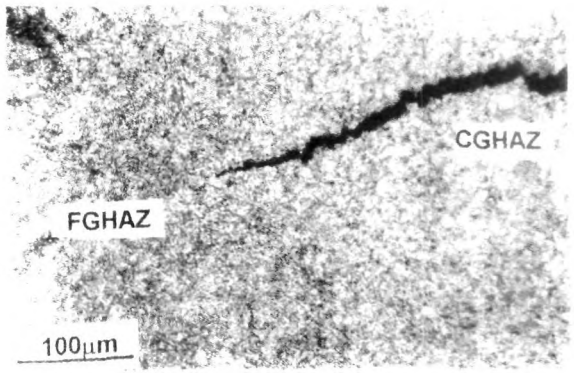

Fig. 6 : Microstructure of 2.25Cr-1Mo Steel HAZ with a Crack

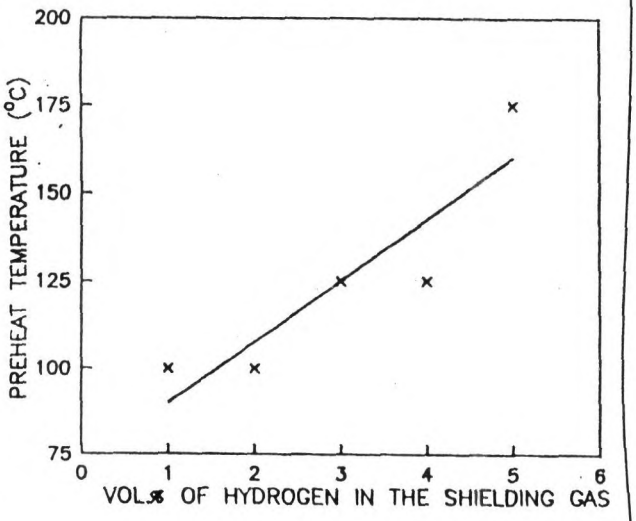

Fig. 7 : Variation of Critical Preheat Temperature with Vol.% of Hydrogen in the Shielding Gas

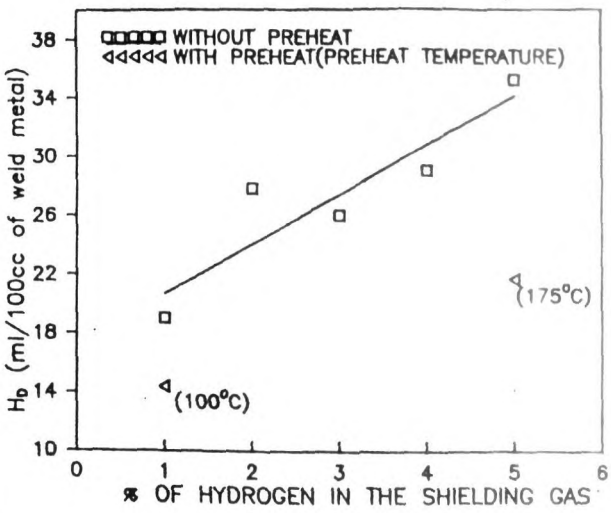

Fig. 8 : Variation of H_D Content with Vol.% of Hydrogen in the Shielding Gas

enous and volume of fused metal is mesured instead of its mass, (H_D) is represented as ml/100 cc of weld metal. It varied from around 20 to 35 cc/100 ml of weld metal with increase in hydrogen content in the shielding gas from 1 to 5%. A comparison of (H_D) values and minimum preheat temperature revealed that the preheat temperature increased with increase in (H_D) in the weld metal. **Fig. 8** also contains the results of (H_D) measurements carried out on welds at minimum preheat temperature required to prevent cracking for 1 and 5% of hydrogen in the shielding gas. It can be seen that (H_D) values are substantially lower than those obtained for welds prepared without preheat.

DISCUSSION

The above results show that both weld metal and base metal are susceptible to hydrogen assisted cracking. This is in agreement with the results of various studies reported on this steel (2-5). However, no systematic study has been carried out to vary the hydrogen content in the weld and determine the minimum preheat temperature required to prevent cracking as a function of hydrogen content in the weld. In one of the studies in which UT-Modified HST has been used to study HAC, authors have assumed that (H_D) content in the weld is of the order of 17ml/100gm of weld metal for 5% hydrogen in the shielding gas. However, the present study shows it is only of the order of 4-5ml/100gm of weld metal (assuming density of the steel = 7.8 gm/cc). It is known that the (H_D) content varies with material (11) and assumption may be based on the measurements carried out on a different material.

A comparison of variation of hardness and (H_D) with preheat temperature reveals that preheating contributes only to removal of hydrogen from the weld metal and doesn't result in change in hardness. Hardness of the order of 400 VHN is typical of a martensitic microstructure in this steel and even at a preheat of 200°C the structure has remained predominantly martensitic. This is only to be expected because of the high hardenability of the steel and fairly high cooling rate experienced by the specimen after welding on a copper fixture even though the fixture is preheated. However, in actual welding conditions, the cooling rate experienced by the weld would be such that the microstructure is predominantly bainitic and bainite has a lower hardness than the martensite. Thus the preheat temperature determined would be conservative and use of preheat temperature determined by this test should ensure crack-free weld in actual welding conditions. It may be noted that the minimum preheat temperature determined in this study is in the same range as recommended for this steel (1,12).

Most of the (H_D) measurements reported in literature are carried out without preheating the sample and hence data is not readily available to compare the results of (H_D) measurements on preheated samples. However, a comparison of the (H_D) values obtained without preheating and with preheating to minimum temperature required to prevent cracking (**Fig. 8**) reveals that there is substantial reduction in the diffusible hydrogen content of the welds with preheating. As there is no significant change in the microstructure during preheating, one would expect the diffusible hydrogen content in the weld prepared with sufficient preheating to prevent cracking, to be lower than the minimum value of 20cc/100cc of weld metal obtained without preheating. For samples prepared with 1% hydrogen in the shielding gas with a preheating of 100°C it is indeed lower (around 14) but for 5% hydrogen in the shielding gas it is slightly higher (around 21). This difference in the (H_D) values may be explained based on the variation in the hydrogen trap density in the two welds prepared at different preheat temperatures. It has to be noted that during (H_D) measurement, only the supersaturated hydrogen that diffuses out as hydrogen atoms and combines to form hydrogen molecules outside the metal is measured. A portion of the supersaturated hydrogen would remain in the metal trapped by various defects like inclusions, grain boundaries, lath boundaries etc.

and hydrogen thus trapped is called residual hydrogen and this does not contribute to cracking. In order to release the residual hydrogen from the metal, degassing has to be carried out at a much higher temperature than that used for (H_D) measurement. As the trap density increases, volume fraction of residual hydrogen increases resulting in a net reduction in the diffusible hydrogen content. Though the structure of the weld metal (and also that of coarse grained heat affected zone) is martensitic, the different cooling rates experienced at different preheat temperatures affect the size of the martensite laths formed from the prior austenite dendrite or grains. Higher the cooling rate (lower the preheat temperature) smaller the size of the laths and higher the area of lath boundaries. It has been shown using autoradiography studies using tritium isotope that lath boundaries trap hydrogen very effectively (13). This study which was carried out on a Cr-Mo steel also showed that a specimen quenched from 890°C (fine lath size) released more hydrogen than the specimen quenched from 1100°C (large lath size) when degassed at 150°C after 30 minutes and 24 hrs of hydrogen charging, indicating higher residual hydrogen content in the sample quenched from lower temperature. In a similar way, in the case of welds prepared at lower preheat temperatures, there will be more traps for hydrogen compared to those prepared at higher preheat temperatures. In other words, the observed difference in the diffusible hydrogen contents in the welds prepared at minimum preheat temperatures required to prevent cracking, may be due to the difference in the relative proportion of the diffusible and residual hydrogen contents in them. It is necessary to carry out measurement of residual hydrogen content in the welds to confirm this.

CONCLUSIONS

The major conclusions from the present study are the following :

1. A commercially available Gas Chromatograph can be suitably modified to measure diffusible hydrogen in the welds.
2. For 2.25Cr-1Mo steel, both weld metal and base metal are susceptible to hydrogen assisted cracking.
3. Minimum preheat temperature required to prevent cracking varies from 100 to 175°C with increase in (H_D) from 20 to 35ml/100cc of weld metal.
4. In the temperature range studied, preheating reduces only (H_D) content in the welds and does not change the microstructure.

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