# Metallurgical control of hydrogen induced cracking in high strength steel weldments

By
**Dr. Tapan Kumar Pal**
Professor, Metallurgical Engg. Department
Jadavpur University, Calcutta – 700 032

## Introduction

Hydrogen Induced Cracking (HIC) is still a major problem encountered during the welding of high strength structural steels. Historically, the risk of HIC has been greatest in the HAZ of the parent metal as a result of the rapid cooling rate experienced during the welding thermal cycle. However, steelmakers have made considerable improvements in terms of solving HAZ hydrogen cracking problems and have implemented strategies involving carbon and alloy control, controlled rolling schedules, on-line accelerated cooling, and so on. However, as we move to steels which have higher yield strength hydrogen-induced cracking problems will shift from the HAZ to the weld metal. HIC formed in weld metal was reported by Flanigan et al [1] for the first time.

It has been well recognised that HIC occurs in the HAZ or in the weld metal if the following four conditions are present simultaneously [2] :
a) critical concentration of diffusible hydrogen

b) stress intensity of sufficient magnitude

c) a microstructure susceptible to hydrogen

d) temperature lower than about 200°C

Although reductions in welding consumable diffusible hydrogen potentials have been achieved, there have been no corresponding developments in weld metal microstructures. Hence, when fabricators have applied the principles of BS 5153 (Arc welding of carbon and carbon manganese steels) to prevent HAZ HIC, cracking of the weld metal has been reported, specially at yield strength levels of 500 MPa and above. For this reason, expensive preheat or post weld heat treatment (PWHT) has been required to prevent weld metal cracking above that necessary to prevent HAZ cracking [3]. Steel fabricators are presently increasing their use of the higher strength low carbon equivalent steels, because this approach can reduce costs through shorter welding times and reduced material coats.

Careful analysis of the welding system i.e. the welding arc, the molten weld pool, and solidified weldment, clearly shows that hydrogen can be picked up from multiple sources, including the welding consumables, plates and shielding gas. By controlling the amount of hydrogen coming from these sources, lower weld metal hydrogen can be obtained. Once the hydrogen is incorporated in the weld metal, however, other strategies are required to minimize the hydrogen content. This paper summarises the interactions between hydrogen and microstructural features along with the prevention of hydrogen induced cracking in high strength steel weldments.

## Hydrogen Source

The diffusible hydrogen content in steel weldment is a critical key factor which determines hydrogen cracking susceptibility and the interaction among hydrogen, nitrogen and oxygen plays a major role in determining the supply of diffusible hydrogen delivered during any arc welding operation. Work is proceeding to reduce the proportion of hydrogen which enters the weld pool by modifying the chemistry of welding arc plasma. This is accomplished through the addition of selected oxides and fluorides in the flux as shown in Fig. 1. The content of hydrogen absorbed during are welding increases as the arc current is increased, when the partial pressure of the hydrogen in the·shielding gas increases and when the moisture content of the flux

| Process | Welding material | Weld metal hydrogen content, ml/100 g | |
| --- | --- | --- | --- |
| | | Diffusible | Residual |
| SMAW | Basic, lime coated (dried at 70°C) | 9.7 | 0.9 |
| | Rutile (from stock) | 24.8 | 6.4 |
| | Acid (from stock) | 18.9 | 7.4 |
| | Low hydrogen E9018-M (dried at 300°C) | 6.6 | — |
| | Cellulose, E8010-6 (as received) | 54.0 | — |
| GMAW | Ar – 20% $CO_2$ (1.6 mm dia., 28V) | 1.8 | — |
| SAW | Fused (4 days after dried) | 4.5 | — |
| | Agglomerated (4 days after dried) | 7.0 | — |

Table 1 : Typical weld metal hydrogen content by various welding processes
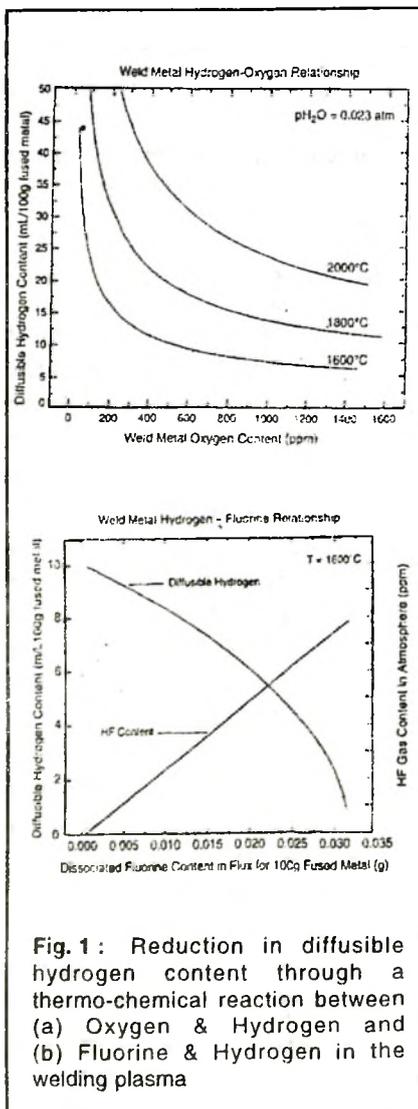


Fig. 1 : Reduction in diffusible hydrogen content through a thermo-chemical reaction between (a) Oxygen & Hydrogen and (b) Fluorine & Hydrogen in the welding plasma

formulation in the shielded-metal-arc, flux-cored arc and in submerged arc welding increases. The weld metal hydrogen content is generally expressed by the content of diffusible hydrogen. Table 1 gives typical values of the diffusible hydrogen content found for various welding processes.

## Metallurgical Effects on HIC

The integrity and the performance of a welded joint depend on the weldment microstructure and properties [4]. During welding, the interaction of the heat source and the material leads to rapid heating, melting and circulation of the molten material in the weld pool. As the heat source moves away from the molten region, solidification of the material takes place. The resulting thermal cycle plays an important role in determining weldment structure and properties.

The metallurgical change in the HAZ results in the embrittlement and the strength mismatching in the HAZ. In particular, sometimes, the significant deterioration of the fracture toughness in the HAZ has been

observed in the common structural steels; compared to its superior in the recent base metals. Among other things, the larger austenite grain size can displace the C.C.T. curve to longer reaction times producing more widmanstatten, or increasing the possibility of martensitic or bainitic transformation products on cooling. This, together with the larger grain size, may result in lower toughness and the risk of HIC.

There is general agreement that the conditions which lead to HIC during welding are : (1) a critical hydrogen concentration, (2) residual and applied stress above a certain level and (3) a susceptible microstructure [5, 6]. All three conditions are interrelated : the critical hydrogen concentration necessary for cracking could be reduced by an increase in the stress or by the presence of microstructures with a greater HIC susceptibility. Changes in the parameter often affect changes in the other two. For example, a change in welding parameters that change the weld metal cooling rate is likely to alter hydrogen concentration, the residual stress, and the microstructure. Twinned martensite has been shown to be more susceptible to HIC than lath martensite [7]. This may be due to either the higher residual stresses present in the twinned martensite structure or an inherent susceptibility of the structure itself.

It is unclear which particular aspect of a microstructure makes it susceptible to HIC. Broadly speaking, the HIC susceptibility of a microstructure increases with increasing hardness (or strength) [8, 9]. A number of workers have indicated that it is the particular

**GROUP**          **EQUATION**

$$CE_{IIW} = C + \frac{Mn}{6} + \frac{Cu + Ni}{15} + \frac{Cr + Mo + V}{5}$$

$$CE_{WES} = C + \frac{Si}{24} + \frac{Mn}{6} + \frac{Ni}{40} + \frac{Cr}{5} + \frac{Mo}{4} + \frac{V}{14}$$

$$CE_{winterton} = C + \frac{Mn}{6} + \frac{Cu}{40} + \frac{Ni}{20} + \frac{Cr}{10} + \frac{Mo}{4} + \frac{V}{14}$$

$$CE_{Stout} = C + \frac{Mn}{6} + \frac{Cu}{40} + \frac{Ni}{20} + \frac{Cr + Mo}{10}$$

$$CE_{Cottrell} = C + \frac{Mn}{6} + \frac{Cr}{5} + \frac{Mo}{5} + \frac{V}{3} + \frac{Nb}{4C} + \frac{0.0001}{S}$$

$$CE_{DnV} = C + \frac{Si}{24} + \frac{Mn}{10} + \frac{Ni + Cu}{20} + \frac{Cr}{5} + \frac{Mo}{4} + \frac{V}{14}$$

$$Pm = C + \frac{Si}{20} + \frac{Mn}{10} + \frac{4P}{3} + \frac{Cu}{20} + \frac{Cr}{30} + \frac{Mo}{20}$$

$$Pcm = C + \frac{Si}{6} + \frac{Mn}{5} + \frac{Cu}{5} + \frac{Ni}{3} + \frac{Cr}{4C} + \frac{Mo}{S} + \frac{V}{S} + 5B$$

$$CE_{HSLA} = C + \frac{Mn}{16} + \frac{Ni}{50} + \frac{Cr}{23} + \frac{Mo}{7} + \frac{Nb}{5} + \frac{V}{9}$$

$$CE_{PLS} = C + \frac{Si}{25} + \frac{Mn}{16} + \frac{Cu}{6} + \frac{Ni}{60} + \frac{Cr}{20} + \frac{Mo}{40} + \frac{V}{15}$$

$$CEN = C + A(C) \left( \frac{Si}{24} + \frac{Mn}{6} + \frac{Cu}{16} + \frac{Ni}{20} + \frac{Cr + Mo + Nb + V}{5} + 5B \right)$$

Where A(C) = 0.75 + 0.25 tank [20(C - 0.12)]

**Table 2 :** Proposed carbon equivalent for assessing weldability

microstructure used to achieve strength which will have a greater influence on HIC than its strength level per se [7, 10] and that neither hardness [11-13] nor strength [14, 15] are reliable indicators of HIC susceptibility.

The microstructure of the HAZ significantly affects HIC; martensitic microstructures are especially sensitive to cracking. The HAZ hardness is regarded as a rough index describing the susceptibility to HIC. The value of 350 HV is often specified as the maximum allowable HAZ hardness for avoiding HIC. Furthermore, line pipe steels and pressed vessel steels for sour gas service are required to satisfy a HAZ hardness of less than 248 HV. It is to be mentioned here that the HAZ hardness is determined by the post-weld cooling rate and the chemical composition of the steel. Therefore, it is necessary to select the proper welding conditions including welding heat input and preheat temperature e.g. proper welding cooling time between 800° and 500°C ($T_{8/5}$) for the phase transformation behaviour in steel welding.

Regulation of HAZ hardness to a certain level has long been considered essential for safe welding practice. However, recent experience in the welding of modern steels has indicated that there is a trend towards cracking which occurs at a lower HAZ hardness level as the carbon or carbon equivalent (CE) decreases. It is to be mentioned here that carbon equivalent can be represented as the resistance to HIC. Table 2 shows some of the important

CES which can be divided into four apparent groups. The carbon equivalents of group C are considered to assess the weldability of low carbon, low alloy steels more satisfactorily than does group A. The carbon equivalents of group B are between groups A and C. The CEN of group D was proposed to assess the weldability of a wide range of steels. In addition to the major alloying elements, microalloy elements and some impurities have been recognised as affecting the HIC, especially in modern high strength steels. The effect of microalloying on HIC is often related to the HAZ hardenability. For example, clean steel has been found the increased risk of HAZ HIC. Therefore, increased hardenability is now recognised to be the main cause for HIC.

The microstructural features of HIC have been studied in low carbon equivalent high strength steel welded structures [16]. Authors suggested that an increase in number of interfaces formed by the second phases contributed to HIC resistance of weld metals by two mechanisms : The first is the ability of an increased number of interfaces to impede the progress of cracking and second is the capacity of there interfaces to distribute diffusible hydrogen more uniformly. Thereby the risk of exceeding some critical local concentration required to initiate and propagate a crack will reduce.

Based on these two mechanisms, the susceptibility to HIC is more with grain boundary ferrite than acicular ferrite even though the former is less stronger. Similarly presence of high volume fraction of inclusion is

beneficial in weld having higher diffusible hydrogen concentration. On the other hand, presence of M/A & carbide at lower diffusible hydrogen concentration appeared beneficial. This was due to the saturation of the M/A and carbide interfaces, which were likely to be acting as traps for diffusible hydrogen.

The susceptibility to HIC of multipass high strength steel weld metal was mainly controlled by local variation of microstructure but little affected by microhardness variation [17]. The most susceptible microstructure was found to be columnar grain structure tempered to some extent. The extent of tempering could play some role in crack formation by the formation of grain boundary carbide. Thereby either decreasing the carbon or the grain boundary decoration with grain boundary ferrite would be beneficial in improving the resistance to HIC of high strength weld metal.

## Prevention of HIC

The HIC in most steel occurs when a weld cools below 100°C. Some hydrogen effuses out from the weld until it reaches 100°C and only the remaining hydrogen contributes to HIC. Therefore, HIC can be prevented by making the cooling time to 100°C, $t_{100}(S)$, longer than the critical cooling time, $(t_{100})_{cr}$. $t_{100}$ is influenced by the preheat temperature, preheat method, preheat width, plate thickness, heat input, ambient temperature, and wind velocity.

The necessary preheat for welding practice seems to be determined mostly on an empirical basis by the use of guideline tables or monographs which consider major

influencing factors in HIC. The british standard BS 5135 (1984) [18] determines the necessary preheat by four major factors : $CE_{IIW}$ carbon equivalent, weld metal hydrogen content, welding heat input and the combined thickness of the plate. The levels of weld metal hydrogen are classified into four groups : high ($H_{IIW}$ > 15 ml/100 g), medium (15 > $H_{IIW}$ > 10), low 10 $H_{IIW}$ > 5) and ultralow (5 > $H_{IIW}$). According to American Welding Society (AWS) D 1.1-86 [19], the method for HIC varies depending on the steel composition, as shown in Fig. 2. For zone I steel, reduction of HAZ hardness is not achieved by increasing heat input,
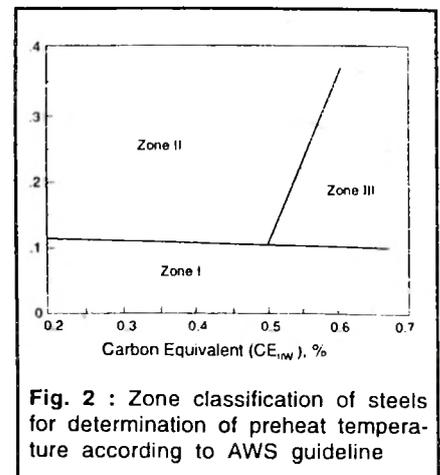
**Fig. 2** : Zone classification of steels for determination of preheat temperature according to AWS guideline

and thus the use of low hydrogen welding material is recommended. The zone II steel, whose carbon content > 0.11%, can expect a reduction in HAZ hardness by increasing the heat input. For zone III steel; use of the hydrogen control method is suggested because heat input becomes excessively high for the control of HAZ hardness below a certain level.

The trend of low carbon in steels has greatly reduced the HIC. The safe welding condition showed be
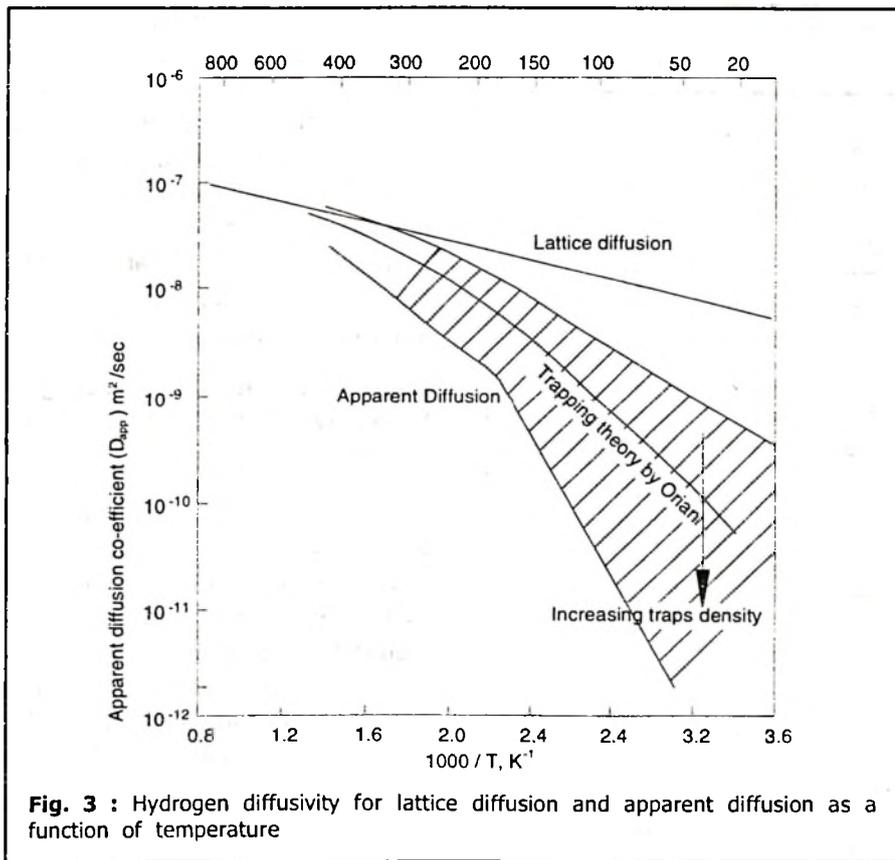
**Fig. 3 :** Hydrogen diffusivity for lattice diffusion and apparent diffusion as a function of temperature

preheating when welding heavy section steels or steel susceptible to cracking. Post-heating is employed immediately after welding or after keeping a weldment at not below the critical temperature.

## Hydrogen trapping

When the weld pool solidifies, hydrogen diffuses from the newly solidified metal due to the concentration gradient.

The suppression of diffusible hydrogen can also be achieved by the introduction of selected rare earth and transition metal additions to the weld metal to serve as hydrogen traps. These traps, in the form of oxides or carbo-nitrides have high binding energies with hydrogen. They are capable of immobilizing hydrogen at temperatures ranges much higher than 100°C, before the risk of cold cracking energies [20].

The ability of trap site to hold a hydrogen atom is associated with the hydrogen trap binding energy. A trapped hydrogen atom must acquire an energy substantially greater than the lattice migration energy to escape the trap and to contribute to the measured diffusivity. Trapping parameters such as the trap binding energy and peak temperature of trap are examined by using the thermal analysis. Since HAC depends on hydrogen diffusivity, hydrogen trapping may serve a beneficial role by hindering the hydrogen cracking mechanism. In other words, hydrogen traps influence the likelihood of cracking by controlling the availability of hydrogen to the critical cracking locations. The effect of hydrogen trapping on the diffusivity of hydrogen is illustrated in Fig. 3, where the apparent

determined to avoid weld metal cracking rather the HAZ cracking. An improved guideline to avoid weld metal cracking is certainly needed. To avoid HAZ cracking, it is preferable to use so-called under match welding, in which a weld metal of low strength, compared with that the base metal, is used. In case of under matching weld metal (less alloyed than base metal), the martensite start temperature is higher in the weld metal than in the base metal and consequently the weld metal will transform before the base metal. Hydrogen transport from the transformed weld metal will be rapid but the austenitic HAZ will serve as a hydrogen barrier and a build up of hydrogen will occur at the HAZ [7]. On the other hand, in weldments of high strength low carbon steels the weld metal is often

more highly alloyed than the base metal. This may be a contributing factor in the increased likelihood of weld metal cracking in low carbon microallyed steels [7]. It is to be mentioned here that the final weld metal composition depends on the base material and filler metal chemistries and on the welding process used. Also, the deposit sequence employed in the field will be important since the root region may have a quite different chemistry from the final passes of multipass welds. In particular, the residual element content in steel weld metal is quite different from that in the HAZ, e.g. conventional arc welding processes deposit steel weld metals containing high oxygen levels (from 250 ppm to as high as 100 ppm). Post-weld heat treatment is sometimes employed in addition to

| Hydrogen | Activation Energy KJ/mol | Peak Temp.(°C) |
|---|---|---|
| **Very reversible** | | |
| interstitial lattice site | 7.7 | — |
| **Reversible** | | |
| Ti substitutional atom | 26 | — |
| grain boundaries | 17.2 | 112 |
| " | 26 | — |
| dislocations | — | 270 |
| " | 24 | — |
| " | 24 | — |
| " | 26.8 | 215 |
| ferrite/carbide interface | — | 115 |
| ferrite cementile interface | 18 | 160 |
| " | 18.4 | — |
| | — | 123 |
| ferrite carbide average | 26.8 | — |
| (tempered martensite) | 29.7 | — |
| (overaged martensite) | 36.4 | — |
| **"Irreversible"** | | |
| microvoids | 48.3 | 338 |
| " | 35.2 | 305 |
| " | 36.8 | 330 |
| iron oxide | 50.6 | 430 |
| | 69.5 | 530 |
| MnS | 72.3 | 496 |
| $Al_2O_3$ | 79 | 580 |
| $Al_2O_3$ or $SiO_2$ | 86.2 | 630 |
| | 112.1 | 750 |
| TiC | 88.2 | |

**Table 3 :** Identified hydrogen traps with a range of trap energies

diffusion coefficient of hydrogen in steel is shown as a function of temperature. An increase in temperature decreases the trap energy then decreasing its tendency to affect hydrogen diffusion. Above about 400°C, the apparent diffusion coefficient approximates the diffusion coefficient of lattice diffusion, while below the temperature the apparent diffusion coefficient is affected by hydrogen trapping.

The hydrogen will be found in different types of trap sites which can be atomic and microstructural imperfections such as vacancies, solute atoms, dislocations, bulk microstructure (FCC and BCC), grain boundaries, voids, and second phase particles [21]. These sites are sinks which retain the hydrogen atom even during thermo-mechanical loading. There are two types of trap sites. If binding energy is small, the corresponding traps are referred to

as reversible traps and can act either as hydrogen sinks, which deposit hydrogen atoms to stronger traps. On the other hand, large binding energy gives rise to traps which are termed irreversible traps. These traps normally will not release hydrogen even at higher tem-peratures [22]. Many traps have been identified and are listed in Table 3.

### Table 3 : Identified hydrogen traps with a range of trap energies

Trapping sites in weld metal have been shown both theoretically [23] and experimentally to reduce hydrogen mass transport and welding consumables are being designed to use this effect to limit the amount of hydrogen available for transport to possible crack sites. The use of hydrogen traps is a promising development in low hydrogen consumable technology.

**Concluding Remarks :**

A number of microstructural modifications could be used to improve the HIC resistance of current high strength steel weld metals. Additionally, a careful examination of the effects of microstructural features on HIC resistance will allow for a more rational development of welding consumables for maximum HIC resistance.

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