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Research to establish a systematic approach to safe welding procedure development using austenitic filler material for fabrication of high strength steel

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Abstract

High strength, thermo-mechanically treated steels often require very careful control of welding procedures to ensure freedom from hydrogen assisted cold cracking (HACC) in the weld metal and heat affected zone (HAZ) whilst avoiding undue loss of heat affected zone strength. In practice this leads to restricted operating envelopes and the requirement for rigorous procedure management. An alternative approach is the use of austenitic welding consumables but whilst this may be expected to expand the process parameter envelope its use is often restricted by adherence to procedures previously developed for ferritic consumables. The current paper describes the research undertaken to establish a systematic approach to the development of optimum, cost effective procedures for welding high strength quenched and tempered steel with an austenitic filler.

Keywords
strength, development, procedure, welding, safe, approach, systematic, establish, research, high, fabrication, material, filler, steel, austenitic

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Abstract:

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Background

High strength, high hardness, thermo-mechanically treated steels are used extensively in wear resisting applications for materials handling and mining and as armour in defence applications. In most cases it is necessary to fabricate the material by welding. During welding the materials may however be subject to loss of heat affected zone hardness if overheated and hydrogen assisted cold cracking if strength matching ferritic filler materials are used with insufficient pre or post heat. The operating tolerance band when using ferritic fillers is therefore very narrow. In repair situations it has been common to use austenitic fillers to reduce the risk of HACC particularly when the parent material has a high carbon equivalent and close procedural controls are difficult to achieve. The present work was carried out to investigate the feasibility of developing less constrained welding procedures when welding with austenitic fillers.

The risk of HACC is related to three factors, figure 1:

1) The presence of hydrogen
2) A susceptible microstructure
3) Stress

All of these three must act in concert to cause HACC if any one of the factors is absent hydrogen cracking will not occur.

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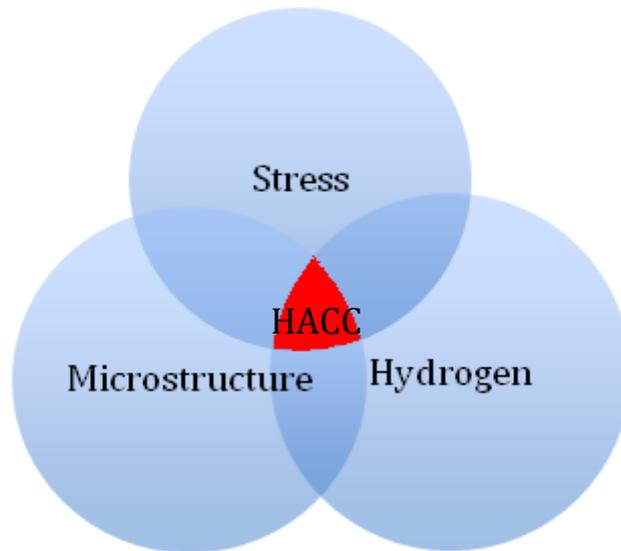
Figure 1 Basic factors responsible for hydrogen assisted cold cracking (HACC).

Stress is almost unavoidable in welded joints due to the contraction stresses and restraint. The level of stress may increase due to the joint design and section thickness. Joint design and welding sequence may be modified to decrease stress but it must be assumed that in more complex fabrications sufficient stress is available to contribute to HACC.

Martensitic structures are normally considered to be susceptible microstructures. Since high hardness is usually a design requirement for use of such materials it is common for the carbon equivalents to be high and normal base microstructures are bainitic or martensitic. Whilst this factor cannot be eliminated there is another issue related to microstructural transformation which influences the hydrogen distribution. If a ferritic filler is used the relative time between transformation of the filler and heat affected zone determines the direction of hydrogen diffusion: the solubility of hydrogen in the molten pool and the high temperature austenitic phase is high but it is much lower in the lower temperature ferritic or martensitic microstructure. If the filler material transforms before the austenitised area of the heat affected zone during cooling of the weld zone any hydrogen in the weld will be 'pumped' into the heat affected zone of the parent material. One obvious way to prevent this 'pumping effect is to use an austenitic filler. In the current investigation a high strength steel was welded with both ferritic and austenitic fillers.

Methodology

The high strength steel used was typical of those used for high hardness wear plate. Its chemical composition is listed in Table 1. The alloying design utilises the novel low manganese approach [1] in order to reduce centreline segregation of MnS inclusions and thus improve mechanical properties and weldability. The steel is microalloyed with Nb, Ti, V, and B in order to increase hardenability and enhance the HAZ properties.

Table 1. Chemical composition of investigated steel in wt% max.

Element	C	P	Mn	Si	S	Ni	Cr	Mo
Wt% max	0.32	0.02	0.4	0.35	0.005	0.35	1.15	0.3

Element	Cu	Al	Nb	Ti	V	B	CE_{IIW}
Wt% max	0.2	0.06	0.005	0.03	0.05	0.002	0.72

After hot rolling the plate is subjected to heat treatment process which consists of two phases: rapid water-quenching from austenite field which results in the formation of untempered martensite and subsequent low temperature tempering which improves the plate's toughness while maintaining high level of hardness. The tempering temperature is based on desired balance between hardness and toughness required for particular application. The microstructure of the investigated plate in Q&T condition consists of tempered martensite with an average hardness of 542 HV20.

Transformation temperature

The continuous cooling transformation behavior of the steel chosen was studied in previous work [2]. It was found that peak hardenability is reached at cooling rates representative of real welding conditions (cooling at 20°C/s to 30°C/s) in the temperature range of 800-200°C. At these cooling rates the austenite to martensite transformation takes place over a range of 200°C with transformation start and finish temperatures being approximately 410°C and 205°C respectively.

Hydrogen investigation

The objective of the tests reported below was to investigate the difference in hydrogen distribution when ferritic and austenitic consumables were used, the effect of preheat (PH) and type of wire on weld hydrogen levels.

Samples were produced using austenitic AWS A5.22 E308 MoT1 type flux-cored wire, an AWS A5.9 EC308Mo type metal-cored wire, an AWS A5.9 ER(307) solid wire and ferritic flux-cored wire complying with AWS A5.20 E71T-1. Both flux cored wires have a similar rutile flux core since it is recognised that the flux system used can have a major influence on the level of hydrogen in weld metal and also its diffusivity.

A robotic GMAW system was utilised to deposit consistent single weld beads onto the test piece assembly clamped in a copper jig. The welding jig is shown in figures 2 and 3. Samples were not degassed prior to welding (as normally required in the relevant standard) in order to simulate the real welding conditions in industry. Preheating was carried out utilising an electric resistance heating blanket and undercooling of the test piece assembly to 7°C was achieved using dry ice (solid CO_2). After deposition of the weld beads, samples were quenched in stirred iced water and transferred in to the bath of liquid nitrogen. Samples were observed for uniform heat-tinting on the reverse side of the test piece assembly and the total length and centre piece to weld crater distance were measured. The run-on and run-off pieces were broken off and centre piece was thoroughly cleaned with a steel wire brush. The intervals spent outside the liquid nitrogen did not exceed 15s.
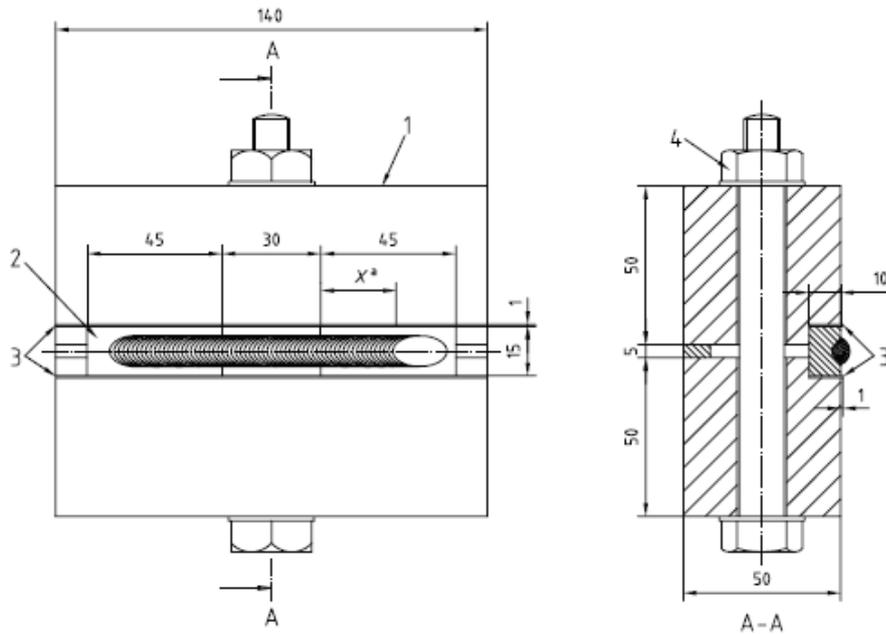
Figure 2 The standard diffusible hydrogen weld test jig assembly.

Figure 3 Welding fixture with test piece assembly before the deposition of the single weld and immediately after deposition of test run.

Parameters that were kept constant for all trials are shown in table 1. Variable welding parameters are displayed in table 2..

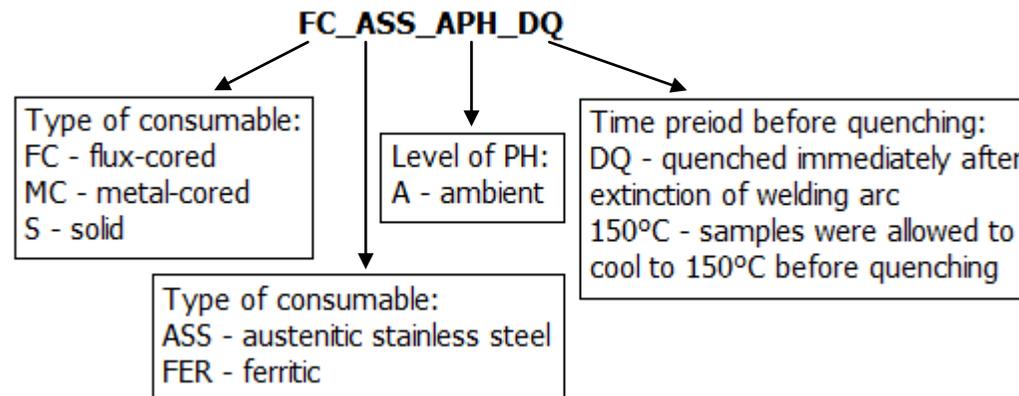**Table 1 Welding parameters common to all welding trials.**

Welding parameter	Value
Travel speed	400 mm/min
Wire feed speed	9.5 m/min
Stick out	16 mm
Voltage	26.3 V
Polarity	DC+
Wire diameter	1.2 mm

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Table 2 Variable welding parameters. Note: The values given in table are the average value of three trials welded under the same conditions.

Welding trial ID	PH [°C]	Quench [°C]	Shielding gas	Flow rate [l/min]	Amperage [A]	Heat input [kJ/mm]	Relative humidity [%]
FC_ASS_APH_DQ	20	Direct	Argoshield 52	16	215	0.85	48
FC_FER_APH_DQ	24	Direct	Argoshield 52	16	253	0.998	68
FC_ASS(PM+WM)_APH_DQ	22	Direct	Argoshield 52	16	218	0.86	70
FC_ASS_7°CPH_150°CQ	6	150	Argoshield 52	16	215	0.85	86
FC_ASS_80°CPH_150°CQ	80	150	Argoshield 52	16	213	0.84	74
FC_ASS_APH_150°CQ	22	150	Argoshield 52	16	220	0.87	65
MC_ASS_APH_150°CQ	22	150	Stainshield	16	294	1.16	58
S_ASS_APH_150°CQ	21	150	Stainshield	20	245	0.97	60

Figure 4 Welding trials identification explanation.

FC_ASS_APH_DQ

Type of consumable:
FC - flux-cored
MC - metal-cored
S - solid

Level of PH:
A - ambient

Time preiod before quenching:
DQ - quenched immediately after extinction of welding arc
150°C - samples were allowed to cool to 150°C before quenching

Type of consumable:
ASS - austenitic stainless steel
FER - ferritic

Diffusible hydrogen testing

Various test methods for determination of diffusible hydrogen content were reviewed. The 'Reference test method' - collection of hydrogen over mercury was selected to measure the diffusible hydrogen content present in the weldments. The gas collection apparatus is shown in figure 4. Testing was performed according to the AS/NZS 3752:2006 standard with few minor modifications. Every test was repeated three times for consistency. The first series of experiments was carried out twice in order to test the welding procedure as well as hydrogen measurement apparatus.
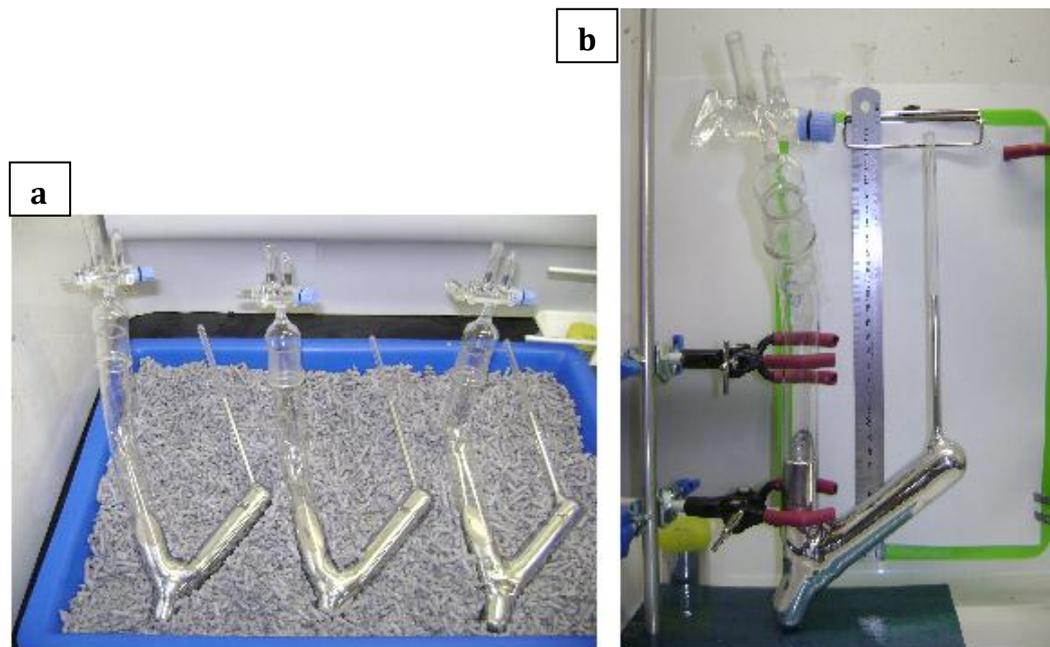

Figure 4 a –Test pieces submerged in mercury in evacuated Y-tube, b- Y-tube during measurement of the volume of evolved hydrogen.

Before transferring the test samples from liquid nitrogen into the y-tube containing mercury and evacuating the inner space, samples were placed into an acetone bath, allowed to rise to room temperature and dried with a jet of air. Samples were left undisturbed in the y-tube for 7 days at room temperature (around 20°C) to allow the majority of diffusible hydrogen to evolve and collect in capillary tube. The readings of atmospheric temperature and pressure were taken every day prior measuring the volume of hydrogen. The volume of the evolved hydrogen was corrected to standard temperature and barometric pressure measured and calculated by the equation:

$$V = \frac{273(P - H)(\pi r^2 \times C)}{760(273 + T)x100}$$

Where: V – volume of collected hydrogen at STP in millilitres; P – barometric pressure in mm of Hg; H – is the differential head of mercury between the two limbs

of the y-tube in mm; C – length of the gas column above the mercury in mm; r – inner radius of capillary tube in mm; T – ambient temperature at the time of hydrogen measurements in °C.

The weight of the test pieces was determined before the deposition of the weld bead (m_1) and also after completion of diffusible hydrogen testing (m_2) to nearest 10mg.The volume of diffusible hydrogen H_D in ml at STP per 100g of weld metal (WM) was calculated from the equation:

$$H_D = V \times \frac{100}{(m_2 - m_1)}$$

Residual hydrogen testing

Various methods of testing residual hydrogen content were also reviewed. The inert gas melt extraction method was selected because it is a rapid process and has a proven reliability. Testing was carried out according to the AS/NZS 1050 standard. The Eltra ONH-2000 total hydrogen testing facility used is shown in figure 5.

After completion of diffusible hydrogen testing, three slices transverse to the longitudinal axis of each centre piece were sectioned utilising a precision cutting machine. The weld metal was separated from the rest of the sample using a grinding wheel. In between grinding steps, samples were cooled in water to prevent heating due to friction and possible loss of residual hydrogen. The resultant samples of weld metal of approximately 1g were washed in acetone, dried and subjected to residual hydrogen testing.

Figure 5 Eltra ONH-2000 - Inert gas melt extraction test facility.

During the test samples were heated to melting temperature to release all residual hydrogen (in case of ASS WM residual hydrogen = diffusible hydrogen in austenite fcc lattice + hydrogen trapped in strong traps). The hydrogen is transported by the carrier gas to the analyser to determine its quantity in ppm. The volume of residual hydrogen in ppm was converted to ml/100g of WM at "normal" atmospheric pressure of 101325 Pa and a temperature of 20°C.

- o   Conversion of ppm to ml of hydrogen per 100g of WM

$$m = 1 \text{ppm of H} = 0.1 \text{mg of H in 100g of WM}$$

$$V[\text{ml ofH per 100g of WM}] = {}^{m}\!/_{\rho}$$

Where, V = volume of the gas, m = weight of the gas, ρ = density of the gas.

The density of hydrogen was calculated from the equation of state of an ideal gas:

$$pV = NkT$$

Where, p = atmospheric pressure, N = number of particles in the gas, k = Boltzmann constant = 1.38×10^{-23} J.K^{-1} and T = temperature.

The formula for calculation of the number of particles:

$$N = nN_A = {}^m\!/_{M_m} N_A$$

Where, n = amount of substance, N_A = Avogadro constant = 6.03×10^{23} mol^{-1}, M_m = molar mass.

Then,

$$pV = {}^m\!/_{M_m} R_m T$$

Where, R_m = molar gas constant = 8.316 $J.K^{-1}mol^{-1}$.

Then, the density of hydrogen can be calculated:

$$\rho = {}^m\!/_V \rightarrow \rho = {}^{pM_m}\!/_{R_m T}$$

Results

Every test was performed three times. The first series of three trials was also repeated twice to gain confidence in the repeatability and consistency of the results. Further, for every residual hydrogen test, three samples of weld metal were sectioned from all three samples welded under the same conditions and were tested for residual hydrogen. Results of both diffusible and residual hydrogen testing showed excellent consistency and repeatability.

Effect of welding consumable type on hydrogen level

Measured levels of diffusible and residual hydrogen are presented in table 3. To be able to directly compare the overall amount of hydrogen introduced into the weldment utilising ferritic and ASS flux-cored wires, samples were quenched in stirred iced water immediately after extinction of the welding arc. Samples were welded without preheat with ambient temperature of approximately 22°C. It was found that the ferritic consumable resulted in a diffusible hydrogen level approximately 18 times greater than that produced by the austenitic wire. The ferritic consumable was expected to produce around 6 to 10 ml/100g .The extremely high value recorded is thought to be due to hydrogen pick-up during prolonged inappropriate storage of the consumable although the original manufacturer's package had never been opened. However, the austenitic consumable was not stored in sealed environment for approximately 3 years. Thus, it is expected that to certain degree "fresh" properly stored consumables would result in lower hydrogen levels and these results could be taken as a 'worst case' scenario. To prove that all the diffusible hydrogen measured for samples welded with austenitic consumable originates only from the HAZ, the austenitic consumable was also deposited onto 304 austenitic plate. This should, based on the above theory, result in zero diffusible hydrogen. This theory was confirmed as there was no diffusible hydrogen whatsoever released during this testing.

Table 3 Levels of diffusible and residual hydrogen measured for ASS and ferritic WM deposited on the investigated steel and also ASS WM deposited on 304 ASS. Note: Presented levels of diffusible hydrogen are the average values of three trials and residual WM's hydrogen levels the average of nine samples produced under the same conditions.

Consumable	Diffusible H content [ml/100g of WM]	Residual H content [ml/100g of WM]
Austenitic	1.3	5.8
Ferritic	23.4	0.9
Austenitic (PM+WM)	0	8

To demonstrate that hydrogen is also present during welding with the austenitic consumable but is safely "locked" in the weld metal, the previous diffusible hydrogen test samples were sectioned and the weld metal was subjected to residual hydrogen testing. The value of 5.8 ml of hydrogen per 100g in the austenitic weld metal indicates that the hydrogen is indeed "locked" within in the microstructure. Whereas the low value of 0.9 ml/100g in the ferritic weld metal indicates that the majority of hydrogen has diffused out of the weld / into the HAZ.

Effect of preheat temperature on hydrogen content

A further objective of the project was to justify reducing or possibly eliminating the recommended preheat temperature of 80°C when welding this type of high strength steel joint when using an austenitic consumable. To be able to compare the influence of various levels of preheat on diffusible hydrogen content, samples were allowed to cool down to 150°C after welding and prior to quenching. Measured levels of diffusible and residual hydrogen of the samples welded with an austenitic filler wire at three different preheat temperatures are shown in table 4. The results show that a decrease in the preheat temperature from 80°C to 7°C resulted in an increased level of diffusible hydrogen from 0.5 to 1.3 ml/100g but decreased the level of weld metal residual hydrogen content from 8.0 to 6.1 ml/100g. The increase in diffusible hydrogen content with decreasing preheat temperature is probably the result of reduced cooling time in the temperature range associated with highly effective hydrogen diffusion out of the weldment.

Table 4 Levels of diffusible and residual hydrogen measured for ASS WM deposited on the investigated steel PM under different PH conditions. Note: Presented levels of diffusible hydrogen are the average values of three trials and residual WM's hydrogen levels the average of nine samples produced under the same conditions.

PH temperature [°C]	Diffusible H content [ml/100g of WM]	Residual H content [ml/100g of WM]
7	1.3	6.1
Ambient (22)	0.8	7.9
80	0.5	8.0

The austenitic weld structure promotes hydrogen diffusion from HAZ to weld metal where it becomes effectively trapped. In other words the austenitic weld metal microstructure having a high solubility and very low diffusivity of hydrogen acts like hydrogen sink. Therefore reduced cooling times (less time for evolution of hydrogen from HAZ into the weld metal) as a consequence of lowering the preheat temperature resulted in lower levels of weld metal residual hydrogen content.

Comparison of hydrogen levels for three types of austenitic consumables

The levels of diffusible and residual hydrogen contents measured for three different types of austenitic wires namely flux-cored, metal-cored and solid wire are displayed in table 5. As expected the highest levels of both diffusible and residual hydrogen were present when employing flux-cored wire. This is due to the presence of the rutile flux, probably the major source of hydrogen contamination. Metal-cored weld deposits resulted in 38% and 42% reduction of diffusible and residual hydrogen respectively. Very low levels of hydrogen both diffusible and residual were expected due to the absence of the flux and seamless structure of external metallic sheath of the wire eliminating hydrogen pick-up by the inner core. Samples produced employing solid wire did not release any diffusible hydrogen at all and the residual hydrogen content was reduced by 53% and 20% compared to flux-cored and metal-cored wires respectively.

Table 5 Values of diffusible and residual hydrogen levels measured for three different types of ASS consumable deposited on investigated stel. Note: Presented levels of diffusible hydrogen are the average values of three trials and residual WM's hydrogen levels the average of nine samples produced under the same conditions.

ASS consumable type	Diffusible H content [ml/100g of WM]	Residual H content [ml/100g of WM]
Flux-cored	0.8	7.9
Metal-cored	0.5	4.6
Solid	0	3.7

Conclusions

From this study it may be concluded that:

1. The use of ferritic consumable with rutile flux type can introduce high levels of hydrogen into the weldment. The low solubility and high diffusivity of hydrogen in the ferritic lattice promotes the transfer of hydrogen into the susceptible HAZ on cooling.

2. When an austenitic consumable is used the majority of the hydrogen introduced into the weldment is contained in the weld metal and does not diffuse into the HAZ or surrounding atmosphere. On the contrary diffusible hydrogen evolves from the HAZ, (which with decreasing temperature becomes heavily oversaturated), into the surrounding atmosphere or into the weld metal.

3. The entire diffusible hydrogen measured in the samples with austenitic weld deposits originated from HAZ.

4. The diffusible hydrogen content of the samples welded with austenitic consumables increased with decreasing preheat temperature. Residual hydrogen levels however increased with increasing preheat temperature indicating that on cooling hydrogen diffuses from supersaturated HAZ into the weld metal.

5. Lowering the preheat temperature from 80°C to 7°C for welding these high strength steels with austenitic consumable resulted in slightly higher level of diffusible hydrogen, but diffusion of this hydrogen is promoted from the HAZ to the weld metal, greatly reducing the risk of HACC compared to a ferritic consumable.

6. Both the diffusible and residual hydrogen levels were decreased when moving from flux-cored to metal-cored and solid austenitic consumable. With zero diffusible hydrogen measured for solid ASS consumable deposited onto investigated steel.

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