

Welding of Hydrogen-Charged Steel for Modification or Repair

The effect of hydrogen in base metal is confirmed, and advice is provided on welding C-Mn and Cr-Mo steel that contains hydrogen

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ABSTRACT

Various types of steel equipment, particularly in refinery service, absorb hydrogen during operation. Material selection and design should ensure that this does not cause any damage, but the presence of hydrogen in the steel also needs to be taken into account if modifications or repairs involving welding are required. Hydrogen in the steel will contribute to hydrogen in the weld and may increase the risk of fabrication hydrogen cracking (cold cracking). The usual approach is to impose a hydrogen removal heat treatment, or hydrogen bakeout prior to welding, to ensure that there is no significant hydrogen left in the steel, in which case normal welding precautions can then be taken to avoid cracking. This, however, is a costly and time-consuming process, with the time taken making a direct and particularly significant contribution to the cost if it results in extended downtime of a refinery. The effects of hydrogen in the base steel have been explored in an experimental program of work, and recommendations for safe welding procedures have been made.

Introduction

When faced with a requirement to weld on hydrogen-charged steel, the welding engineer will typically perform some sort of calculation to determine the time and temperature required to remove the hydrogen. For example, the hydrogen re-

moval curves presented in Ref. 1 may be used. Any such calculation will require a knowledge of the coefficient of diffusion of hydrogen for the steel of concern, which is generally not known with any confidence, and some judgment will have to be made with regard to the amount of hydrogen that needs to be removed. One solution to these uncertainties is to employ direct measurement of hydrogen effusion, and successful use of such an approach has been reported (Ref. 2). Nevertheless, both downtime and uncertainty would be removed if the effects of hydrogen in the steel could be accommodated in the welding procedure. Welding conditions that avoid hydrogen cracking can be devised for high-hydrogen consumables, and a similar approach should be possible for a hydrogen-charged steel.

A further advantage of devising welding procedures that accommodate hydrogen in the steel is that trapped hydrogen, which may not be driven out by relatively low-temperature bakeout treatments or, therefore, registered by direct hydrogen flux measurements, would be taken into account. There is some risk associated with assuming that a hydrogen bakeout treatment has been completely effective by calculation or measurement.

Bearing the above considerations in mind, a program of experimental work was carried out at TWI. The aim was to demonstrate and quantify the effect of hydrogen in the steel such that appropriate modifications to welding procedures could be recommended.

KEYWORDS

Controlled Thermal Severity (CTS)
Hydrogen
C-Mn Steel
Cr-Mo Steel
Heat-Affected Zone (HAZ)

Experimental Program

Approach

The issue of hydrogen charging of steels in service arises both in low-temperature corrosive conditions and when handling hot, high-pressure hydrogen. For corrosive conditions, the primary concern is with carbon manganese (C-Mn) steels, particularly when operating in sour (H₂S-containing) environments. For hot, high-pressure hydrogen, chromium molybdenum (Cr-Mo) alloyed steels resistant to hydrogen attack are used. In view of both the significant differences in the materials and the hydrogen charging routes, the work included tests on corrosively charged C-Mn steel and Cr-Mo steel charged in a hydrogen autoclave.

The controlled thermal severity (CTS) test (Ref. 3) was selected as a weldability test to provide a comparison between hydrogen-charged and hydrogen-free steels. Data generated using this test at TWI formed the basis of the guidelines currently in BS EN 1011-2, Appendix C (Ref. 4), and in the TWI book on welding steels without hydrogen cracking (Ref. 1). Thus, it was considered to be an appropriate test that would allow the data to be fed directly into guidelines for avoidance of cracking. It was also possible to carry out hydrogen charging and analysis with only slight modification of the test method. To facilitate hydrogen analysis, extension pieces were included on all test blocks, which were removed for analysis immediately prior to test welding.

Bearing in mind the common constraints on heat input in a repair weld situation, 'crack: no crack' boundaries were determined in terms of preheat at a single heat input typical of repair practice.

Materials

The test steels consisted of three C-Mn steels and one Cr-Mo steel. Chemical

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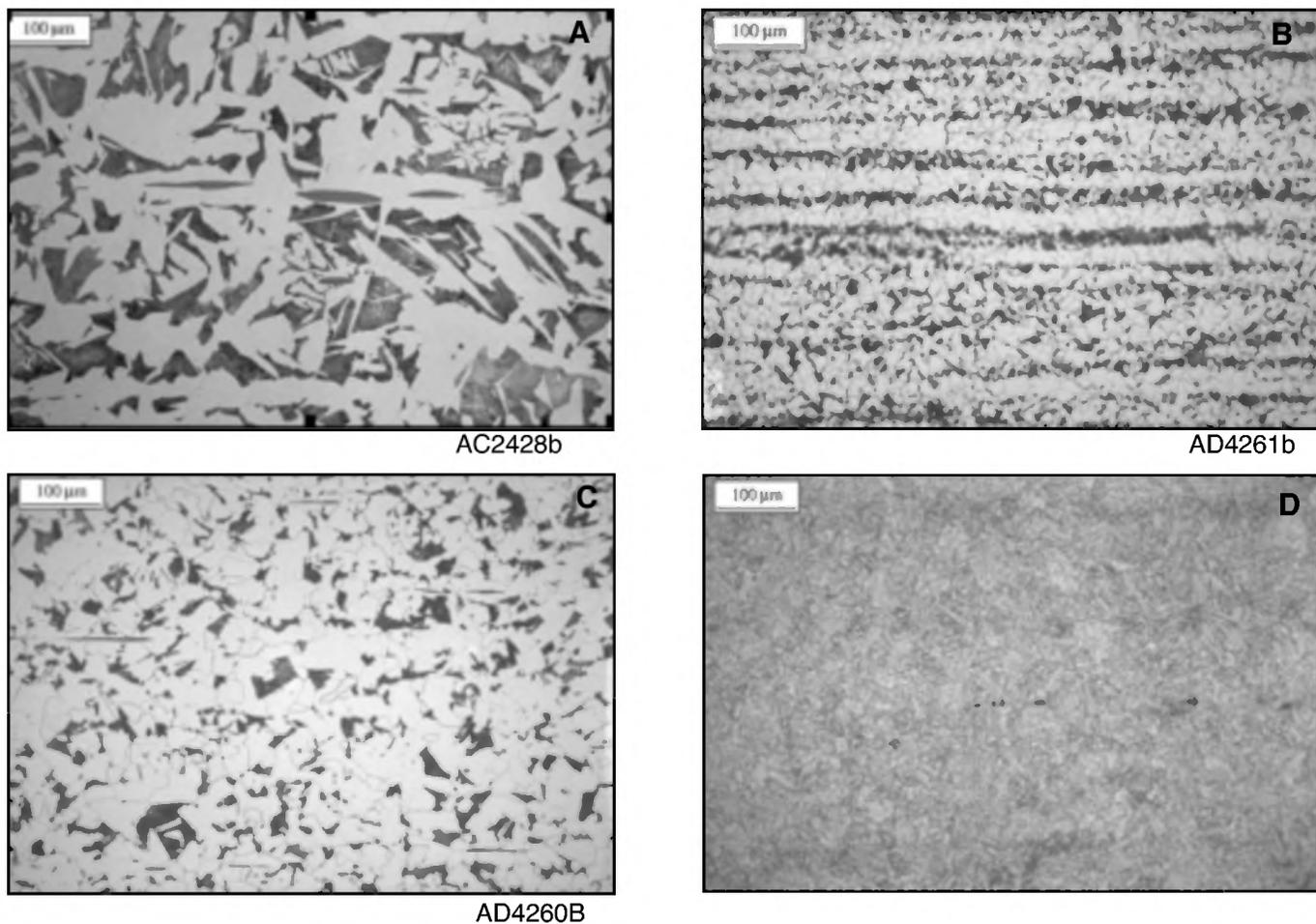


Fig. 1 — Micrographs of test steels. All etched in 2% nital. Magnifications given by micron marks. A — High CE, high S, C-Mn steel (A); B — high CE, low S, C-Mn steel (B); C — low CE, high S, C-Mn steel (C); and D — 2¼ Cr-1 Mo steel (D).

compositions are presented in Table 1. Micrographs are also presented — Fig. 1.

Three C-Mn steels were used to investigate the effects of two material variables. The steels selected cover high and low carbon equivalents (CE, see Table 1), the parameter that best describes this class of material's relative susceptibility to fabrication hydrogen cracking. The selected steels also cover both high and low sulfur contents.

The higher sulfur level was selected to allow the effect of hydrogen trapping around inclusions (which might slow diffusion of hydrogen in the steel) to be explored.

A high CE, high sulfur C-Mn steel (A) was chosen as representative of older, dirty (high sulfur and oxygen) steels likely to be encountered in repair welding operations. It conformed to the old British Standard BS 1501-221 Grade 32A and had

a CE of 0.43 wt-% and a sulfur content of 0.032 wt-%.

A steel with the same CE (steel B), but with a very low sulfur content of <0.002 wt-%, provided a comparison in terms of steel cleanliness. Furthermore, this was the only Al-treated steel of the three C-Mn steels tested, and it had a low oxygen content of 4 ppm. This steel complied with BS 4360: 1990 Grade 50D. The grain size

Table 1 — Chemical Compositions of Base Materials (wt-%)

	C	S	P	Si	Mn	Ni	Cr	Mo	V	Cu	Nb	Ti	Al	O	N	CE*
High CE, high S C-Mn steel, A	0.21	0.029	0.018	0.22	1.12	0.08	0.07	0.02	<0.002	0.12	<0.002	<0.002	<0.003	0.0121	0.0062	0.43
High CE, low S C-Mn steel, B	0.19	<0.002	0.021	0.28	1.38	0.01	0.02	<0.005	<0.002	0.005	0.024	<0.002	0.047	0.0004	0.0041	0.43
Low CE, high S C-Mn steel, C	0.14	0.033	0.035	0.20	1.25	0.04	0.03	0.005	<0.002	0.02	<0.002	<0.002	0.004	0.0100	0.0038	0.36
2¼ Cr 1Mo steel, D	0.14	0.002	0.004	0.20	0.43	0.10	2.17	0.96	0.002	0.02	<0.002	<0.002	0.024	0.0024	0.0034	—

* $CE_{IIW} = C + Mn/6 + (Cr + Mo + V)/5 + (Ni + Cu)/15$

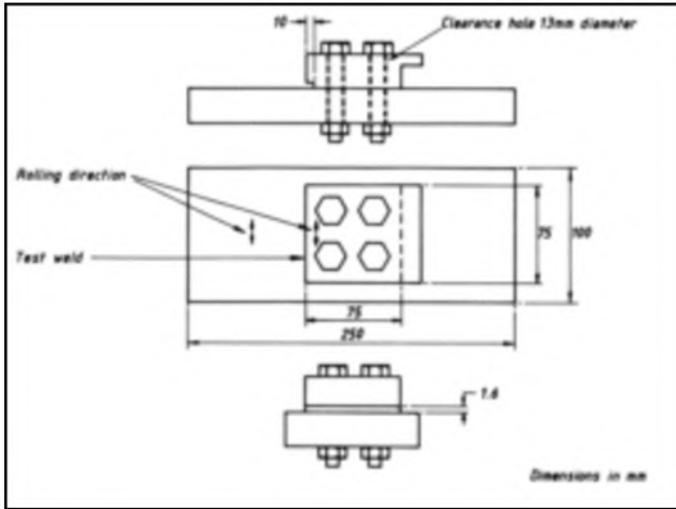


Fig. 2 — Modified CTS test design.

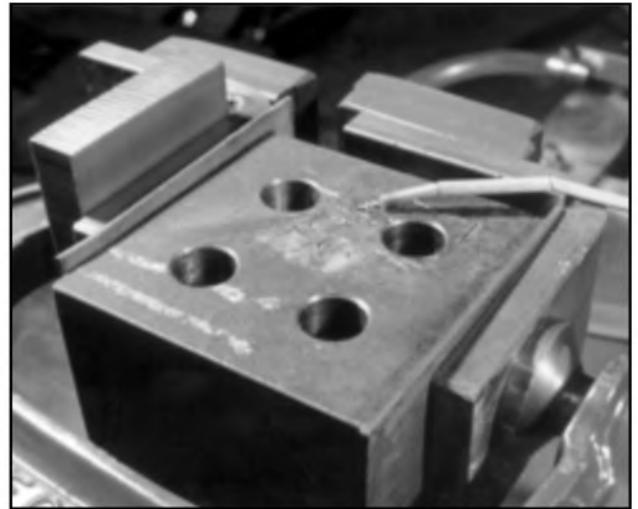


Fig. 3 — Modified CTS top block setup for preheating by electrical resistance.

was finer than that of steel A, but it was not unusual for a steel of this grade.

Steel C was another high sulfur (0.033 wt-%) steel to BS 1501-121 but with a lower CE of 0.36 wt-%. The grain size was relatively coarse and closer to that of steel A

than steel B.

The Cr-Mo steel (B) was a 2.25Cr 1Mo type, with a tempered bainitic microstructure, and low sulfur and oxygen contents.

The manual metal arc/shielded metal arc process was selected, and part dried

welding consumables were procured that could be conditioned to the desired hydrogen level. For C-Mn steels, AWS E7018 consumables were used. These were dried at 330°C for 1 h to give 9.7 mL/100 g deposited metal hydrogen, which is toward

Table 2 — Test Results, C-Mn Steel, A

Weld No.	Heat Input* kJ/mm	Preheat °C	HAZ Hardness HV10	No. of Faces Showing Cracking	Hydrogen, mL/100 g in Top Block	
			$\frac{\text{max-min}}{\text{mean}}$		Diffusible	Total
W2	0.83	20	$\frac{425-390}{406}$	3/6	Not charged	
W3	0.83	100	$\frac{376-354}{365}$	0/6	Not charged	
W4	0.86	50	$\frac{429-360}{400}$	0/6	Not charged	
W6	0.82	38	$\frac{429-401}{415}$	0/6	Not charged	
W8	0.88	85	$\frac{383-357}{377}$	0/6	6.3	6.4
W9	0.88	35	$\frac{421-401}{409}$	0/6	Not charged	
W11	0.78	60	$\frac{394-380}{386}$	1/6 inclusion cracking 3/6 conventional	5.9	6.1
W12	0.81	35	$\frac{394-376}{385}$	5/6 inclusion cracking 2/6 conventional	8.5	8.7
W13	0.82	77	$\frac{401-376}{391}$	6/6 inclusion cracking 0/6 conventional	7.4	7.6
W15	0.88	100 top 95 base	$\frac{376-357}{365}$	3/6 inclusion cracking 0/6 conventional	9.2	9.9
W17	0.72	120	$\frac{366-342}{349}$	5/6 inclusion cracking 0/6 conventional	12.0	12.1
Not welded	—	As charged	—	—	14.0	14.2
Not welded	—	150	—	—	9.5	9.7

* Using arc efficiency of 0.8.

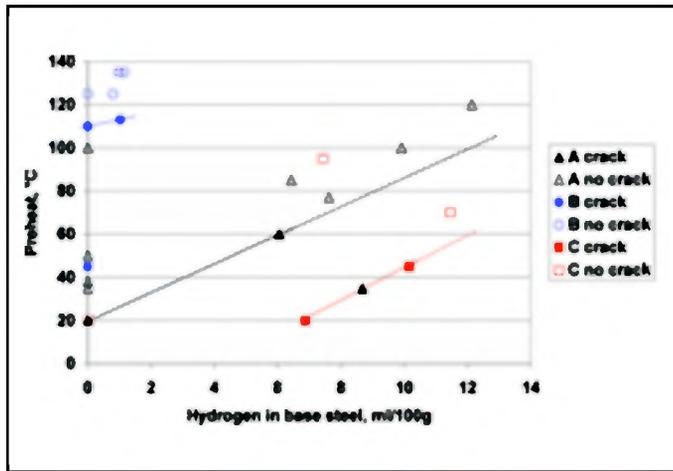


Fig. 4 — Summary of results of CTS tests on C-Mn steels A, B, and C.

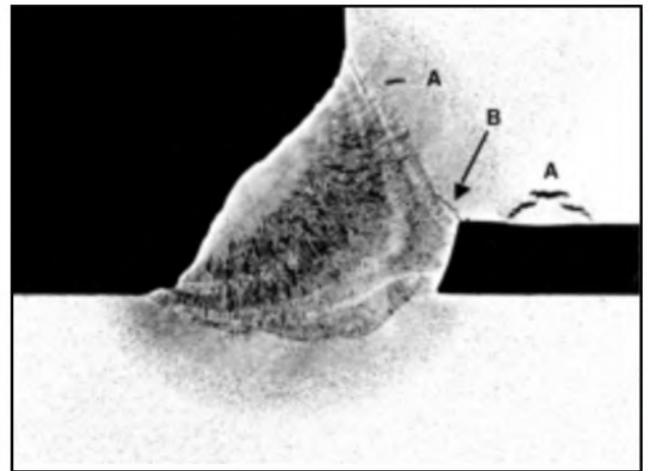


Fig. 5 — Inclusion cracking (A) and HAZ hydrogen cracking (B) in sample W12 (hydrogen-charged, high sulfur C-Mn steel).

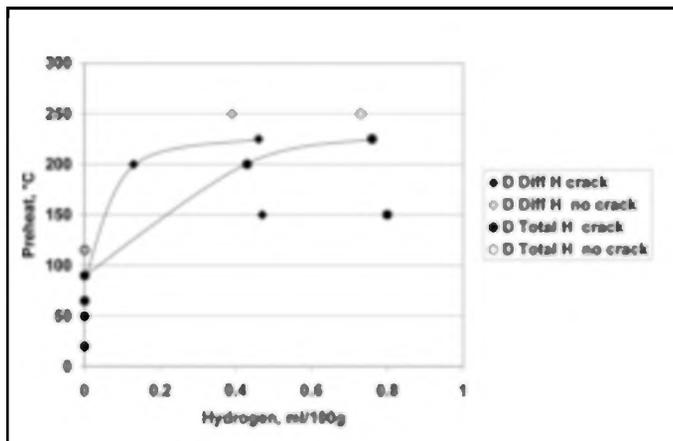


Fig. 6 — Summary of results of CTS tests on 2 1/4Cr-1Mo steel D.

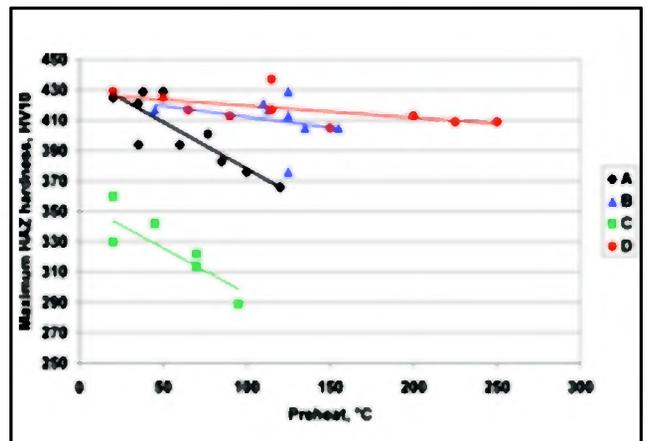


Fig. 7 — Effect of preheat on maximum hardness for the different steels.

the upper end of scale C in BS EN 1011-2 (5–10 mL/100 g deposited metal). For the Cr-Mo steel, matching AWS E9018 consumables were used, dried at 250°C for 1 h to give 4.3 mL/100 g hydrogen, which is toward the upper end of scale D (3–5 mL/100 g deposited metal). The aim was to generate cracking conditions with reasonable preheat/heat input combinations so that any effect of base metal hydrogen on these thresholds could be determined.

Hydrogen Charging of C-Mn Steels

Hydrogen charging of the C-Mn steels was achieved by placing the modified CTS top blocks (see below) in acidified (pH3) 5 wt-% NaCl solution saturated with H₂S (standard NACE TM0177 solution A (Ref. 5)), with an exposure time of 96 h. After charging, the blocks were removed from the solution, lightly cleaned, then packed in dry ice and stored in a freezer

until required for testing. Results of diffusible and total hydrogen determinations on the tabs from charged CTS top blocks are included in Tables 2–4. Diffusible hydrogen analysis was performed by full evolution and collection over mercury at ambient temperature and residual hydrogen (to give total hydrogen by addition) by vacuum hot extraction at 650°C.

Hydrogen Charging of Cr-Mo Steel

Hydrogen charging of the Cr-Mo steel was carried out using an autoclave. In this case, CTS top blocks were exposed to a hydrogen atmosphere at a temperature of 450°C and pressure of 10.3 MPa (1500 lb/in.²) for 48 h. This temperature was chosen as it is approximately that at which many vessels operate, and the pressure places this condition just within the safe region of a Nelson curve plot for this material. The exposure time of 48 h was estimated to be suf-

ficient to saturate the material under these conditions. Previous published work (Ref. 6) indicates that a hydrogen content of approximately 4 ppm should be obtained (1 ppm = 1.12 mL/100 g). The same source quotes a calculated hydrogen content of 4.8 ppm for an α iron under similar conditions.

After charging, the Cr-Mo blocks were stored in the same way as the C-Mn blocks. However, in this case some loss of hydrogen is expected between charging and storage because the blocks must cool to 250°C before pressure can be released and the blocks extracted and quenched. This can take up to 1 h.

Results of diffusible and total hydrogen determinations on the tabs from charged CTS top blocks are included in Table 5. Diffusible hydrogen analysis was performed by evolution and collection over mercury at ambient temperature and residual hydrogen (to give total hydrogen by addition) by vacuum hot extraction at 650°C.

Table 3 — Test Results, C-Mn Steel, B

Weld No.	Heat Input* kJ/mm	Preheat °C	HAZ Hardness HV10 $\frac{\text{max}-\text{min}}{\text{mean}}$	No. of Faces Showing Cracking	Hydrogen, mL/100 g in Top Block	
					Diffusible	Total
W27	0.83	45	$\frac{417-405}{412}$	6/6 conventional	Not charged	
W28	0.88	110	$\frac{421-405}{414}$	4/6 conventional	Not charged	
W29	0.82	125	$\frac{413-345}{376}$	0/6	Not charged	
W31	0.83	125	$\frac{429-413}{419}$	0/6	Not charged	
W32	0.83	160 top 150 bottom	$\frac{405-366}{390}$	0/6	Not analyzed	
W34	0.82	135	$\frac{405-387}{393}$	0/6	0.8	1.0
W35	0.85**	135	$\frac{405-373}{395}$	0/6	1.0	1.2
W37	0.88	125	$\frac{376-357}{370}$	0/6	0.7	0.8
W40	0.81	110–115	$\frac{417-394}{403}$	1/1 conventional	0.9	1.0

* Using arc efficiency of 0.8.

** Estimated using $V=23$ because true voltage reading unavailable.

CTS Tests

CTS testing was performed as far as possible to BS EN ISO 17642-2:2005 (Ref. 3), but slightly modified to take account of the special requirements of this work. All welding was performed using 4-mm-diameter electrodes at a nominal heat input of 0.8 kJ/mm (arc energy of 1 k/mm), representative of likely practice for repair welds. Specially designed robotic equipment was used, and travel speed and arc length were controlled automatically. Current and voltage were monitored and recorded throughout. Tests were carried out at different preheats to define 'crack: no crack' boundaries. The threshold preheat is defined as the highest preheat for which cracking was observed, established to within 25°C of a no crack result and confirmed with a second no crack result.

Test Block Design

The test assembly itself was modified so that restraint was provided by four bolts (tightened to 100 Nm torque) rather than by anchor welds — Fig. 2. This was to avoid hydrogen loss from the hydrogen-charged CTS test assemblies during anchor welding. In addition, top blocks that were hydrogen charged were machined with small (10 × 12 × 75 mm) tabs on the

face opposite the weld. These were sawed off immediately prior to preheating and used for hydrogen determinations. The modified design was used for all tests to ensure consistent comparisons.

In order to establish the validity of this procedure, including uncharged baseline, hydrogen determinations were performed on tabs from two simultaneously charged blocks (steel A) before and after preheating to 150°C. The results of this trial are reported with the rest of those for steel A below.

Preheating Procedures

In order to minimize hydrogen loss during preheating, a procedure for rapidly heating hydrogen-charged top blocks and bolting these to the separately preheated bottom blocks was developed. The larger bottom blocks and bolts were brought to the required temperature by soaking in a furnace (1 h/in. thickness min). The hydrogen-charged top blocks were heated using an electric resistance heating method. The equipment used was a weld thermal simulator which, for the size of specimen used in this work, i.e., 50 mm CTS top block, was found to be capable of heating to a temperature of 180°C in 70 s — Fig. 3. During heating and subsequent assembly of the CTS components, tem-

perature was monitored with a portable digital thermometer and by taking readings from a Chromel-Alumel thermocouple located on the upper face of the top block.

Prior to carrying out the CTS tests, an investigation of temperature variations between this top face and the block center was performed. The results showed that when heating to 180°C, the temperature measured on the top face rose more quickly, exceeding the temperature at the center by up to 20°C. As the current was turned down (as 180°C was approached), the difference decreased. With the current off, a uniform temperature was reached after 10 s. The temperature measured on this upper face was, therefore, taken as being representative of the whole block.

Once at temperature, top blocks were transferred to the welding area and the CTS assembly bolted together ready for welding. This incurred a delay of approximately 7 min. During trials, an assembly heated to 180°C was found to cool to 160°C in this time (obviously heat loss will be less at lower preheat temperatures). The temperature quoted for individual tests is the average temperature of the top and bottom blocks immediately prior to welding measured using a contact thermometer. Following welding, water cooling was applied as per BS EN ISO 17642-2:2005.

Table 4 — Test Results, C-Mn Steel, C

Weld No.	Heat Input* kJ/mm	Preheat °C	HAZ Hardness HV10 <u>max–min</u> mean	No. of Faces Showing Cracking	Hydrogen, mL/100 g in Top Block	
					Diffusible	Total
W20	0.78	20	<u>330–314</u> 322	0/6	Not charged	
W21	0.78	20	<u>330–306</u> 321	0/6	Not charged	
W22	0.85	20	<u>360–304</u> 342	2/6 conventional 2/6 inclusion cracking	6.8	6.9
W23	0.79	45	<u>342–327</u> 332	3/6 inclusion cracking 1/6 conventional HAZ	10.0	10.2
W24	0.86	70	<u>332–281</u> 303	1/6 inclusion crack into HAZ from base plate	11.3	11.5
W25	0.81	70	<u>314–302</u> 309	2/6 inclusion cracks	11.4	11.5
W26	0.90	95	<u>289–283</u> 285	2/6 inclusion cracks	7.3	7.4

* Using arc efficiency of 0.8.

Examination of CTS Test Welds

After welding, sectioning and examination of the CTS assemblies was also carried out to BS EN ISO 17642-2:2005 following a delay of at least 72 h. Metallographic sections were taken as required, polished to a 3- μ m finish, etched in 2 wt-% nital, and then examined for cracking in the top block heat-affected zone (HAZ) using an optical microscope. Following this, hardness surveys were performed using a Vickers hardness machine with a load of 10 kg.

Results

High-Carbon-Equivalent, High-Sulfur, C-Mn Steel (A)

A ‘crack: no crack’ preheat threshold of 20°C (room temperature) was established for this steel in the as-received condition (at a heat input of 0.8 kJ/mm and with a weld metal diffusible hydrogen level of 9.7 mL/100 g). CTS test results for this steel are presented together with hardness survey results in Table 2. The threshold is lower than would be predicted by the nomogram in Ref. 1, apparently due to the low hardenability of this steel, as discussed in the next section.

The results of diffusible and residual hydrogen determinations performed on the tabs cut from hydrogen-charged top blocks are included in Table 2. These results show a wide scatter in hydrogen contents with a maximum total of 12.14 mL/100 g, a minimum total of 6.05 mL/100 g, and an average total of 8.47 mL/100 g in

the welded blocks. This variation was not expected considering previous experience (Ref. 7) with this charging method. Although greater variations in hydrogen pickup have been seen between different steels, this present level of scatter has not been seen with one steel. It is possible that this is due to the high inclusion content. Table 2 also shows the results of hydrogen determinations on tabs removed before and after preheating. These seem to indicate that for the two blocks in question, the hydrogen level measured in a tab is reduced during heating to 150°C and subsequent cooling. However, two separate blocks were used for this trial, and the differences between the two may just be experimental scatter in light of the other results for this steel. It is not, therefore, thought possible to draw any firm conclusions from this trial.

Results of CTS tests on this material after hydrogen charging are shown in Table 2 and included on a graph — Fig. 4. It can be seen that the hydrogen content (average 8.47 mL/100 g) of the charged base material has increased the preheat threshold by approximately 40°C.

In this work, the ‘crack: no crack’ preheat threshold for charged material has been defined by the occurrence of conventional HAZ-type cracking. No weld metal cracking was observed. However, cracking associated with blistered inclusions was also observed, a photomicrograph showing typical examples of both types of crack is shown in Fig. 5. Examination of unwelded hydrogen-charged material has shown that such cracking can exist even before preheating and, therefore, al-

though its occurrence has been noted, it has not been used to define ‘crack: no crack’ thresholds. Nevertheless, this type of inclusion cracking did seem to be exacerbated by preheating when preheated but unwelded blocks were examined, and it was also noticeably worse in the HAZs of welded blocks. Although the preheating in this work was unusually rapid, similar effects principally due to an increase in hydrogen pressure at preheat temperatures would be expected unless heating rates were very slow, as hydrogen escape from voids at such temperatures is very slow.

High-Carbon-Equivalent, Low-Sulfur, C-Mn Steel (B)

Controlled thermal severity results for this steel in the as-received condition and after hydrogen charging are shown in Table 3. Hydrogen determinations on the tabs from charged specimens are also included in Table 3. The results for total hydrogen content show the maximum to be 1.15 mL/100 g, the minimum 0.80 mL/100 g, and the average 0.99 mL/100 g. It can be seen from these results that the level of hydrogen, particularly diffusible hydrogen, in this steel is substantially lower than for the higher-sulfur steel A.

All the CTS data for this steel are included in Fig. 4. It can be seen from this graph that preheat thresholds are effectively the same in both conditions, at 110°C, and both are significantly higher than for the high-sulfur steel A that has the same CE. In the as-received condition, this is not unexpected considering the difference in sulfur contents between the two

Table 5 — Test Results, 2½Cr-1Mo Steel, D

Weld No.	Heat Input* kJ/mm	Preheat °C	HAZ Hardness HV10 <u>max-min</u> mean	No. of Faces Showing Cracking	Hydrogen, mL/100 g in Top Block	
					Diffusible	Total
W5	0.70	20	<u>429-397</u> 410	6/6	Not charged	
W7	0.78	115	<u>417-401</u> 411	0/6	Not charged	
W10	0.85	90	<u>413-401</u> 405	1/6	Not charged	
W14	0.75	65	<u>417-409</u> 413	4/6	Not charged	
W16	0.76	115	<u>437-409</u> 422	0/6	Not charged	
W18	0.76	50	<u>425-405</u> 411	6/6	Not charged	
W30	0.86	150	<u>405-401</u> 403	4/6	0.5	0.8
W33	0.87	200	<u>413-390</u> 403	2/6	0.1	0.4
W36	0.96	250	<u>409-401</u> 406	0/6	0.4	0.7
W39	0.90	225	<u>409-401</u> 404	2/6	0.5	0.8
W41	0.88	250	<u>409-397</u> 402	0/6	Not analyzed	

* Using arc efficiency of 0.8.

Table 6 — Summary of Effects of Hydrogen on Necessary Preheat to Prevent Hydrogen Cracking

	High CE (0.43) High S (0.029%)	High CE (0.43) Low S (<0.002%)	Low CE (0.36) High S (0.033%)	2½ Cr 1Mo
Average total hydrogen in steel, mL/100 g	8.5	1.0	9.5	0.7
Increase in preheat after charging	40°C*	Zero*	>25°C*	130°C**

* Heat input of 0.8 kJ/mm and 9.7 mL/100 g consumable hydrogen.

** Heat input of 0.8 kJ/mm and 4.3 mL/100 g consumable hydrogen.

steels and the previously reported observation that high sulfur can reduce relative cracking susceptibility (all other things being equal) (Ref. 8). Critical hardnesses for the two steels are, however, very close in the as-received condition at 425 HV10 for the high-sulfur and 421 HV10 for the low-sulfur steel (Tables 2, 3), consistent with the idea that high-sulfur alleviates

cracking risk via its effect on hardenability. As expected, no weld metal cracking and no blistered inclusion-type cracking were observed in this low-sulfur steel.

Low-Carbon-Equivalent, High-Sulfur, C-Mn Steel (C)

Controlled thermal severity test results

for this steel, in both the as-received and hydrogen-charged condition, are given in Table 4 and included on Fig. 4. Results of hydrogen determinations after charging are also included in Table 4 showing a variation in total hydrogen content from 6.88 to 11.47 with an average of 9.48 mL/100 g. These results are similar to those for the other high-sulfur steel A.

No cracking was observed in tests carried out at room temperature with the material in the as-received condition. In the hydrogen-charged condition, however, a preheat threshold of 45°C was established. This equates to a shift of >25°C by comparison with the as-received steel.

Blistered inclusion-type cracking was also observed in this steel but not used to define the threshold as with the other high-sulfur steel A. There was no weld metal cracking.

2.25Cr 1Mo Steel (D)

Controlled thermal severity test results for as-received and hydrogen-charged material welded at 0.8 kJ/mm heat input and weld metal hydrogen levels of 4–5 mL/100

g are given in Table 5. Results of hydrogen determinations on charged material are also included in Table 5. These results indicate a very small amount of diffusible hydrogen present in the tabs (0.36 mL/100 g). The extent to which this is representative of the charged blocks as a whole is unknown, but it would be expected that hydrogen loss during cooling in the autoclave immediately following charging will be greater from the tab because of its larger surface area to volume ratio. The average total hydrogen content was 0.68 mL/100 g.

Despite the low measured hydrogen content, the effect of charging on the preheat threshold was very marked as shown — Fig. 6. The shift in threshold preheat being approximately 130°C from 90°C for the as-received to 225°C after hydrogen charging. All cracking was conventional HAZ root cracking.

Discussion

C-Mn Steels

It was not intended to vary charged hydrogen level in this program. The original intention was to determine any increase in preheat necessary to prevent hydrogen cracking for the three C-Mn steels after hydrogen charging, assuming the hydrogen charging had resulted in a constant material hydrogen level. The results of the program are presented in these terms in Table 6.

For C-Mn steels containing high levels of sulfur, hydrogen charging using the NACE TM0177 solution A (Ref. 5) results in total base material hydrogen contents of approximately 9 mL/100 g, and this in turn increased the preheat necessary to prevent cracking by approximately 25–40°C. The increases in preheat necessary for both the high and low CE steels equate roughly to what would be expected on increasing consumable hydrogen level to scale B from the scale C level actually used. This indicates that, at this level of hydrogen in base material, the effect of hydrogen introduced by welding cannot be considered in isolation or assumed to swamp any base material hydrogen effects.

The lack of any observed increase in cracking following hydrogen charging of the low-sulfur steel (B) is probably due to the low hydrogen content, this being sufficiently low that any effect it may have is masked by the hydrogen from the consumable.

As noted above, the hydrogen charging did not result in consistent levels for all the C-Mn steels, and in some cases there was less consistency within one steel type than was expected. The scatter in levels attained for the three C-Mn steels makes it possible to plot a graph of total hydrogen

content vs. preheat. This graph has been presented in Fig. 4 (note that it has been assumed that as-received materials contain no hydrogen). From this graph, the differences in behavior between the three types of C-Mn steel can be seen more clearly. In addition, the need for data at other hydrogen levels or, in the case of the high-CE low-sulfur steel (B) at a lower consumable hydrogen level, can be seen.

For the low-sulfur steel B, the reported results show considerably lower diffusible hydrogen content than the higher-sulfur steels but a roughly comparable residual hydrogen content. Other unpublished work at TWI (Ref. 9) confirms this trend. In a project looking at the effect of hydrogen on fracture toughness of steels, a C-Mn steel of intermediate sulfur content (0.015 wt-%) was charged in an identical manner. Hydrogen determinations revealed a diffusible hydrogen content of 3.97 mL/100 g and a residual content of 0.1 mL/100 g. Thus, it appears that diffusible hydrogen content rather than residual hydrogen content is a function of sulfur content. This is perhaps surprising as it would be expected that more sulfide inclusions would provide more deep traps and result in a greater residual hydrogen content. It is, however, thought that the split between diffusible and residual hydrogen may be misleading, as it is possible that some hydrogen present in inclusion voids may be able to diffuse out even at room temperature over the time period of the measurements. It is also worth noting that only the low-sulfur steel was aluminum treated.

The effect of hydrogen in causing blistered inclusion-type cracking with these high-sulfur steels must also be considered. Although this form of cracking was seen in two of the unwelded base materials (because of the severity of the corrosive charging medium and the quality of the steels) as well as in HAZs, it was observed that such cracking could also be induced or exacerbated in these steels by both preheating and welding. In view of this, it is thought that a less severe charging condition that does not result in cracking during charging could be used to elucidate this effect in any future work. It should also be recognized that preheat and/or welding could exacerbate this type of cracking in a repair situation.

In the absence of data over a wider range of consumable and precharged hydrogen levels, it is unwise to propose any comprehensive guidelines for the behavior of C-Mn steels in general. Nevertheless, this work has shown that base material hydrogen content does indeed have an exacerbating effect on the risk of fabrication hydrogen cracking, but that the possibility exists to overcome this without the need for preweld hydrogen release heat treatments. Specifically, it has been shown

that an increase in preheat of the order of 50°C when welding at scale C (5–10 mL/100 g) consumable hydrogen level is sufficient to prevent cracking at a heat input of ~0.8 kJ/mm if the base steel contains up to 12 mL/100 g hydrogen, as might be encountered following sour service.

2.25Cr 1Mo Steel

The effect of hydrogen charging was the most marked with this steel, although the measured hydrogen level was low (average total 0.68 mL/100 g). However, it is thought that these hydrogen levels will be underestimates of the hydrogen present in the actual top blocks owing to the short diffusion distances in the analysis tab. It has been estimated, using a simple diffusion model, that the hydrogen levels in the main body of the blocks are probably consistent with the hydrogen levels expected (approximately 4 mL/100 g).

The very marked effect of the base material hydrogen content (see Fig. 6) is again indicative of an add-on effect as observed with the high-sulfur C-Mn steels. In this case, the much greater increase in threshold preheat is probably a reflection of the difference in behavior between the Cr-Mo steel and C-Mn steels. For the former, preheat only decreases the likelihood of cracking by enhancing hydrogen diffusion, whereas with C-Mn steel preheat also influences HAZ hardness to a certain extent, rendering the HAZ microstructure, in turn, less sensitive to hydrogen. The effect of preheat on maximum HAZ hardness is shown — Fig. 7. With a bulk source of hydrogen available in the steel, hydrogen level control via preheat and hence weld cooling time is difficult, but at least in the size of block tested here, possible.

The implications of these results are twofold. First, although for a real repair situation the possibility does exist of preventing cracking through preheat control of weld cooling times, it is probable that this will be impractical for 2.25Cr 1Mo steel in most cases due to the magnitude of the change in cooling rate required. A more practical solution may be the application of postheat to enhance hydrogen diffusion.

A second implication is that any hydrogen release heat treatments on this type of Cr-Mo steel should be combined with conservative (i.e., high preheat) welding procedures as this work has shown that even a very small amount of hydrogen remaining in the base material will greatly increase the risk of cracking during welding.

Summary and Conclusions

An experimental procedure has been developed for CTS tests on hydrogen

charged base material. Tests have been performed, following this procedure, using a modified test block assembly. Results have enabled threshold preheats to be established for three C-Mn and one 2.25Cr 1Mo steel, at a heat input of 0.8 kJ/mm, both in the as-received and hydrogen-charged conditions.

Specific conclusions are as follows:

1. Hydrogen charging increases the risk of fabrication hydrogen cracking in both C-Mn and Cr-Mo steels. For C-Mn steels, this can be overcome by controlling weld cooling times through the application of preheat, negating the need for preweld hydrogen-release heat treatments.

2. The high-sulfur C-Mn steels of both high- and low-carbon equivalent required an increase in preheat to prevent cracking following charging to give hydrogen levels of up to 12 mL/100 g. This increase was of the order of 40°C for the high-carbon-equivalent steel and equal to or in excess of 25°C for the low-carbon equivalent steel when welded using a scale C consumable hydrogen level (5–10 mL/100 g deposited weld metal) at ~0.8 kJ/mm heat input.

3. For the high-sulfur steels, both conventional HAZ cracking and blistered inclusion cracking were observed to occur after welding in the hydrogen-charged condition. There is some indication that preheat and welding exacerbates blistered inclusion cracking.

4. No change in the preheat required to prevent cracking was observed for the low-sulfur, high-carbon-equivalent C-Mn steel following hydrogen charging under the same conditions as used for the high-sulfur steels. For the low-sulfur steel, the charged hydrogen levels were, however, significantly lower.

5. A significant increase of 130°C in preheat was required to prevent cracking in the 2.25Cr 1Mo steel following hydrogen charging in an autoclave at 450°C and 10.3 MPa pressure for 48 h.

Recommendations

General Recommendations for Welding Steels Containing Hydrogen

When welding either C-Mn or Cr-Mo steels containing or suspected of containing hydrogen, it should be assumed that this will result in an increased risk of fabrication hydrogen cracking during repair or alterations.

Recommendation for Welding C-Mn Steels Containing Hydrogen

For C-Mn steels that have been in sour service, the possibility exists of overcoming the increased risk of cracking in many circumstances by raising the preheat temperature.

In particular, if the steel to be welded has an IIW carbon equivalent of ≤ 0.45 wt-% and is to be welded at a heat input of ~0.8 kJ/mm with consumables of scale C, i.e., 10 mLH₂/100 g deposited metal (or lower) hydrogen content, then the increased tendency to crack can be compensated for by specifying a preheat of at least 50°C or 50°C in excess of that recommended in BS EN 1011-2 for a comparable hydrogen-free situation, whichever is the higher.

The use of consumables of scale D or lower is strongly recommended, but an assumed consumable hydrogen level of scale C should still be used when deciding upon an initial (hydrogen free) level of preheat. In situations where a heat input in excess of ~0.8 kJ/mm is to be used, it is recommended that the level of preheat be decided upon assuming the heat input is ~0.8 kJ/mm. If it is not possible to use a heat input of greater than or equal to ~0.8 kJ/mm, then the recommendations set out above cannot be fully relied upon to prevent cracking.

Recommendations for Welding Cr-Mo Steels Containing Hydrogen

For these types of steels, it appears unlikely that the increased risk of cracking can reliably be negated by the use of increased preheat, at least for those steels that have been in typical high-temperature hydrogen service.

In view of this, it is recommended that where such welding operations are to be undertaken, a preweld hydrogen-release heat treatment is performed and that this should be combined with conservative welding procedures, i.e., high preheat and low consumable hydrogen levels. Consideration should also be given to the use of a postheat to further reduce welding hydrogen levels.

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