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Cold Cracking Susceptibility of Weld Metal Deposited by Gas Shielded Rutile Flux-Cored Wire

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Cold cracking susceptibility of weld metal deposited by gas shielded rutile flux-cored wire

M. Pitrun and D. Nolan

This paper presents the results of an investigation of the hydrogen assisted cold cracking (HACC) susceptibility of low strength rutile flux-cored seamless (H5) and seamed (H10) wires, with nominal diffusible hydrogen (H_D) levels of 5 and 10 mL/100 g, respectively. The objective was to assess the influence of key welding parameters on the susceptibility of the weld metal to cold cracking. Parameters investigated were the welding current, the contact-tip to work-piece distance (CTWD), the shielding gas and the preheat temperature.

The gapped bead-on-plate (G-BOP) test was used to examine the effects of these parameters on weld metal transverse cracking. Tests were carried out at different preheat temperatures and the percentage of cracking was recorded.

It was found that, without preheat, the H5 wire weld deposits did not crack, whereas all those produced using the H10 wire exhibited cold cracking. The overall results indicate that the susceptibility to cold cracking strongly correlates with H_D. Apart from the intrinsic hydrogen content of the wire, the concentration in the deposit is affected by the welding current, the CTWD, the shielding gas and the preheat temperature. Preheat has a strong effect and was found to substantially decrease the amount of cold cracking in the H10 welds. Further, weld metal deposited using 75Ar-25CO_2 shielding gas resulted in higher H_D levels than for CO_2 shielding and a higher susceptibility to cold cracking for no or low preheat.

Keywords

FCAW, diffusible hydrogen, welding current, CTWD, shielding gases, G-BOP test, HACC, weld metal cracking, preheat temperature.

Introduction

Hydrogen assisted cold cracking can be initiated in either the parent metal HAZ or the weld metal when hydrogen is present in the welded joint and accumulates at a site of high stress concentration within a susceptible microstructure. Traditionally, processing factors such as preheat temperature, plate thickness, selection of welding process, welding consumable strength and nominal hydrogen content are chosen to avoid HACC in the HAZ of the parent plate, as specified in the welding standards.

Modern steels have become more resistant to HAZ hydrogen cracking as a result of reduced alloying content and the introduction of thermo-mechanically controlled processing (TMCP). The current generation of structural steels is characterized by leaner chemistry and more sophisticated thermo-mechanical processing, particularly lower carbon contents and the development of strength through grain size control and micro-alloying with strong carbide forming elements. The reduction in the carbon and carbon equivalent levels[1] has significantly lowered the risk of hydrogen cracking in the HAZ. As a result, the focus of attention has switched to the weld metal, particularly the development of transverse weld metal cracking in thick plate welds[2]. Consequently, it is becoming increasingly important to develop reliable testing methods that provide accurate data for the development of guidelines for the avoidance of weld metal hydrogen cracking.

Although there are guidelines and welding standards for avoidance of HACC in the HAZ [AS/NZS 1554.1-2000, AS/NZS 1554.4-1995, AWS D1.1-2000 and EN 1011.2-2001], a universal and reliable model for HACC avoidance in the weld metal is expected to be more complex and difficult than for hydrogen cracking in parent metal[3]. Therefore, independent management procedures for avoiding HACC in the weld metal are yet to be developed.

In general, the susceptibility of weld metal to hydrogen cracking appears to increase with an increase in weld metal strength, hydrogen content and section thickness[3,4] and is more complex than the case of HAZ cracking[5].

Weld metal hydrogen cracking transverse to the welding direction has been reported in a thick multi-pass weld FCAW welds using high strength[6] and low strength[7] rutile flux-cored wires. Interestingly, no cracking was observed in the HAZ in either work.

The aim of investigation reported in this paper was to analyse G-BOP test results for the FCAW process in the light of information previously reported on the effect of welding parameters on hydrogen content of weld metal generated by flux-cored wires[8]. The examination of two low strength rutile wires of the same classification, but different nominal hydrogen levels, has provided the opportunity to evaluate the effect of hydrogen on the HACC- susceptibility of low strength weld metal.

Weldability tests

The first test methods for cold cracking emerged in the 1940s[9], when the formation of martensite in the HAZ was the main cause of cracking. Following World War II, there was progressive development of hydrogen-induced cracking tests for a range of weld configurations and applications. These tests became gradually more sophisticated and some were designed specifically for the investigation of the mechanism of HACC and for the proper selection of welding materials and welding conditions

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Table 1. A list of reported test methods for determining hydrogen cracking in the parent metal HAZ and/or the weld metal.

TEST	Mode of Cracking		Weld Pass(es)
	HAZ	WM	
Reeve restraint cracking	x		S
Non-restraint fillet	x		S
Tekken (Y groove)	x		S
Controlled thermal severity (CTS)	x		S
Implant	x		S
Tensile restraint cracking (TRC)	x		S
Rigid restraint cracking (RRC)	x		S/M
H slit restraint cracking	x		S/M
Cruciform	x		M
Cranfield	x		M
Lehigh (U groove) restraint cracking	x	x	S
Lehigh (slot grove) restraint cracking	x	x	S
Welding Institute of Canada (WIC)	x	x	S
Circular patch (BWRA)	x	x	M
Longitudinal bead - tensile restraint (LB-TRC)		x	S
Longitudinal restraint cracking		x	S
V groove weld		x	M
Gapped bead on plate (G-BOP)		x	S/M

Note: S = single M = multiple

for its avoidance during weld fabrication. Historically, most of the methods were designed to simulate some particular application in which cracking was experienced. The main objective of weldability tests is to examine the effects of various factors on cracking susceptibility, including parent metal composition, type of welding consumable, preheat temperature and other welding conditions.

The basic idea of all testing methods is to obtain a reliable and representative indication of crack susceptibility in relation to a defined set of test criteria. Cold cracking tests are used to:

- examine sensitivity to welding variables and other surrounding conditions that affect hydrogen cracking;
- examine the relationship between welding consumable and parent metal;
- provide a preliminary examination of the cracking mechanism; and
- establish welding conditions that avoid or minimize hydrogen cracking for a particular given combination of welding process, consumable and parent metal.

In view of the crack location, testing methods for susceptibility to HACC are divided into two groups, those that study HACC in the HAZ or the weld metal. Although the earlier tests were developed primarily to measure susceptibility to HAZ cracking, several tests have been designed specifically for weld metal cracking, or both, as shown in Table 1.

The majority of these tests were designed as small scale laboratory tests, using a single weld pass. Other, more expensive weldability tests were designed for multi-pass welds that take into account the interacting effects of thermal cycles, changes in thermal stresses and increase in restraint associated with the progress of welding through the plate thickness. A number of studies have comprehensively reviewed the most commonly used weldability testing methods for HACC in both the HAZ and weld metal[5,10-12].

Although there are fundamental differences between the testing methods, particularly in terms of the different levels of restraint imposed, a number of cracking tests have proven to be sufficiently reliable that they have been accepted in American [API 4009-1977], British [BS7363-1990], French [NF A89-100-1991] and Japanese [JIS Z3158-1996] standards. The Lehigh, CTS, G-BOP, Implant and Tekken tests are the most widely adopted tests.

Development of the G-BOP Test

Transverse cracking can occur when welding over a small gap that acts as a stress concentrator. This situation can arise for poor fit up in highly restrained joints.

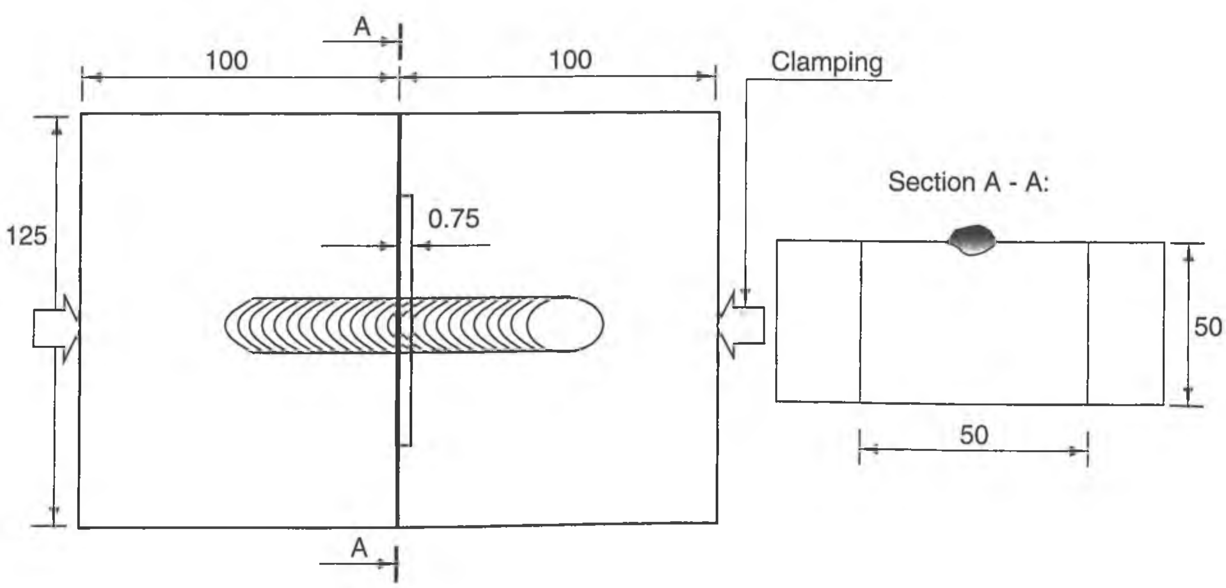

Figure 1. A diagram of the G-BOP test configuration (after Graville and McParlan, ref 15). Note: all dimensions in millimetres, [not to scale].

In the early 1960s, the Brown-Boveri test was introduced to examine cracking sensitivity. This test was initially designed for austenitic stainless steels and the sample consisted of several machined thick plates bolted together[13]. During the early 1970s, the E.O. Paton Welding Institute designed a test piece with an artificial notch that enabled initiation of a cold crack in a transverse direction to the weld[14]. Following this concept, a self-restrained gapped bead-on-plate test (G-BOP) was established[15], in which a gap underneath the weld introduces a large stress concentration assisting initiation of transverse weld metal cracking.

Early work on the G-BOP test by Graville and McParlan[15], initially applied to austenitic stainless steels, showed that the test was suitable for the determination of cold cracking susceptibility as a function of preheat temperature. The results indicated that by increasing preheat temperature, cracking was gradually suppressed. The mechanism of cold cracking was also elucidated through measurement of the stress level variation across the gap.

The G-BOP test sample consists of two 50 mm thick steel blocks, one of which has a machined 0.75 mm wide recess. The blocks are clamped together to prevent any relative movement and a bead is deposited along the top surface over the gap as shown in Figure 1.

After welding, the blocks are left in clamps for a minimum of 48 hours to provide necessary restraint and to allow hydrogen cracking to develop. The welds are then heated to a dull red heat in the vicinity of the gap to allow heat tinting of the fracture surfaces. The samples are then allowed to cool to room temperature and are broken open to reveal the fracture surfaces of the weld. Any cracks that developed in the weld metal during the dwell time of 48 hours are heat tinted, revealing a dark blue or gray discoloration of the fracture surface. Any non-cracked weld metal cross-section has as a light gray metallic appearance.

Several researchers later modified the test[12,16-19]. These modifications mainly included variations of test block dimensions, incubation periods, clamping forces and releasing time of clamps. Further modification of the testing procedure allowed rotation of test blocks to deposit 4 weld beads[12]. Although the G-BOP test method is primarily used to assess the susceptibility of the welding consumable to cold cracking, this method was also successfully used for a study of parent metal dilution in a weld metal[20].

In order to minimize the dilution effects, a modified G-BOP test has been developed to test the weld metal composition without the influence of dilution with parent metal. The plate is prepared by weld surfacing ('buttering') with the weld metal before machining of the test piece[18]. This technique is particularly applicable to study of hydrogen cracking in alloyed and multi-pass weld deposits[18].

The G-BOP test can be quantified by a room temperature cracking parameter (RTC), or a cracking parameter for preheat temperatures above 20°C.

However, for the case of consumables containing higher levels of diffusible hydrogen, RTC is usually 100% and the parameter is inadequate[18]. Therefore, a parameter known as the 10% crack preheat temperature (10% CPT), defined as the preheat temperature required to limit cracking to ≤10%, was found to be more suitable[17,18]. This parameter may be useful where two consumables produce similar amounts of diffusible hydrogen in their weld deposits and exhibit 100% RTC, but may respond differently to an increase in preheat temperature. That is, the 10% CPT values are different. Another useful parameter is the critical preheat temperature, obtained by extrapolation, at which the cracking percentage is expected to be <5%[12]. The major benefit of the G-BOP test is that it can be used as a quick and inexpensive 'go' or 'no-go' comparative method to rank consumables with respect to susceptibility to cold cracking. The standard procedure can be also enhanced by an instrumented G-BOP test. This enhancement can be achieved by recording temperature history and cooling rates, or longitudinal stresses across the gap during weld bead cooling[21].

The main aim of this current study was to observe the effect of preheat temperatures on susceptibility to cold cracking for a range of welding conditions in the FCAW process. Previous work by Pitrun et al.[8] established the levels of diffusible hydrogen in the weld bead for the same welding conditions and consumables, under controlled laboratory conditions.

Experimental procedure

Equipment and materials

Standard G-BOP tests were carried using the same welding equipment as for the diffusible hydrogen testing program previously reported[8]. A conventional 3-phase DC welding machine, Transmig 400, was used that has been widely adopted by industry for continuous gas-shielded wire processes (GMAW and FCAW). To allow full control of the welding parameters of travel speed, position of welding torch and CTWD, the welding torch was fixed onto a travelling mechanism mounted on the top of a support the frame. This enabled continuous horizontal movement under controlled conditions.

All of the G-BOP tests were conducted using the identical spools of wire that were used in the welding trials for the weld metal hydrogen testing[8]. Therefore, other than for the effects of varying atmospheric conditions (recorded for each set of test samples), the probability of significant variations in diffusible hydrogen levels between the two sets of results is low.

Table 2. Chemical compositions (weight %) of all-weld metal deposits of H5 and H10 FCAW consumables, using 75Ar-25CO_2 and CO_2 shielding gases.

All-weld metal chemical analysis of wire samples (weight %)																	
(H5)	C	Mn	Si	S	P	Ni	Cr	Mo	Cu	V	Nb	Ti	B	Al	N	O	CE_{IIW}	Pcm
75Ar-25CO_2	0.043	1.46	0.59	0.009	0.011	0.051	0.055	0.008	0.14	0.014	0.011	0.044	0.0043	0.008	0.0084	0.0430	0.31	0.173
CO_2	0.050	1.25	0.47	0.010	0.011	0.052	0.054	0.008	0.15	0.013	0.010	0.042	0.0032	0.007	0.0097	0.0590	0.29	0.159
(H10)																		
75Ar-25CO_2	0.070	1.41	0.61	0.011	0.012	0.025	0.027	0.003	0.029	0.015	0.012	0.051	0.0090	0.004	0.0042	0.0585	0.32	0.210
CO_2	0.065	1.15	0.48	0.012	0.012	0.024	0.025	0.002	0.025	0.015	0.010	0.039	0.0077	0.003	0.0067	0.0545	0.28	0.183

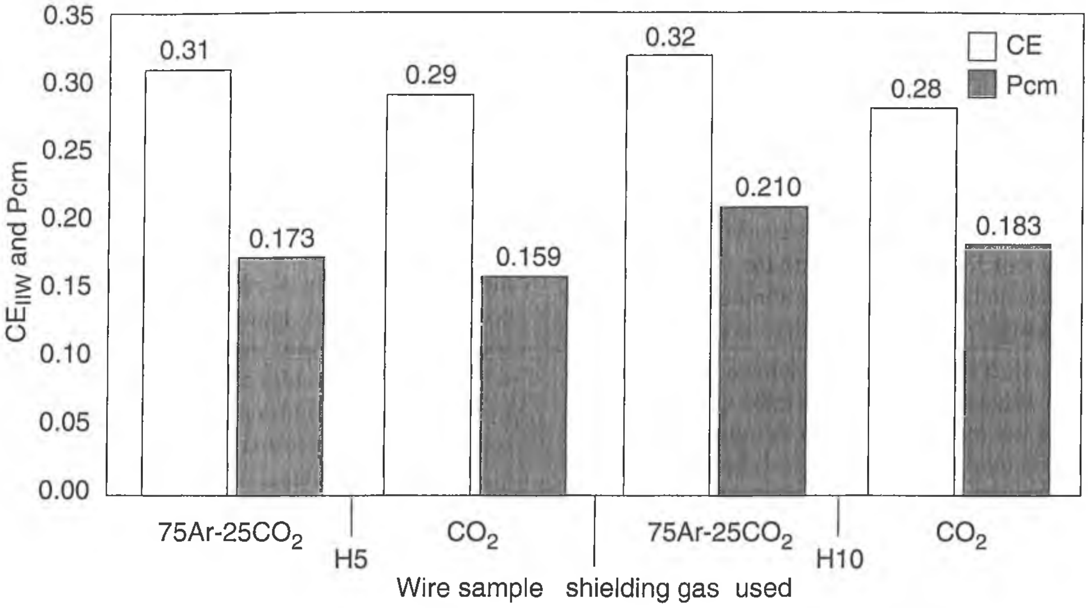

Figure 2. Graph showing the CE_{IIW} and Pcm values for H5 and H10 'all-weld metals' welded using 75Ar-25CO_2 and CO_2 shielding gas.

Welding consumables

In Australia there are two standards for carbon steel flux-cored wires that are commonly used by industry, ANSI/AWS A5.20-95 and AS 2203.1-1990. However, the wire classification system referred to in these standards and the specification of nominal hydrogen levels vary significantly.

The current work adopts the American ANSI/AWS A5.20-95 classification system for flux-cored consumables, rather than the more complex AS 2203.1-1990 Australian terminology. In this way the low strength rutile consumables used in the current work are referred to as E71T-1 rather than ETP-GMp-W503A as in AS 2203.1-1990. In addition, specification of hydrogen levels in the current work adopts the ISO 3690-2000 and AS 2203.1-1990 benchmarks, with the wires further designated as H5 and H10, with nominal levels of diffusible hydrogen in deposited weld metal of 5 and 10 mL/100 g, respectively. These designations are commonly used in Australia for hydrogen levels of FCAW consumables.

Two commercially available, seamless and seamed tubular gas shielded flux-cored wires of 1.6 mm diameter were used in this current work. These two wire types are considered to be the most widely used for FCAW of C and C-Mn steels for all positional applications in the Australian construction industry.

Both the seamless (H5) and seamed (H10) wires are micro-alloyed rutile types based on a titanium-boron flux composition. The wires not only differ significantly in the nominal hydrogen levels but also in their cross-section design, as shown in Figure 3 of reference 8. The butt seam of H10 wire was not fully closed leaving approximately 0.1 mm gap, thereby allowing the ingress of moisture or wire lubricant through the seam during the manufacturing process. Similar gaps have been also observed on equivalent wires supplied by a range of manufacturers.

The chemical compositions of 'all-weld metal' deposits, carried out in accordance to the Australian Standard AS 2203.1-1990, for the two wires used in the current work are presented in Table 2. The calculated carbon equivalent (CE_{IIW} and Pcm) values were determined after welding with both mixed 75Ar-25CO_2 and CO_2 shielding gases.

While the CE_{IIW} values are very similar for both the H5 and H10 wires, the Pcm values for the H10 wire samples were noticeably higher for both shielding gases, most likely due to the higher levels of boron in the H10 weld metal deposits, as B is an important factor in the Pcm carbon equivalent formula. Carbon levels are also higher in the H10 weld metal, and C is a dominating element in Pcm, more so than the IIW formula. The carbon equivalent values are shown diagrammatically in Figure 2.

It should be noted that the CE_{IIW} and Pcm values were calculated from multiple layer 'all-weld metal' deposits and different values would be obtained from a single weld bead due to the dilution effects. The use of CO_2 shielding gas reduced Mn, Si and B recovery, resulting in marginally lower CE_{IIW} and Pcm values for both consumables.

G-BOP test plates

The G-BOP test samples were prepared from a 50 mm thick rolled plate made from AS 3678-1999 Grade 250 steel with the chemical composition shown in Table 3.

In order to avoid a variation of results caused by a possible inconsistency between the batches of examined wires, the entire

Table 3. Chemical composition of parent material used for G-BOP test.

	Chemical analysis (weight %)															
	C	Mn	Si	S	P	Ni	Cr	Mo	Cu	V	Nb	Ti	N	Al	Fe	CE_{IIW}	Pcm
Check	0.165	1.23	0.34	0.008	0.015	0.024	0.021	0	0.01	0	0	0.018	0.0015	0.029	Rem	0.38	0.240
Ladle	0.15	1.25	0.32	0.009	0.014	0.024	0.023	0.003	0.007	0.004	0.001	0.019	0.0028	0.028	Rem	0.37	0.226

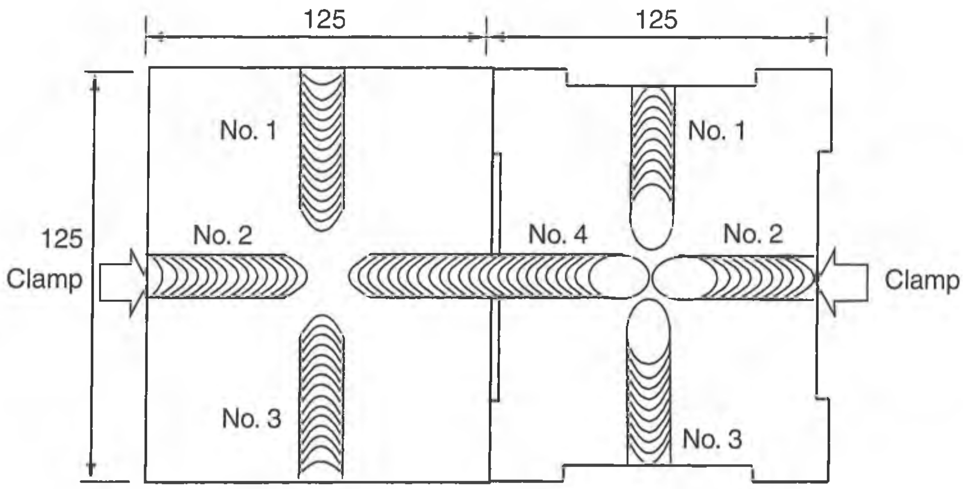

Figure 3. Schematic diagram of the G-BOP test block design that allows four weld beads to be deposited on each test block. Note: all dimensions in millimetres [not to scale].

experimental work was carried out by using only one spool from each wire supplied. After completion of each experiment, the wire was re-packed into its original packaging and stored in dry conditions at ambient temperature. In this way, the effect of long time exposure of wires to varying atmospheric conditions between the experiments was kept to a minimum.

Testing procedure and welding parameters

As proposed by Graville et al[15], four 0.75 mm deep recesses were machined in one of the mating blocks to introduce a notch so as to initiate transverse weld metal cracking. The multiple recesses enabled the use of the same pair of blocks for four weld passes, as shown in Figure 3.

The G-BOP tests were carried out on both H5 and H10 weld deposits using preheat temperatures of 20, 50, 80, 100 and 120°C. Samples were preheated at temperature for 60 minutes and the furnace temperature was set 10°C higher, thereby allowing for the time required to align and clamp the pair of blocks together prior to welding. Once the temperature of the blocks was stabilized at desired preheat temperature (+10°C), the mating blocks were quickly joined together by a large G-clamp, approximately in the middle of the block thickness. In order to allow a uniform loss of heat through radiation, the assembled blocks were rested on two supporting steel plates positioned across the block ends.

Weld beads of 100-120 mm length were deposited. Relative humidity and ambient temperature were recorded for each welded test block. Immediately after completion of welding, the blocks were allowed to cool down to ambient temperature in still air, while remaining restrained in the clamp for a minimum period of 72 hours. After this period of time, the restraining clamp was released and a small area in the vicinity of the weld bead (just over the gap of the two mating halves) was heated up to a dull, cherry red color using a gas flame and maintained for about 10 seconds. This procedure was designed to heat tint any pre-existing crack surfaces. The samples were then allowed to cool in still air to ambient temperature. Subsequently, the test weld was fractured open by simple bending, allowing visual examination of fractured surfaces. The fractured faces of all weld deposits were digitally recorded and visually examined at a magnification of 20x. The proportion of discolored transverse crack area, A_C , and total fused metal area, A_F, were precisely measured by using computer drawing software. From the measured areas, the percentage cracking was then calculated following equation:

Percentage cracking = $\frac{A_C}{A_F}$ x 100

Table 4. Matrix of welding parameters investigated in the current work to determine the effect of diffusible hydrogen content on cold cracking susceptibility.

Welding Parameter	Test values
Welding Current [A]	280 – 300 – 320
CTWD [mm]	15 – 20 – 25
Heat Input [kJ/mm]	1.26 – 1.35 – 1.44
Shielding Gas [18 l/min]	75Ar-25CO_2 and CO_2
Welding Voltage* [V]	29 – 30
Travel Speed* [mm/min]	400

*Note: * Not used as variable parameters*

In addition to the room temperature cracking (RTC), for each set of G-BOP samples the 10% crack preheat temperature (10%CPT) and the critical preheat temperature (CPT) were also determined.

Three welding variables: welding current, CTWD and shielding gas were selected to study their effects on susceptibility to cold cracking. Values used are given in Table 4. The chosen ranges of welding parameters were within the recommended ranges from both wire manufacturers and reflect the general industrial practice for welding in the downhand position.

Results

The results from testing of H10 and H5 weld metals using a range of preheat temperatures are summarised in Tables 5, 6, 7 and 8.

The G-BOP test results for H10 weld metal showed cold cracking for all combinations of welding parameters selected at the no-preheat condition of 20°C. In contrast, the H5 weld metals exhibited no cracking at room temperature. Therefore, the examination of H5 at higher preheat temperatures was not pursued. The results for the G-BOP tests are shown diagrammatically in Figure 4.

Table 5. Percentage of cold cracking for H10 welds deposited using 75Ar-25CO_2 shielding gas.

Welding parameters			Preheat temperature [°C]					
			20	**50**	**80**	**100**	**120**	
Welding current [A]	Shielding gas [18 L/min]	CTWD [mm]	Percentage cracking					H_D [mL/100 g]
280	75Ar-25CO_2	15	100	100	88	0	-	17.0
280		20	100	98	16	18	0	14.8
280		25	100	100	29	0	-	12.0
300		15	100	96	39	17	0	14.3
300		20	100	100	53	11	0	12.9
300		25	100	100	31	0	-	11.0
320		15	97	70	18	0	-	13.8
320		20	100	82	36	4	-	13.1
320		25	94	83	30	0	-	10.5
RH [%]-(Temperature [°C])			50-(17)	37-(18)	45-(25)	28-(26)	44-(22)	

Table 6. Percentage of cold cracking for H10 welds deposited using CO_2 shielding.

Welding parameters			Preheat temperature [°C]					
			20	**50**	**80**	**100**	**120**	
Welding current [A]	Shielding gas [18 L/min]	CTWD [mm]	Percentage cracking					H_D [mL/100 g]
280	CO_2	15	89	57	24	0	-	11.7
280		20	88	50	38	16	0	9.5
280		25	67	66	38	12	0	8.3
300		15	58	37	17	0	-	12.7
300		20	76	65	0	0	-	10.7
300		25	67	70	27	16	0	8.4
320		15	25	0	0	0	-	12.8
320		20	75	48	22	13	0	11.0
320		25	73	70	26	8	0	8.6
RH [%]-(Temperature [°C])			44-(22)	44-(22)	45-(25)	42-(26)	44-(22)	

Table 7. Percentage of cold cracking for H5 welds deposited using 75Ar-25CO_2 shielding gas.

Welding parameters			Preheat temperature [°C]					
			20	**50**	**80**	**100**	**120**	
Welding current [A]	Shielding gas [18 L/min]	CTWD [mm]	Percentage cracking					H_D [mL/100 g]
280	75Ar-25CO_2	15	0	-	-	-	-	3.5
280		20	0	-	-	-	-	2.2
280		25	0	-	-	-	-	1.5
300		15	0	-	-	-	-	3.1
300		20	0	-	-	-	-	2.1
300		25	0	-	-	-	-	1.7
320		15	0	-	-	-	-	2.6
320		20	0	-	-	-	-	1.6
320		25	0	-	-	-	-	1.6
RH [%]-(Temperature [°C])			42 -(17)	-	-	-	-	

Table 8. Percentage of cold cracking for H5 welds deposited using CO_2 shielding gas.

Welding parameters			Preheat temperature [°C]					
			20	50	80	100	120	
Welding current [A]	Shielding gas [18 L/min]	CTWD [mm]	Percentage cracking					H_D [mL/100 g]
280	CO_2	15	0	-	-	-	-	1.9
280		20	0	-	-	-	-	1.8
280		25	0	-	-	-	-	1.1
300		15	0	-	-	-	-	1.5
300		20	0	-	-	-	-	1.3
300		25	0	-	-	-	-	0.9
320		15	0	-	-	-	-	1.7
320		20	0	-	-	-	-	1.5
320		25	0	-	-	-	-	0.9
RH [%]-(Temperature [°C])			42 -(17)	-	-	-	-	

Discussion

Effect of welding current – (a) 75Ar-25CO_2

From the diagrams presented in Figure 4 for the H10 weld deposits, it is apparent that the welding current affects the weld metal cracking differently with varying combinations of the welding conditions. Perhaps the most noticeable difference is that an increase in current from 280 to 300 to 320 A, when using 75Ar-25CO_2 shielding gas, appears to have very little effect on percentage cracking at room temperature for all of the selected CTWD values.

Despite the significant differences in the weld metal hydrogen content range from 10.5 to 17.0 mL/100g, the percentage room temperature cracking (RTC) for eight out of ten G-BOP samples revealed 100% cracking. The samples welded using the highest welding current of 320 A exhibited only marginally less than 100%RTC.

As expected, by increasing the preheat temperatures from 50 to 120°C, the percentage of cold cracking was progressively reduced. At the shortest CTWD of 15 mm, for 75Ar-25CO_2 shielding gas (see graph (A15) in Figure 4), the plotted lines for each current level are further apart than those plotted in diagram (A25), for a CTWD of 25 mm. This effect is probably due to the generally higher and wider range of hydrogen levels for weld metal produced at 15 mm CTWD (13.8-17.0 mL/100g) than at 25 mm CTWD (10.5-12.0 mL/100g).

It should be noted that the weld metal deposited using the lowest welding current of 280 A at 15 mm CTWD contained the maximum amount of diffusible hydrogen (17.0 mL/100 g) and also exhibited a significantly higher percentage of cracking up to the preheat temperature of 80°C. This observation confirms an expectation from the earlier work that an increase in welding current reduces the weld metal diffusible hydrogen levels[8] and therefore a current increase would be expected to reduce susceptibility to cold cracking for 75Ar-25CO_2 CO_2 shielding gas and a CTWD of 15 mm.

Effect of welding current – (b) CO_2

The percentage cracking observed, when using CO_2 shielding gas, showed a more complex relationship with welding current, particularly at the shortest CTWD of 15 mm. The increase of welding current, which resulted in a slight increase of weld metal diffusible hydrogen content, produced a significant and unexpected reduction of susceptibility to cold cracking at room temperature, as shown in Figure 5 and diagram (C15) of Figure 4. This effect was also observed when preheating was employed. Possible explanations for this phenomenon are the geometry of the G-BOP welds cross sections, as shown in Figure 5, or differences in weld metal microstructure. The fused weld metal profiles varied significantly with increasing weld current for 20°C preheat. For example, the sample deposited using the lowest welding current of 280 A was characterised by a very flat and wide bead profile (89% RTC), whereas the sample welded using the highest welding current (320 A) was characterised by deeper penetration and a higher bead height and showed only 25% RTC. Increase in preheat temperature appears to suppress this bead shape effect and the weld deposit contours gradually become more uniform.

Although the diffusible hydrogen range for 15 mm CTWD is relatively narrow in the case of CO_2 shielding gas (11.7-12.8 mL/100g), an increase in welding current was found to be beneficial, significantly reducing the weld metal cold cracking susceptibility at all preheat temperatures examined in this work. For the lowest CTWD of 15 mm the welding current appears to be the governing variable in reduction of cracking percentage at 20, 50 and 80°C preheat temperature. However, for an increase of CTWD from 15 to 25 mm, the weld metal contains lower hydrogen levels and a narrower range of diffusible hydrogen contents, 8.3-8.6 mL/100 g, and a change in welding current has a less significant effect, see diagram C(25) in Figure 4.

Effect of CTWD

The effect of CTWD in the range 15 to 25 mm on cracking susceptibility is best illustrated in Figure 4, which shows that the relationship between the CTWD and percentage weld metal cracking is ambiguous. Although the CTWD appears to be a significant variable in terms of the hydrogen content, its effect on weld metal hydrogen cracking varied for the 75Ar-25CO_2 and CO_2 shielding gases.

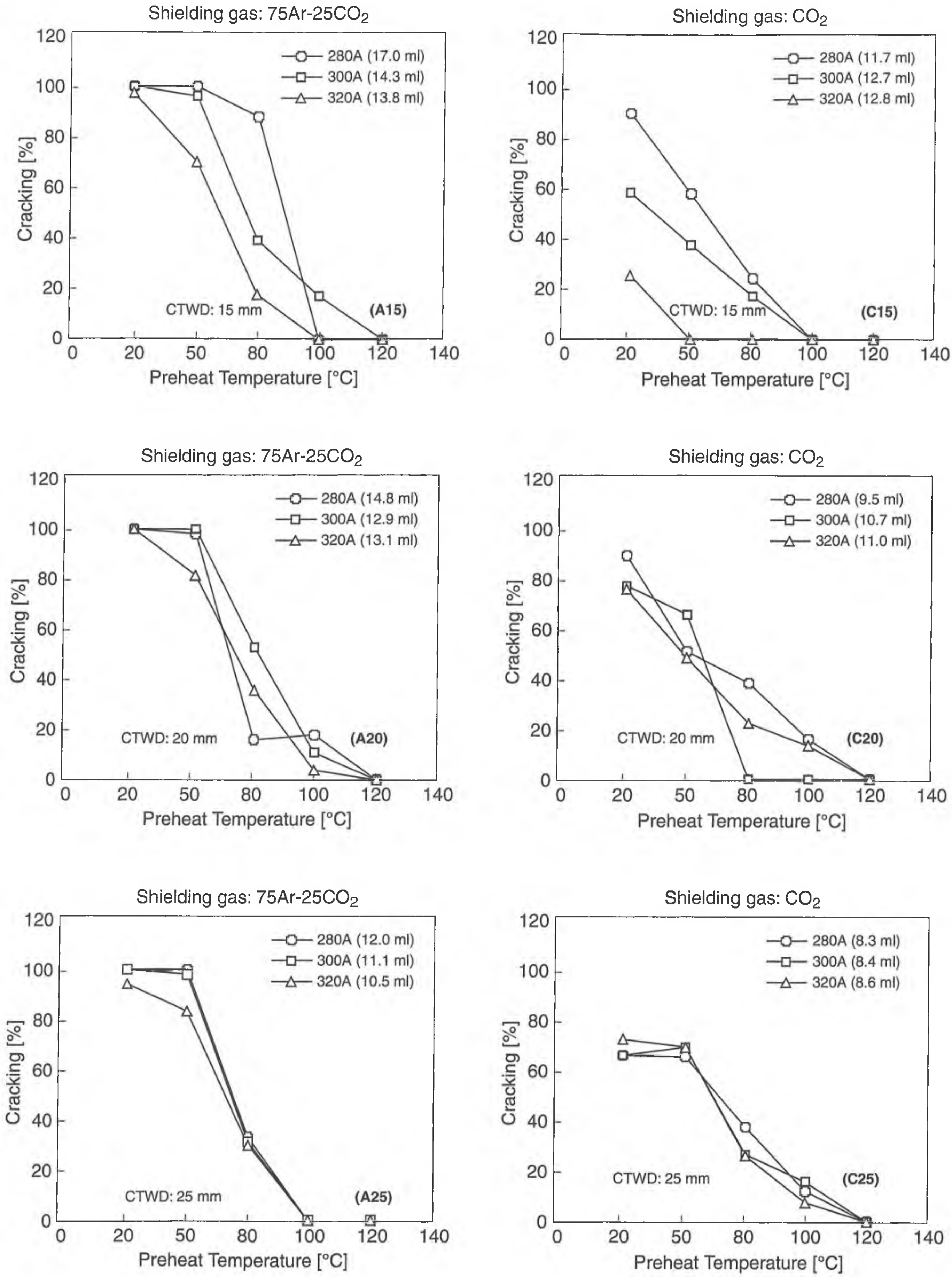

Figure 4. Graphs showing the percentage cracking for H10 weld metal in G-BOP tests, using 75Ar-25CO_2 and CO_2 shielding gases at welding currents 280, 300 and 320 A and CTWD of 15, 20 and 25 mm. Hydrogen contents are shown in parentheses for each welding current.

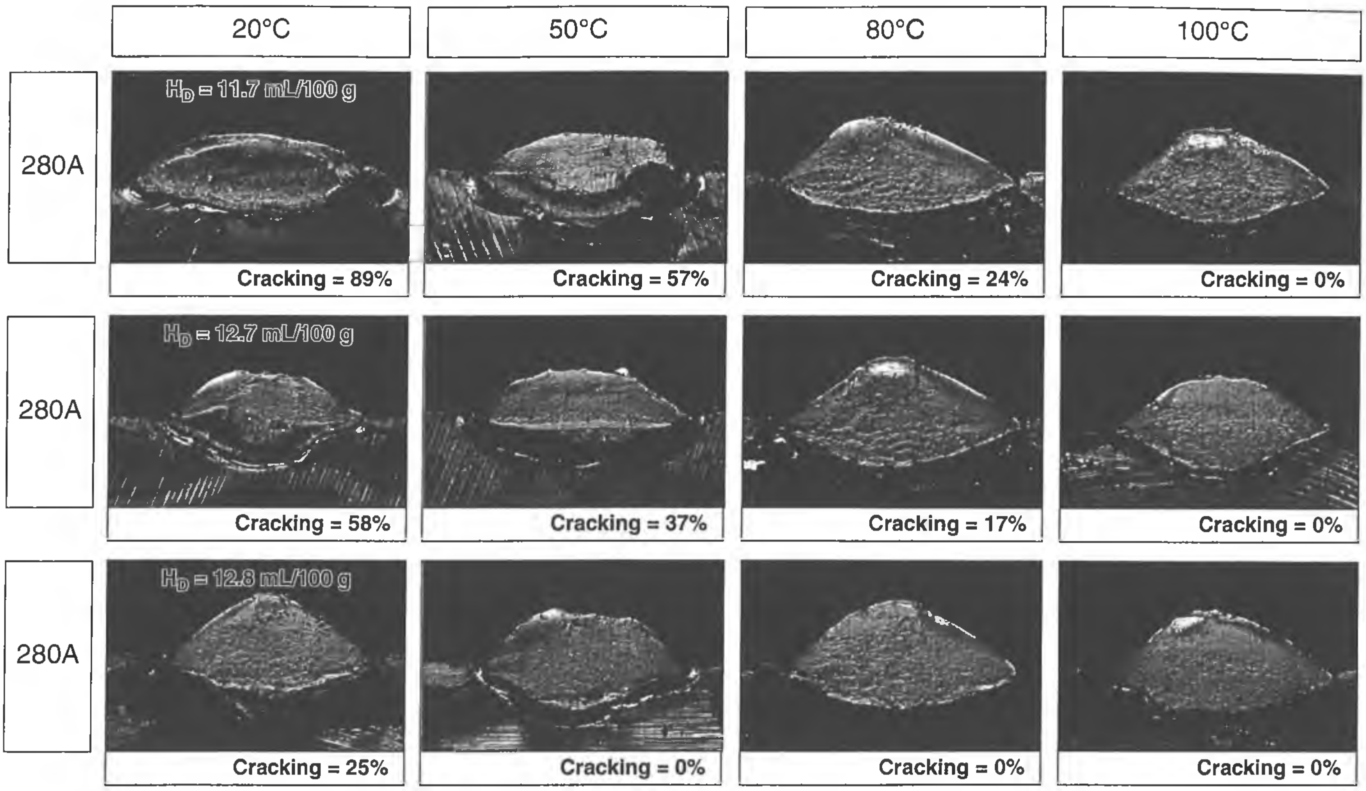

Figure 5. Macrographs showing the G-BOP fracture faces with cracking percentage of H10 weld metal deposited at preheat temperatures of 20, 50, 80 and 100°C using CO_2 shielding gas and a CTWD of 15 mm.

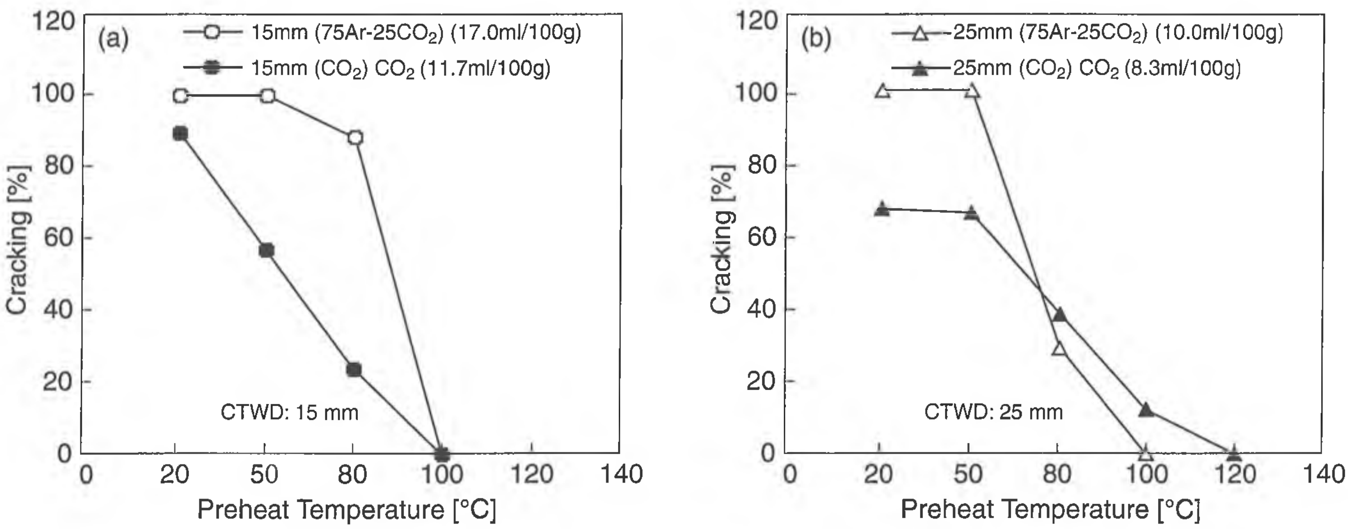

Figure 6. Graphs showing the percentage cracking for H10 weld metal in G-BOP tests, using 75Ar-25CO_2 and CO_2 shielding gases at constant welding current 280 A and CTWD of (a) 15 mm and (b) 25 mm.

Regardless of weld metal diffusible hydrogen levels at ambient temperature, the CTWD increase had no effect on percentage RTC at this temperature when welding involved 75Ar-25CO_2 shielding gas. The majority of G-BOP samples exhibited close to 100%RTC. However, when using CO_2 shielding gas the weld metal cracking was found to be more complex, as shown in diagrams (C15), (C20) and (C25) of Figure 4. Interestingly, at a CTWD of 15 mm using CO_2 shielding gas there appears to be a large scatter in RTC caused by the welding current. It should be noted that the weld metal hydrogen levels varied marginally (11.7-12.8 mL/100 g), but by increasing welding current the percentage of RTC decreased from 89 to 25%. By increasing the preheat temperature to 50, 80 and 100°C, the variation of the % cracking parameter was gradually narrowed (see (C15)). The bead deposited with no preheat at 15 mm CTWD and a welding current of 320 A contained the highest level of diffusible hydrogen (12.8 mL/100g), yet it exhibited the smallest RTC of 25%.

Although it is generally recognised that increasing CTWD significantly reduces weld metal diffusible hydrogen, the results indicate that the lowest value of RTC occurred in weld metal containing the highest diffusible hydrogen content in the range

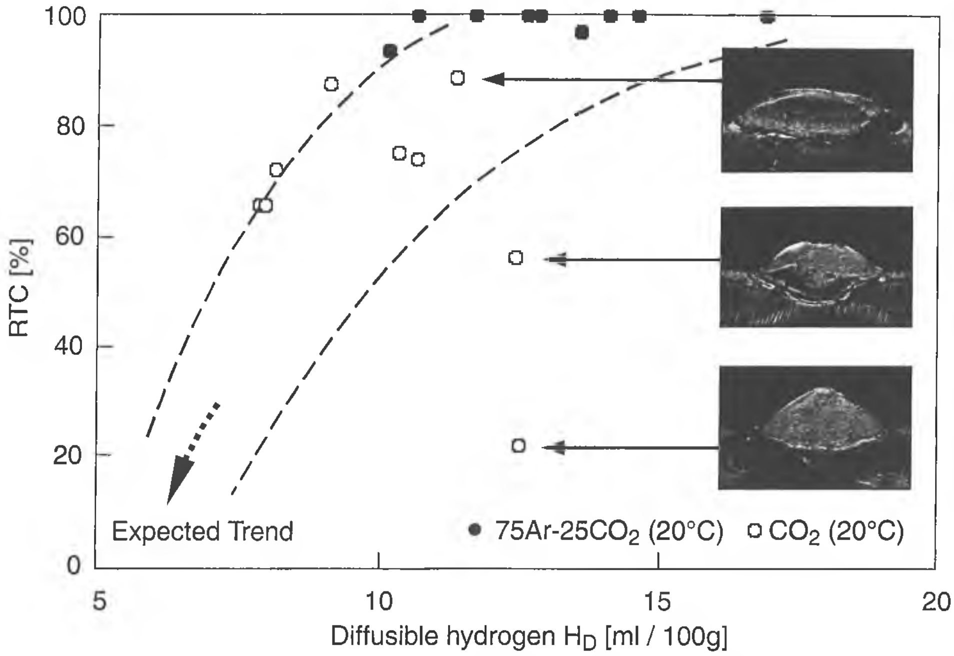

Figure 7. Graph showing the percentage cracking for room temperature welding using H10 weld metal, 75Ar-25CO_2 and CO_2 shielding gases, welding currents of 280, 300 and 320A and CTWD values of 15, 20 and 25 mm.

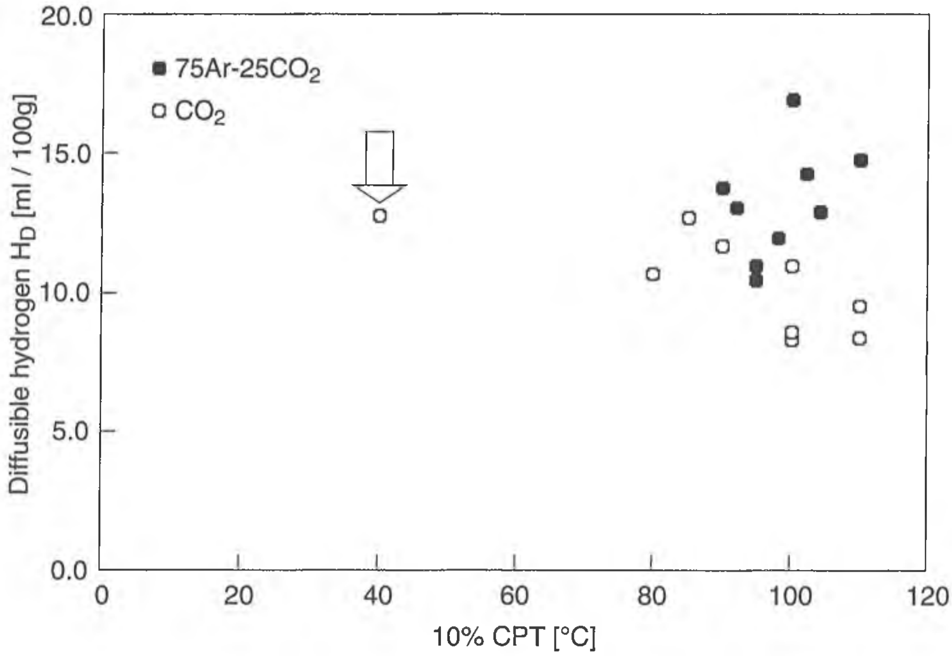

Figure 8. Graph showing the relationship between 10%CPT and diffusible hydrogen levels for G-BOP welds deposited using H10 wires, welded using 75Ar-25CO_2 and CO_2 shielding gases.

11.7 -12.8 mL/100 g for a CTWD of 15 mm and CO_2 shielding gas (see Figure 4, diagram C15). Despite the smallest diffusible hydrogen range (8.3-8.6 mL/100 g) for welds deposited using a CTWD of 25 mm and CO_2 shielding gas (see Figure 4, diagram C25), increasing the preheat temperature was not as effective in reducing % cracking as for welds deposited using the CTWD of 15 mm and the same shielding gas.

Effect of shielding gas

The investigation revealed that the shielding gas affects the susceptibility of the weld metal to transverse cold cracking, as shown by the results for the various preheat temperatures, Figure 6.

For constant CTWD and welding current, weld metal deposited using the mixed 75Ar-25CO_2 shielding gas generally exhibited a larger percentage of cracking than for the CO_2 shielding gas at room temperature. At preheat temperatures of 50 and 80°C, the decrease in percentage of cracking was more noticeable under CO_2 shielding gas at a CTWD of 15 mm, as shown in Figure 6(a). However, an increase in preheat temperature from 80 and 100°C resulted in a significant decrease in weld metal cracking for mixed shielding gas. Despite the fact that the % cracking values were significantly lower for CO_2 shielding gas for all G-BOP samples, the critical preheat temperatures for no cracking were found to be similar for both shielding gases.

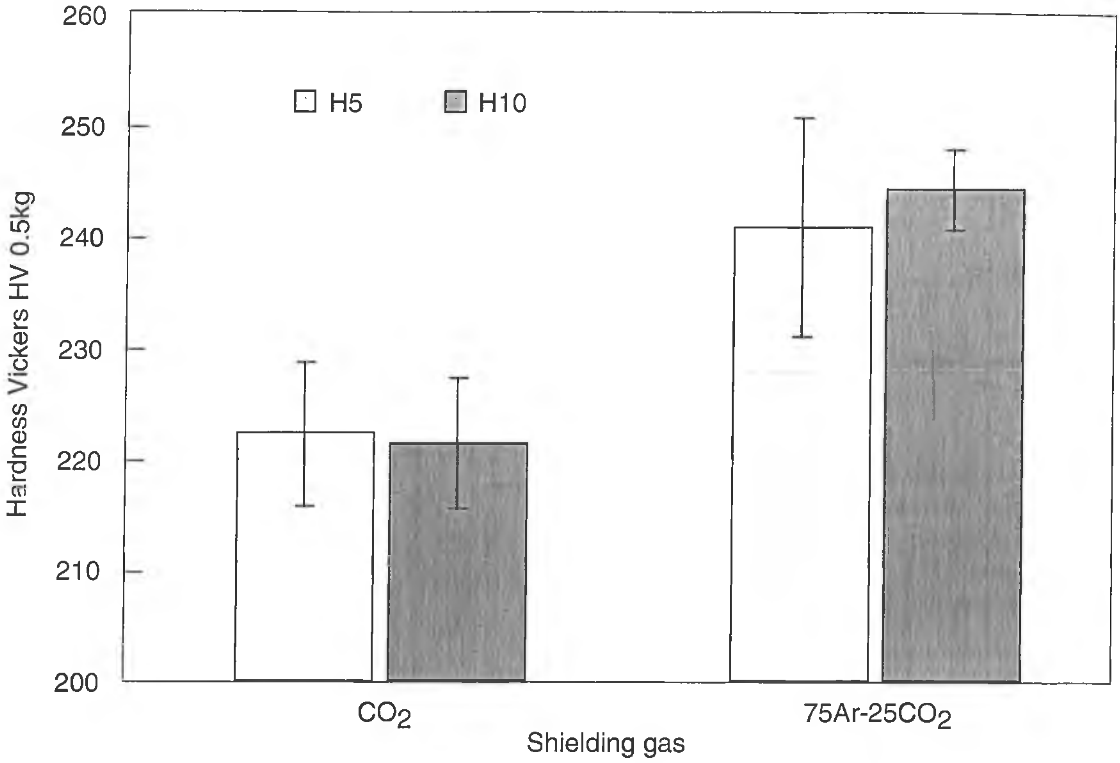

Figure 9. A plot of hardness values for G-BOP welds deposited using H5 and H10 wires for welding at ambient temperature with a welding current of 300 A, a CTWD of 25 mm and the two shielding gases.

A different relationship between percentage cracking and shielding gas was observed for a CTWD of 25 mm, as shown in Figure 6(b). At room temperature, the weld metal deposited using CO_2 shielding gas was characterised by significantly less cracking (67% RTC) compared with the 75Ar-25CO_2 deposit (100%RTC), However, a more rapid decrease in percentage of cracking was observed as preheat temperature was increased from 50 and 80°C on welds deposited using the 75Ar-25CO_2 shielding gas. This steeper reduction of cracking percentage in samples welded using mixed gas resulted in no cracking at 100°C, whereas in CO_2 shielding gas the cracking was present until the preheat temperature reached 120°C.

The effect of weld metal diffusible hydrogen content on cracking susceptibility at room temperature for 75Ar-25CO_2 and CO_2 shielding gas deposits is illustrated in Figure 7. It is apparent that CO_2 shielding gas deposits exhibited generally lower RTC values than 75Ar-25CO_2 weld deposits at similar levels of diffusible hydrogen. Although the ranges of hydrogen levels only partly overlap, the percentage cracking at room temperature was significantly lower in welds deposited using CO_2 shielding gas. This finding not only illustrates the differences in diffusible hydrogen generated in weld metal by the two shielding gases, but also demonstrates that the weld metals have different sensitivities to cold cracking. The two points residing outside of the expected band represent welds with bead contours that were noticeably different to the other weld beads. The diffusible hydrogen contents for the three beads identified in Figure 7 were similar (in the range 11.7 to 12.8 mL/100 g), but the amount of cracking at room temperature varied significantly. These differences may be due to the marked differences in the weld profiles, illustrated by the macrographs included in the figure.

It is important to note that by increasing preheat temperature, the samples welded using 75Ar-25CO_2 exhibited a steeper reduction of percentage of cracking than the CO_2 weld deposits, especially in samples welded with a CTWD of 25 mm (see Figures 4 (A25) and (C25)).

RTC versus 10%CPT

The effect of increase in preheat temperature on the reduction of cracking can be expressed by the 10%CPT value. This parameter is particularly useful when the weld metal containing higher hydrogen levels gives 100%RTC.

The H10 welds deposited using the 75Ar-25CO_2 shielding gas, characterised by higher diffusible hydrogen levels, display a higher cracking susceptibility at room temperature compared to those deposited using CO_2 shielding gas. However, as illustrated in Figure 8, the welds deposited using 75Ar-25CO_2 and CO_2 shielding gas revealed similar values of 10%CPT, in the range 95 - 110°C. This finding demonstrates that although the welds deposited using CO_2 shielding gas exhibited a higher resistance to cold cracking to those deposited using 75Ar-25CO_2 shielding gas at room temperature, both types of welds showed a similar response to preheat. The generally higher initial hydrogen levels in the 75Ar-25CO_2 welds did not appear to affect the 10%CPT value. Note that the outlying point in Figure 8 (arrowed), represents a weld sample with bead contour different to the other weld beads, as discussed earlier. This sample not only exhibited the smallest 10%CPT value of 40°C, but also the lowest amount of cold cracking at room temperature (25%RTC). It is concluded that this result is an anomaly resulting from an unusual bead geometry.

So although welds deposited using 75Ar-25CO_2 may exhibit a higher degree of cracking at room temperature this does not necessarily mean that the weld will require significantly higher preheat temperature to eradicate cracking.

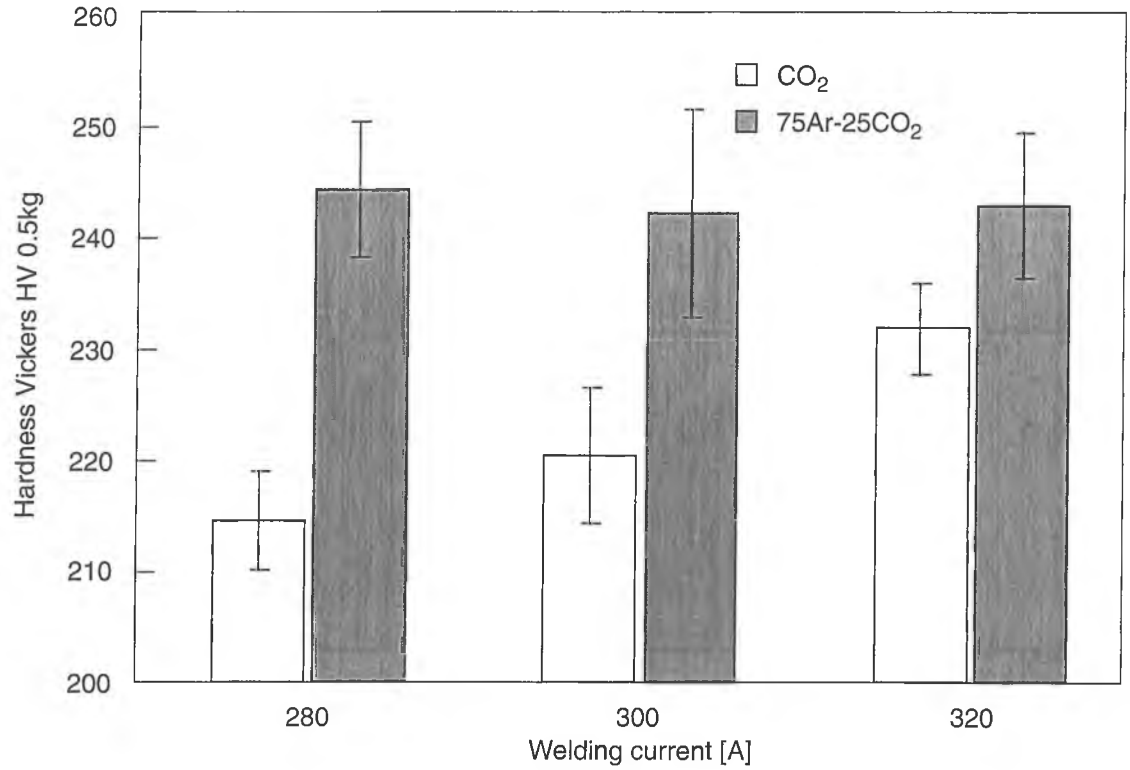

Figure 10. A plot of hardness values for G-BOP welds deposited using the H10 wire at various welding currents (280, 300 and 320 A) using CO_2 and 75Ar-25CO_2 shielding gases, CTWD of 25 mm and preheat temperature of 20°C.

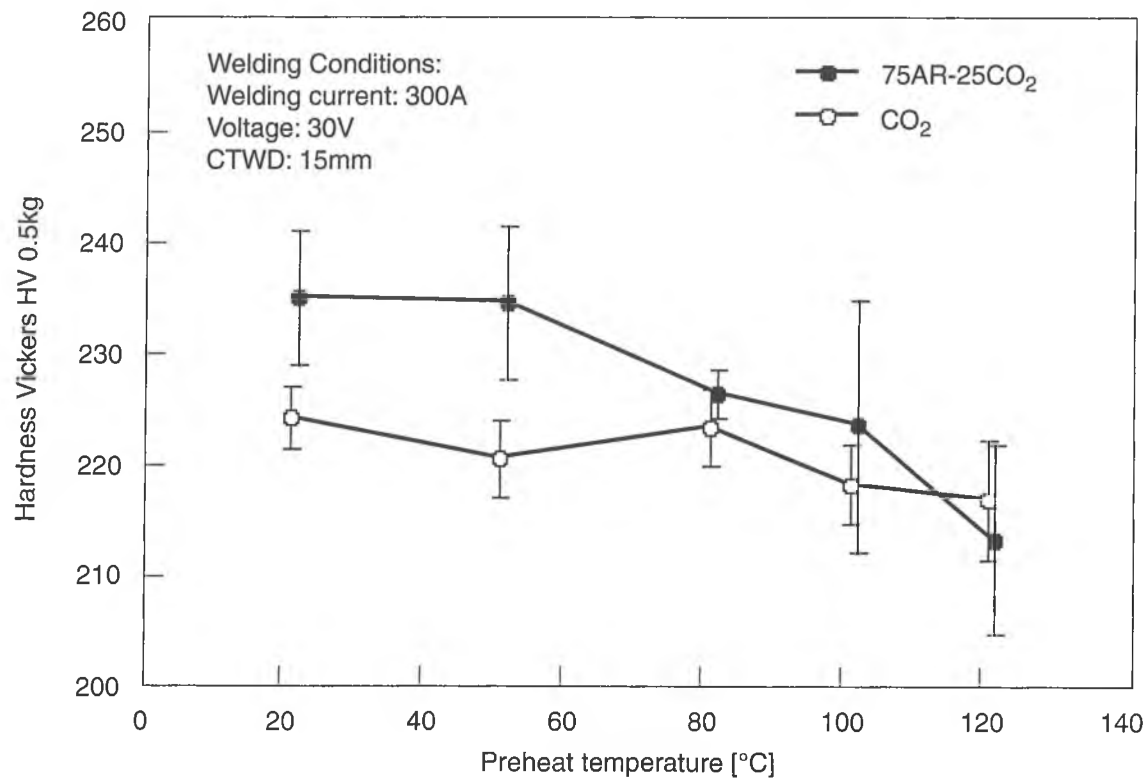

Figure 11. A plot of hardness values for G-BOP weld deposited using H10 wire at various preheat temperatures and using 75Ar-25CO_2 and CO_2 shielding gas.

Weld metal hardness

Samples of welds from the G-BOP tests were extracted for Vickers micro-hardness measurements using 0.5 kg load from both H10 and H5 weld deposits. The hardness values reported are averages determined from a minimum of five measurements (Figure 9).

In general, the H10 welds deposited using 75Ar-25CO_2 revealed higher hardness values than those using CO_2 shielding gas. This effect was more apparent at lower preheat temperatures of 20 and 50°C, but at higher preheat temperatures the difference was found to diminish. Since the G-BOP samples of H5 weld metal exhibited no cracking at room temperature and no welding was carried out at higher preheat temperatures and the effects of increasing preheat temperatures on H5 weld deposit hardness were not investigated.

The hardness measurements of weld metals deposited using H5 and H10 wires reacted similarly to a change in shielding gas for a given welding condition at 20°C. This is shown in Figure 9 for welds deposited at a current of 300 A and a CTWD of 25 mm for both the H5 and H10 wires. Both welding consumables exhibited a similar hardness increase (20 HV0.5) due to a change from CO_2 to 75Ar-25CO_2.

The relationships between the welding current and hardness values for H10 welds deposited at the preheat temperature of 20°C, and a CTWD of 25 mm are shown in Figure 10 for both shielding gases. The weld metal hardness was found to increase with increasing welding current using CO_2 shielding gas: increase in welding current from 280 to 320 A resulted in an increase in weld metal hardness from 215 to 233 HV0.5. However, the weld metal hardness remained unchanged in welds deposited using 75Ar-25CO_2 shielding gas. It is inferred that increasing current for CO_2 welding changes the microstructure, whereas mixed gas welding does not. Further investigation is required to clarify this difference.

A gradual decrease of weld metal hardness with increasing preheat temperature was observed, as shown in Figure 11. The samples welded using 75Ar-25CO_2 exhibited a greater reduction of hardness with increasing preheat temperature than those welded using CO_2 shielding gas. For example, the hardness decreases for an increase in preheat temperature from 20 to 120°C were 22 and 7 hardness points for 75Ar-25CO_2 and CO_2 shielding gas, respectively. The lower decrease in hardness for CO_2 shielding gas is probably related to the leaner chemistry of the welds established by the higher oxidizing potential of the gas. Table 2 indicates that the Mn and Si contents of the CO_2 welds are significantly lower than for 75Ar-25CO_2 welds. The higher as welded hardness of the more highly alloyed 75Ar-25CO_2 welds is more markedly affected by increasing preheat because of structural coarsening due to the slower postweld cooling rate.

In summary, weld metal hardness was found to be reduced by an increase in preheat temperature for both shielding gases, although the effect was more pronounced for the mixed gas. The results of weld metal hardness measurements confirmed a consistent difference in weld metal hardness for welds deposited using the different shielding gases. This effect was observed for both H5 and H10 weld deposits.

Conclusions

This paper reports the findings of a G-BOP test program to assess the effects of welding parameters and shielding gases on the HACC-susceptibility of weld metal deposited by seamless (H5) and seamed (H10) rutile wires. The major conclusions drawn from this investigation are as follows.

1. E71-T1 (rutile) weld metal containing diffusible hydrogen of ≥10 mL/100 g is highly susceptible to cold cracking at room temperature.
2. In contrast the H5 weld deposits exhibited no cracking at room temperature under any of the welding conditions investigated.
3. The G-BOP test results indicated that although the same H10 welding consumable deposited using the different shielding gases can show different responses to preheat temperature, a preheat temperature of 120°C decreases cracking to ≤10% for welds deposited using both shielding gases.
4. The results of room temperature G-BOP tests show that the susceptibility of weld metal to HACC is reduced in welds deposited using CO_2 shielding gas for all of the combinations of CTWD and welding current investigated. This effect can be explained by the lower levels of diffusible hydrogen in welds deposited using CO_2.
5. Increasing preheat was found to decrease the percentage of cracking in the H10 weld deposits in all cases. A major effect of increasing preheat temperature is to decrease the diffusible hydrogen concentration.
6. Preheat temperature increase from 20 to 120°C reduced the weld metal hardness by 22 and 7 points HV 0.5 for welds deposited using 75Ar-25CO_2 and CO_2 shielding gas, respectively. This reduction in hardness also contributes to the reduced HACC- susceptibility.
7. For tests welds deposited at 20°C, using 75Ar-25CO_2 shielding gas and a CTWD of 15 mm, an increase in the welding current was found to reduce the weld metal diffusible hydrogen levels, but not the susceptibility to cold cracking. In the case of the G-BOP tests welds deposited under CO_2 shielding gas, an increasing welding current resulted in a significant reduction of cold cracking at room temperature at a CTWD of 15 mm, despite a slight increase in H_D. This difference was less evident at higher CTWDs.
8. Shielding gas composition influenced the chemical composition of the weld deposits. For both welding consumables (H10 and H5), welds deposited using 75Ar-25CO_2 shielding gas exhibited higher CE_{IIW} and Pcm values than the weld metal deposited using CO_2 shielding gas. This compositional difference is consistent with the observed hardness trends of the welds. The measured hardness results for welds produced with the same weld metals parameters were about 20 HV0.5 points higher for 75Ar-25CO_2 compared to those deposited using CO_2.

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