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The Hydrogen-Induced Cracking
Resistance of Consumables
for Use in the Fabrication of the
COLLINS Class Submarines

James L. Davidson

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The Hydrogen-Induced Cracking Resistance of Consumables for Use in the Fabrication of the COLLINS Class Submarines.

James L. Davidson

**Ship Structures and Materials Division
Aeronautical and Maritime Research Laboratory**

DSTO-TR-0354

ABSTRACT

A MIL-121TM flux cored arc welding consumable, Alloy Rods Dualshield® II 120-M2, has recently been qualified for use in the fabrication of the pressure hull of the Royal Australian Navy COLLINS class submarines. Gapped bead on plate testing has been carried out to compare the hydrogen-induced cracking resistance of welds produced using the flux cored arc consumable with welds produced using the manual metal arc and submerged arc consumables currently used on the submarines. These tests will ensure that the introduction of the flux cored arc welding consumable does not increase the risk of hydrogen-induced cracking in the submarine pressure hull. The resistance of the flux cored arc welding consumable was found to be better than either the manual metal arc or the submerged arc consumables and the introduction of the flux cored arc welding consumable for use in the welding of the COLLINS pressure hull does not increase the risk of hydrogen induced cracking. The lower strength level of the deposited flux cored arc weld metal, contributed to its greater hydrogen-induced cracking resistance by reducing the residual stress state in the weld metal and by improving its inherent resistance to hydrogen "embrittlement". It is shown that there is no fundamental reason for a "one to one" relationship between hydrogen-induced cracking resistance and toughness in nominally 690 MPa yield stress weld metal. The hardness of the heat affected zones produced by each of the processes were not significantly different and gave no information regarding the likelihood of weld metal hydrogen induced cracking.

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The hydrogen-induced cracking resistance of consumables for use in the fabrication of the COLLINS class submarines.

Executive Summary

The construction of the COLLINS class submarines has been relatively free from hydrogen-induced cracking during welding. This has been achieved by great care and attention to keeping levels of hydrogen low in welding, particularly through the diligent use of a high degree of preheat (120°C) in all pressure hull welding. The latter involves a large degree of conservatism and expense but is necessary because the underlying process of weld metal hydrogen-induced cracking is not well understood.

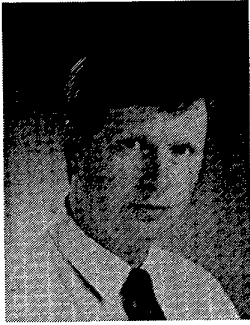
This good welding record has been achieved using the manual metal arc and submerged arc processes, although the flux cored arc welding process has recently been qualified for use in the fabrication of the COLLINS pressure hull. A crucial part of the backup to this qualification process is to establish, by controlled laboratory testing, the relative resistance of the flux cored arc welding consumables to hydrogen induced cracking. These tests will ensure that the introduction of the flux cored arc welding consumable does not increase the risk of hydrogen-induced cracking in the pressure hull of the COLLINS class submarine.

The gapped bead on plate test was used to compare the hydrogen-induced cracking resistance of welds produced using the flux cored arc consumable with welds produced using the manual metal arc and submerged arc consumables currently used on the submarines. The hydrogen-induced cracking resistance of the flux cored arc welding consumable was found to be better than either the manual metal arc or the submerged arc consumables and so the introduction of the flux cored arc welding consumable for use in the welding of the COLLINS pressure hull will not increase the risk of hydrogen induced cracking. The hydrogen induced cracking resistance of the manual metal arc consumable was found to be better than the submerged arc consumable, in agreement with shop floor experience.

The present investigation has produced two findings of general importance. Firstly, the notion that good weld metal toughness will confer good resistance to hydrogen-induced cracking is examined and it is shown that there is no scientific basis for a "one to one" relationship between toughness and hydrogen-induced cracking resistance. Secondly, the hardness of the heat-affected parent metal adjacent to the weld is often used to flag the possible propensity to hydrogen-induced cracking. However, the present results show that the hardnesses of the heat-affected zones produced by each process were similar to each other and give no information regarding the likelihood of weld metal hydrogen-induced cracking.

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1. Introduction

The push for increased operability of naval platforms continues to drive the requirement for stronger, lighter structures, necessitating the welding of steels of increasing strength and propensity for hydrogen-induced cracking. The COLLINS class submarines are fabricated from a 690 MPa yield strength steel. Hydrogen-induced cracking (HIC) occurs in such weldments due to the combined presence of tensile residual stresses and hydrogen absorbed during welding. It is the most common cracking problem encountered when welding ferritic steel structures [1]. HIC which occurred during the fabrication of the SSN21 SEAWOLF submarine necessitated the reworking of 15% of welds at a cost of tens of millions of dollars (US) [2].

HIC usually occurs a few hours to a few days after welding and for this reason is often called cold cracking or delayed cracking in the welding industry. There is an incubation period prior to cracking because hydrogen diffusion to regions of high triaxial stress around inclusions and other crack like defects is required. HIC can occur at various positions in the weldment depending on the degree of restraint and the composition of the plate and weld metal (see Figure 1). With the improvements in steel compositions and processing, such as those found in BIS 812 EMA, the hull steel used in the COLLINS class submarines, the risk of heat affected zone (HAZ) HIC has been greatly reduced and the avoidance of weld metal HIC is becoming increasingly more important [3,4].

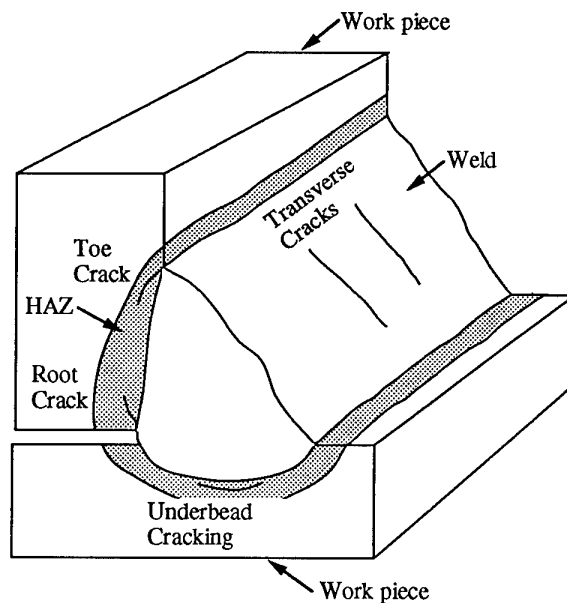


Figure 1. Location and nomenclature of hydrogen-induced cracks. (after ref. [1])

Sources of hydrogen for HIC include (i) rust, primer, paint and hydrocarbons which may be present on the weld preparation, and (ii) moisture on the weld preparation and from the atmosphere entrained in the arc (iii) moisture or hydrocarbons in or on the welding consumables (Figure 2) [5].

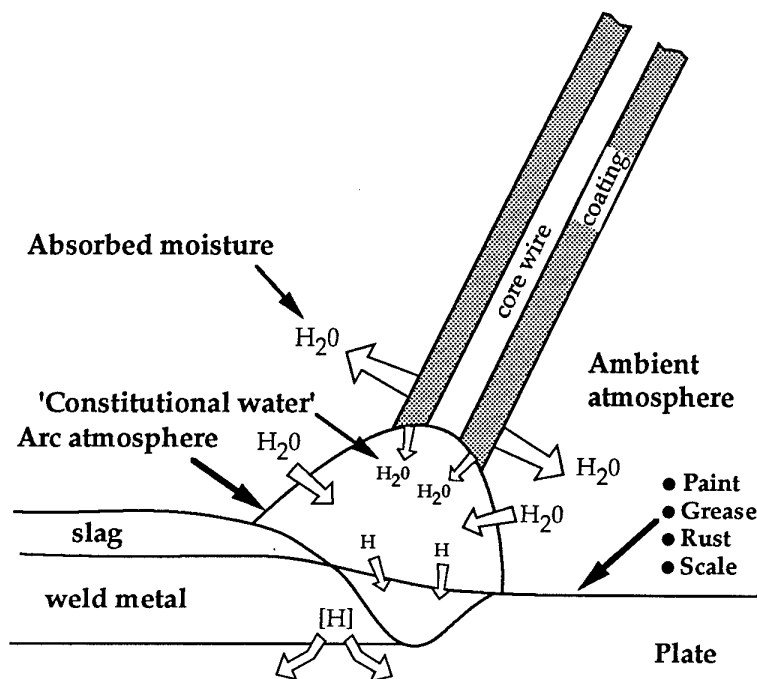


Figure 2. Schematic illustration of hydrogen entering the weld pool during manual metal arc welding (after Ref. [5]).

The risk of HIC can be reduced by preheating the weld preparation and by maintaining the welding heat input above some critical level. These measures accelerate the escape of hydrogen from the weldment and limit the weld metal and HAZ hardnesses. Hardness is an important parameter since it is an indicator of strength and it may flag the presence of martensitic microstructures which are deleterious to toughness [6]. The technical specification for the welding of COLLINS sets an upper limit on the allowable weld metal and HAZ hardnesses. While the use of higher heat input, to avoid HIC, is consistent with higher productivity welding, the necessity of preheat adds up to \$10,000/tonne of steel welded to the cost of fabrication.

The need for preheat during welding of ferritic steels has traditionally been dictated by the susceptibility of the HAZ to HIC. However, major improvements in steel compositions and processing, such as those found in BIS 812 EMA steel, have reduced the preheat required to safely avoid HIC in the HAZ. Whilst technological advances in

the control of HIC, such as the use of lower carbon and hydrogen levels, have been made in the development of welding consumables they have not kept pace with developments in the production of the so-called 'preheat free' steels [7,3,8]. Consequently in modern low carbon microalloyed steels, such as BIS 812 EMA, the focus of attention in the control of HIC has shifted from the HAZ to the weld metal.

Weld metal hydrogen cracks are often tightly closed and are convoluted, making their detection by radiographic techniques difficult [9]. Furthermore, hydrogen cracks in multipass weldments in steels of 690 MPa yield strength usually occur within the weld metal, transverse to the weld length. These cracks can and have been missed by routine ultrasonic inspection, which is conducted transverse to the weld.

Every precaution has been taken to avoid HIC during the fabrication of the COLLINS class submarines. Previously only MMAW and SAW were qualified for use in fabrication of the COLLINS pressure hull. However a flux cored arc welding (FCAW) consumable has recently undergone qualification trials as a potential replacement for MMAW. The FCAW process has the advantages of enhanced productivity, user appeal and can be used in all positions. The present work focuses on the hydrogen-induced cracking resistance of the FCA welds compared with the MMA and SA processes and consumables currently used in the fabrication of the COLLINS hull.

2. Experimental Method

The processes and consumables investigated are listed in Table 1 together with the treatment of the flux and electrodes, the heat input and the weld metal diffusible hydrogen content. Diffusible hydrogen contents were measured according to Australian Standard 3752 using an Oerlikon-Yanaco G-1006H gas chromatograph hydrogen analyser. The SAW and MMAW consumables employ basic flux formulations while the FCAW wire contains a rutile based flux. The chemical compositions and mechanical properties of welds deposited using the three consumables are listed in Tables 2 and 3 respectively. The arc energy produced by the power source is given by the welding parameters,

$$\text{Arc energy} = \text{Current} \times \text{Voltage} / \text{Travel Speed}$$

whereas the heat energy transferred from the arc to the workpiece, heat input, is dependent on the arc efficiency factor, η , and is given by

$$\text{Heat input} = \eta \times \text{Arc Energy}.$$

Heat input is the more relevant parameter in HIC studies since the energy transferred to the workpiece affects both the microstructure and the level of hydrogen in the weldment.

Table 1. Details of the three processes/consumables under investigation.

Process	Consumable	Flux/Electrode Treatment	Storage Temp. (°C)	Arc Energy (kJ/mm)	Heat Input* (kJ/mm)	H _D ** (ml/100g)
SAW	Ø 4mm LTEC® 120S Oerlikon OP121TT flux	3hrs @ 350°C	100	2.0	2.0	6.5
FCAW	Ø1.2mm Alloy Rods Dualshield II 120-M2, Argoshield™ 52 gas	Nil	50	1.7	1.4	6.0
MMAW	Ø 4mm Alloy Rods Atom Arc® 12018-M2	1 hr @ 300°C	70	1.9	1.4	3.5

* calculated using η from [10]** H_D= diffusible hydrogen content per 100 g of deposited weld metal

Table 2. Chemical composition and carbon equivalent of undiluted weld deposit.

	C	Mn	Si	S	P	Ni	Cr	Mo	Cu	V	Nb	B	Ti	P _{cm} **
FCAW	0.02	1.71	0.2	0.007	0.008	2.41	0.04	0.01	0.03	0.022	0.010	0.0018	0.022	0.17
SAW	0.06	1.47	0.40	0.006	0.014	2.35	0.31	0.48	0.03	0.008	0.003	0.0008	0.012	0.24
MMA *	0.04	1.55	0.20	0.008	0.01	3.67	0.03	0.26	Nil	0.01	-	-	-	0.21

** $P_{cm} = C + (Mn+Cr+Cu)/20 + Si/30 + Ni/60 + Mo/15 + V/10 + 5B$ [11]

* from [12]

Table 3. Typical weld metal mechanical properties [13].

	Yield R _{el} (MPa)	Tensile R _m (MPa)	Elongation A5	Charpy Impact Energy Joules(% fibrosity)	
				-51°C	-18°C
FCAW	735	776	19	58 (55)	86 (62)
SAW	763	853	21	88 (54)	126 (66)
MMAW	794	844	19	75 (55)	99 (62)

2.1 HIC susceptibility of MMAW, SAW and FCAW consumables

A range of mechanical tests is available to measure the HIC susceptibility of weldments produced by a given process or consumable. Each test has a bias toward producing either HAZ or weld metal cracking and will affect the orientation of cracking. This is due to the geometry of the test and the resulting stresses imposed on the weldment. The Gapped Bead on Plate (GBOP) test [14] was chosen for this investigation since it provides a good measure of the resistance to transverse weld metal cracking. The GBOP test piece is made up of two blocks, one of which has a machined recess on one face (Figure 3). The blocks are clamped together so that the machined recess forms a slot between the two blocks.

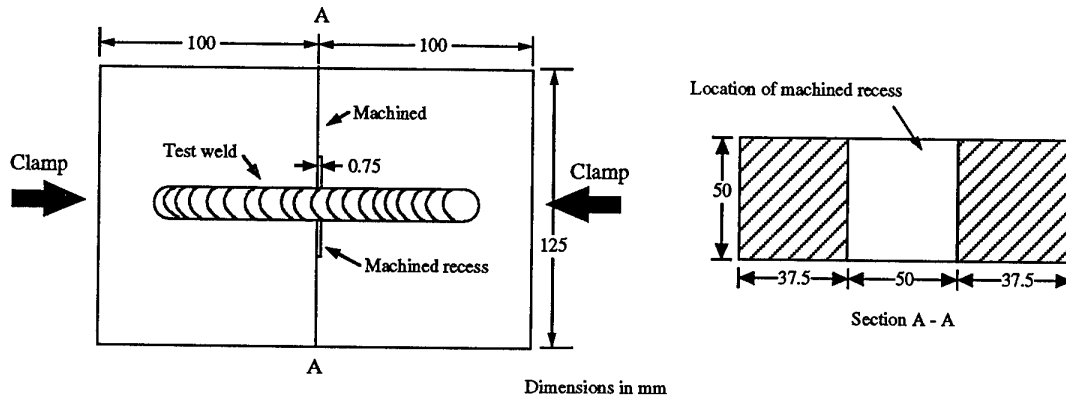


Figure 3. The GBOP test specimen.

In the present investigation, the test pieces of BIS 812 EMA steel were heated to the preheat temperature using resistive heating pads and a 150 mm long weld bead was deposited over the slot in the flat position. The test piece was then removed from the heating pad and placed on a wooden benchtop to cool. After 48 h the clamp was removed and the weld bead was heat tinted before being broken apart, by simple bending at ambient temperature, for inspection. The crack face was then photographed and the image digitised so that the percentage area of the hydrogen-induced crack could be measured using image analysis software (Figure 4). The percentage area of the hydrogen-induced crack was then used as an indicator of the welds resistance to HIC. To compare the FCAW, MMAW and SAW processes, welding was carried out with each consumable over a range of preheats from 20 to 120°C and at the heat inputs indicated in Table 1.

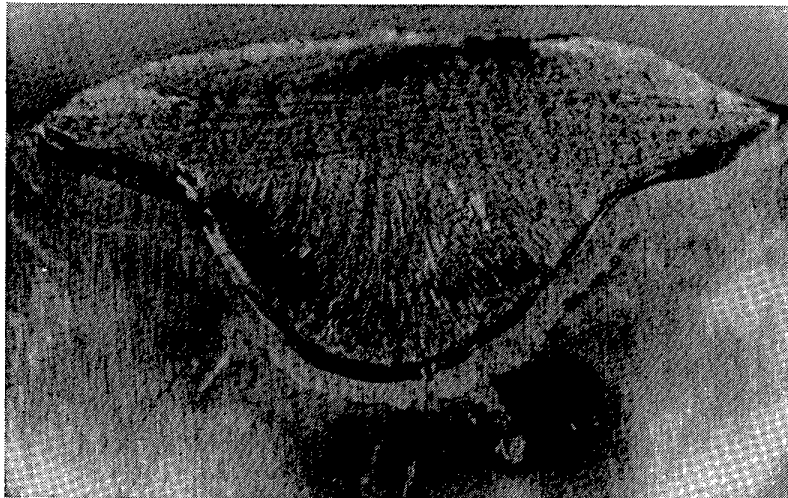


Figure 4. The crack face of a GBOP specimen which has been broken open for inspection. The HIC portion of the crack is delineated by the dark heat tinting.

Hardness was measured according to ASTM standard E92 using a pyramidal diamond indenter and a 10 kg load. Reported weld metal hardnesses are an average of 6 measurements and HAZ hardnesses are an average of the four highest measurements taken from the coarse grained structure in the re-entrant portion of the HAZ.

3. Results

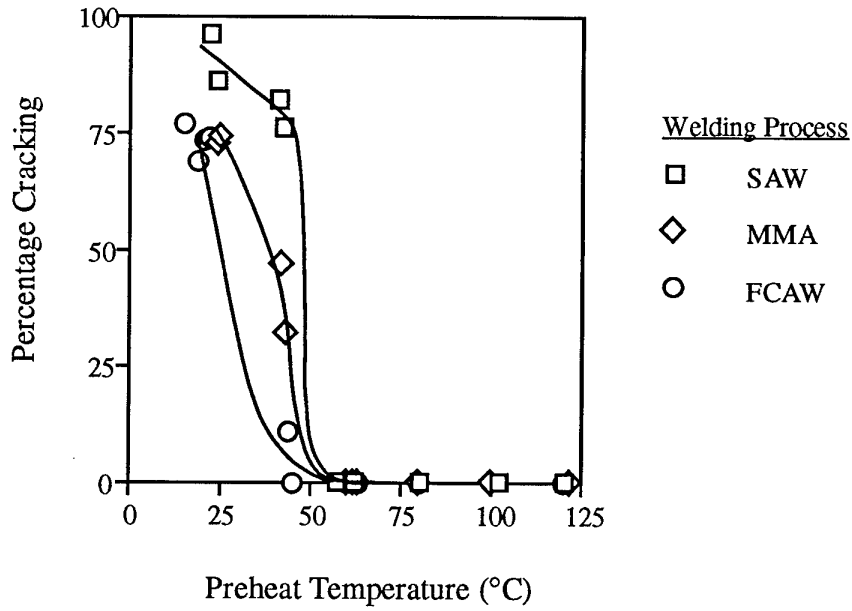
3.1 Comparison of the HIC resistance of FCAW, MMAW and SAW consumable

The results of the GBOP testing of the three consumables are illustrated in Figure 5(a). None of the processes produced cracks when a preheat of 60 °C or higher was applied. However, all exhibited some degree of cracking at 40 °C preheat and the percentage of cracking generally increased with decreasing temperature. The largest percentage of cracking at 40°C occurred in the specimens welded using the SAW process (76 and 80%) and the least cracking was observed with the FCAW process (0 and 11%). Corresponding weld metal and HAZ hardness measurements are presented as a function of preheat in Figure 5(b). Both weld metal and HAZ hardnesses increase with decreasing preheat. The hardness of the FCA weld metal is considerably less than either the MMA or SA weld metal hardnesses. The HAZ hardnesses resulting from the three processes are similar to each other.

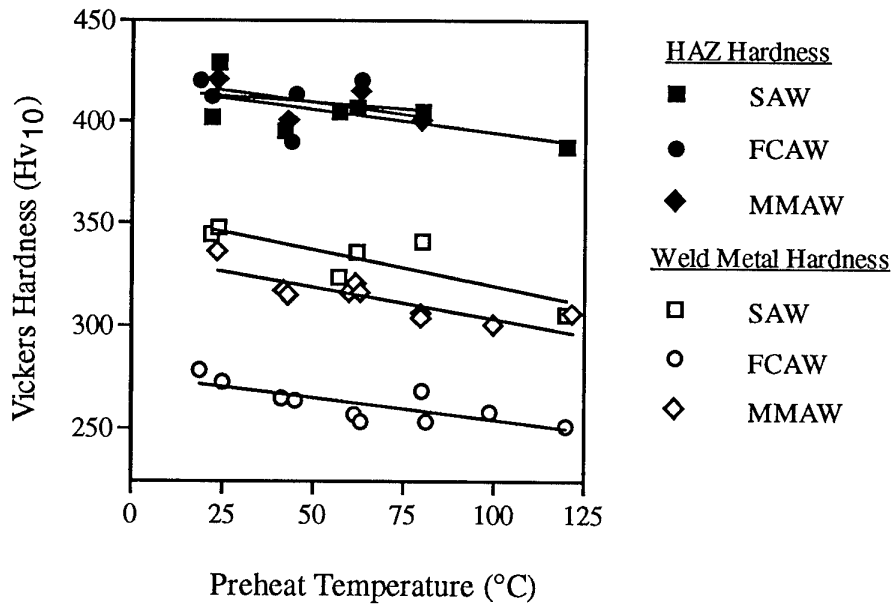
4. Discussion

4.1 HIC resistance of the FCA consumable

The present results (Figure 5) demonstrate that the FCA weld metal has the lowest percentage of cracking in the GBOP test of the processes investigated, despite the fact that its diffusible hydrogen content is 2.5 ml/100g (ie. 70%) higher than that of the MMA weld metal. This observation may be explained in terms of the relative yield stress of the weld metals as follows. The local stress intensity factor must reach a critical level for a given hydrogen concentration before cracking will initiate [15]. The local stress intensity factor at a potential cracking site will be a function of the size and nature of the cracking site and the stress state in the weld. In the GBOP test the stress state results from the welding residual stresses and is equal to the yield stress of the weld metal [16]. Since the yield stress of the FCA weld metal is lower than the yield stress of the MMA or SA weld metal (Table 3), the residual stress, and the local stress intensity factor for a given potential cracking site, will be lower in the FCA weld metal.



(a)



(b)

Figure 5. (a) GBOP results for the SAW, MMAW and FCAW consumables and (b) the corresponding HAZ and weld metal hardness.

The strength will also affect the inherent resistance of the weld metal to the “embrittling” effects of hydrogen. The greater the resistance of a weld metal to hydrogen embrittlement, the higher the hydrogen concentration required to initiate cracking at the potential cracking sites within it. It is well established that the resistance to hydrogen embrittlement is inversely related to strength: lower strength metals being more resistant to hydrogen embrittlement [17,18]. For example, Figure 6 illustrates the effect of strength on the stress intensity at which HIC occurs in a 2.25Cr-1Mo steel. From the above arguments it follows that the lower yield strength of the FCA weld metal will contribute to its greater HIC resistance both by reducing the residual stress in the weld metal and by improving its inherent resistance to hydrogen embrittlement.

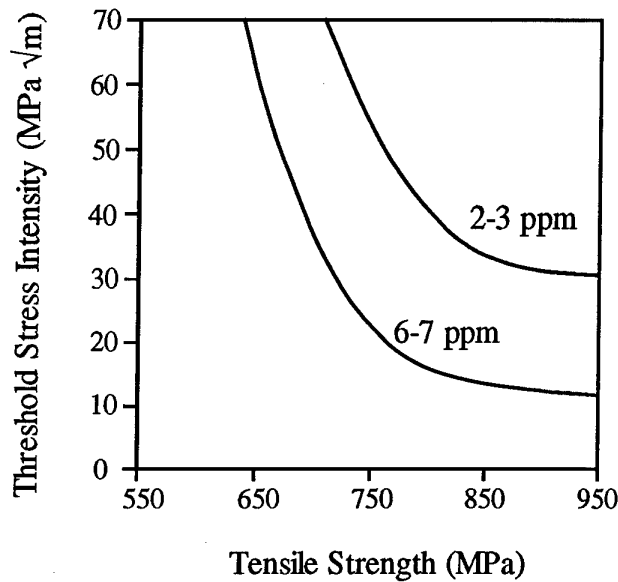


Figure 6. The threshold stress intensity at which hydrogen-assisted cracking occurs in a 2.25Cr-1Mo steel charged 2-3 and 6-7 ppm hydrogen (after ref. [19]).

4.2 Correlation between weld metal HIC resistance and toughness

At first glance, the present results, which show that the FCA weld metal has superior HIC resistance despite its lower toughness, seem to be somewhat paradoxical. This is because conventional wisdom suggests that HIC resistance and weld metal toughness are intimately related: good toughness supposedly conferring good resistance to HIC. This tenet is widely accepted throughout industry, is often implied in the literature [20] or is stated explicitly [3, 21, 22, 23]. It is important to reconcile this paradox, to allow for a scientifically rational selection of welding consumables for maximum HIC resistance.

If the factors influencing HIC are examined critically it can be confidently deduced that there is no scientific basis for a "one to one" relationship between toughness and the resistance to HIC in weld metals of this strength level. However HIC resistance and toughness are indirectly related through their qualitative mutual dependence on certain material parameters.

The reasoning which leads to this conclusion is as follows. Some parameters which control toughness affect HIC resistance while some do not. To illustrate this point each of the material parameters which influence toughness will be discussed in terms of their likely effect on the inherent resistance of a weld metal to HIC. The resistance to cleavage and ductile fracture will be treated separately in the discussion, since each occurs by different mechanisms and hence have different dependencies on material parameters. Specifically, cleavage fracture resistance, or the lower shelf energy in the Charpy transition curve, is dependent upon ferrite grain size (or colony size in the case of ferrite with aligned second phase), yield stress and the size and type of crack nuclei, whereas ductile fracture, or the upper shelf energy, depends largely on the volume fraction of non-metallic inclusions and the ultimate tensile strength (UTS) [24].

4.2.1 Ferrite grain size

Since cleavage fracture in ferrite occurs along {100} planes, a fracture must change direction when it crosses a high angle boundary between adjacent grains of acicular ferrite. The smaller the grain size the more tortuous the fracture path and the greater the resistance to cleavage fracture. [24]. The upper shelf energy, is not controlled directly by the ferrite grain size.

Hart observed a decrease in hydrogen induced cracking resistance with a decrease in ferrite grain size in a weld metal where the crack morphology was transgranular [3]. However, it is unlikely that the ferrite grain size will affect HIC in the weld metals under investigation since HIC in weld metals above 690 MPa yield strength occurs as an intergranular fracture along prior-austenite grain boundaries: a mode of failure unaffected by ferrite grain size.

4.2.2 Non-metallic inclusions

Although non-metallic inclusions are known to influence the resistance to HIC as well as the resistance to both cleavage and ductile fracture, distinctions can be made between the way in which inclusions affect the three failure modes. Cleavage fracture resistance is reduced by inclusions at the upper end of the inclusion size distribution ($>1\mu\text{m}$) because they act as nuclei for cleavage fracture [25, 26]. Since ductile fracture occurs by the growth and coalescence of microvoids nucleated at inclusions, the volume fraction of inclusions is of paramount importance in controlling ductile fracture resistance [24,27,28]. The role of the larger inclusions in ductile fracture is much less significant than in cleavage fracture [29]. A series of experiments was conducted to determine the effect of weld metal oxygen content, and therefore the oxide inclusion content, on the resistance of an HSLA100 and an HY100 weld metal to

HIC [30]. With increasing oxygen content the HIC resistance of the HY100 weld metal decreased but the HIC resistance of the HSLA-100 weld metal increased despite an increase in the number of oxide inclusions on prior-austenite grain boundaries and within the prior-austenite grains. The relationship between inclusions and the resistance to HIC was obscured in the HSLA100 weld metal by a change in microstructure and a reduction in strength (as measured by hardness) which occurred with increasing oxygen content. However, the HY100 weld metal had sufficient inherent hardenability from its alloying content so that its microstructure was unchanged by the increased oxygen content. When a special quenching technique was employed to ensure consistency of microstructure and strength of the HSLA100 weld metal with increasing oxygen content, the resistance to HIC was then observed to decrease with increasing oxygen content (see Figure 7). The decrease in the HIC resistance coincided with an increase in intergranular failure, believed to be due to the increased number of oxide inclusions on prior-austenite grain boundaries. The grain boundary inclusions were believed to promote grain boundary fracture by locally concentrating the hydrogen via a hydrogen trapping mechanism and by acting as a stress raiser.

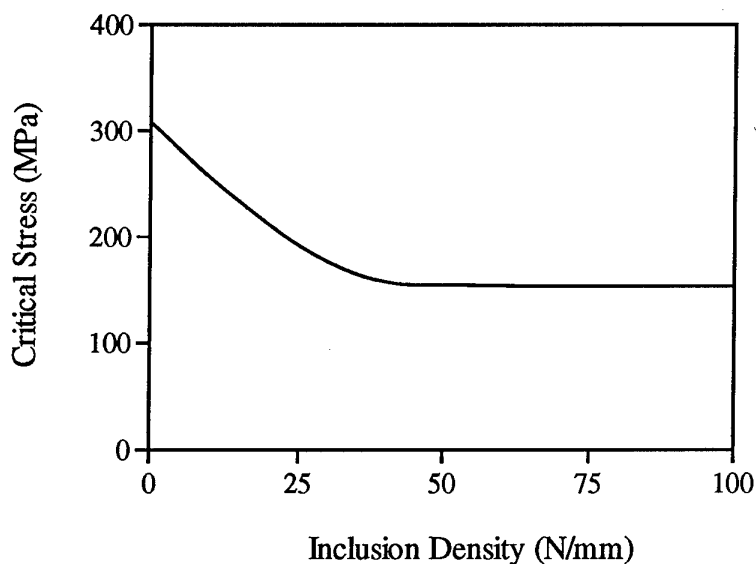


Figure 7. The critical stress for HIC in the longitudinal bead tensile restraint cracking test as a function of the grain boundary inclusion density (after [30]).

These observations bring out two important points. Firstly, that for a given microstructure and strength level, an increase in the number of oxide inclusions on prior-austenite grain boundaries decreases weld metal resistance to HIC and secondly, that the effect of inclusions is of secondary importance relative to strength.

In summary, each failure mode is related to the inclusion population in a different manner: cleavage fracture is dependent on the largest inclusions, ductile fracture is dependent on the overall volume fraction of inclusions and HIC resistance is known to be dependent on the number of inclusions on prior-austenite grain boundaries.

4.2.3 Strength

The cleavage fracture resistance, upper shelf energy and HIC resistance are all reduced by an increase in strength [31,32]. While qualitatively each of the failure modes is dependent on strength it is difficult to quantify the relationships for a direct comparison. However, a linear relationship has been found between the resistance to ductile fracture (as measured by the Charpy upper shelf energy) and the ultimate tensile strength [33], whereas there is an exponential decrease in the threshold stress intensity for HIC with increasing strength (see Figure 6). This indicates that HIC resistance is more sensitive to strength than is ductile fracture.

Table 4 is a summary of the material parameters which influence ductile fracture, cleavage fracture and HIC resistance as discussed above. It is clear from Table 4 and the above analysis that toughness and HIC resistance are not intimately related and that there is no basis for the "one to one" relationship assumed to exist between the two.

Table 4. A summary of the material parameters which influence ductile fracture, cleavage fracture and HIC resistance.

Parameter	HIC Resistance	Ductile Fracture Resistance	Cleavage Fracture Resistance
Ferrite Grain size	- dependence unlikely where fracture intergranular - proportional to d where fracture is transgranular [3]	No dependence	inversely proportional to \sqrt{d} [21]
Strength	Inverse exponential dependence on UTS (and therefore hardness) [17]	Inverse linear dependence on UTS (and therefore hardness) [31]	Inverse dependence on yield strength [21]
Oxide Inclusions	Number on prior-austenite grain boundaries[28]	Volume fraction [27]	Largest inclusions ($>1\mu\text{m}$) [22,24]

4.3 The unconservative nature of the GBOP test

It should be emphasised that the level of preheat necessary to produce crack free welding in the GBOP test does not guarantee freedom from HIC in multipass welds of a full scale structure [34]. A number of factors conspire to make multipass welds more susceptible to HIC than the single pass weld used in the GBOP test. In particular, there is a build of hydrogen and stress with successive layers in a multipass weld (see Figure 8). Although yield strength level longitudinal stresses occur in the GBOP test there will be greater triaxiality of the stress within the multipass weld due to the

restraint imposed by the increasing thickness of the weld deposit. Hydrogen-induced cracks have been shown to grow more quickly in plane strain than in plane stress [35]. The test should therefore only be used to rank different process and consumable combinations in terms of their resistance to transverse weld metal HIC.

Moreover, the propensity for HIC of any process or consumable will depend on the application in which it is used. For example, if the FCAW and MMAW consumables were used in the same application, then the HIC resistance of the FCA weld would be better than the MMA weld. However, if the FCAW consumable were used in an application with higher restraint, or shorter interpass times than those applications in which the MMAW consumable were used, then the propensity for HIC of the FCA weld might well be greater due to the onerous conditions under which it was deposited.

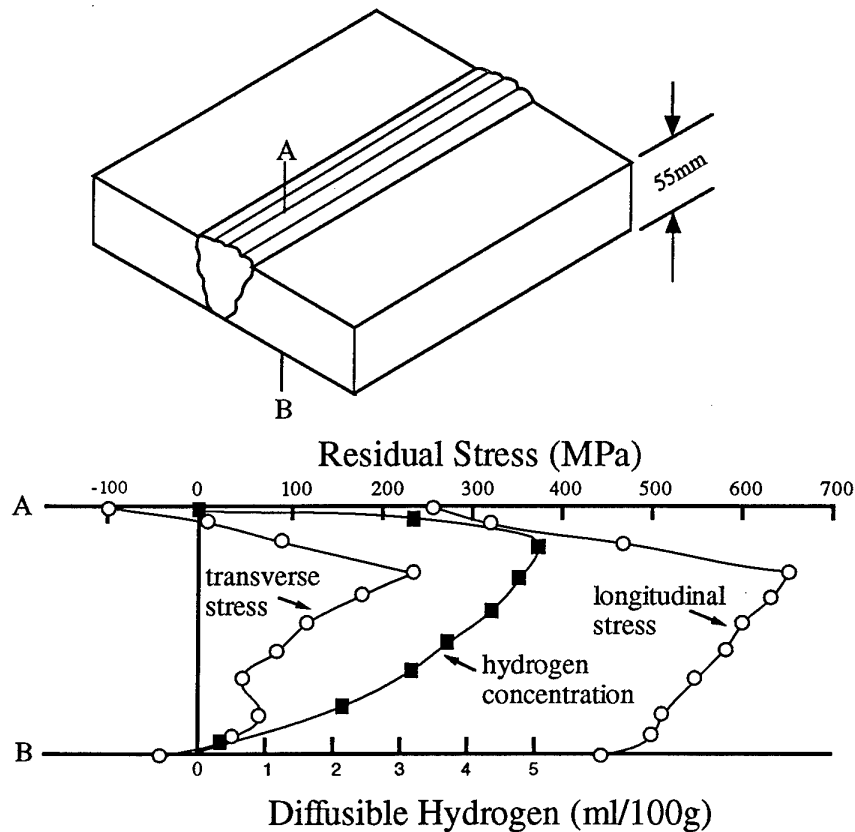


Figure 8. Residual stresses and hydrogen concentration as a function of vertical position in a butt weld in a 690 MPa yield strength steel. Note that the greatest concentration of hydrogen nearly coincides with the position of maximum longitudinal stress (After Ref. [36]).

4.4 HAZ hardness as a weld metal HIC indicator

Numerous build specifications in ship and submarine construction set upper limits on HAZ hardness. However, the hardness of the HAZs produced by each of the processes offers no information regarding the likelihood of weld metal HIC since the HAZ hardnesses resulting from each process were similar. The HAZ hardnesses depend on the hardenability of BIS 812 EMA and the HAZ cooling rate. The former is fixed by the composition of BIS 812 EMA while the latter is a function of welding variables. One would expect the MMA and FCAW HAZs to be harder than the SAW HAZ due to the lower HI and therefore higher HAZ cooling rate of the MMAW and FCAW processes. However, the HAZ hardness reaches a plateau at higher cooling rates (Figure 9). Near this plateau there is a reduced sensitivity of the HAZ hardness to cooling rate. The HAZ cooling rates which would result from MMA, FCA and SAW welds at 20 °C are noted in Figure 9. Given the scatter in the hardness measurements and the reduced sensitivity of the HAZ hardness to cooling rate in this region, it is not surprising that the HAZ hardnesses of the three processes are similar. Although the HAZ hardnesses give no information regarding the likelihood of transverse weld metal cracking, HAZ hardnesses are of interest because they will highlight hard HAZ microstructures which could have a low notch toughness resulting in poor performance under shock loading.

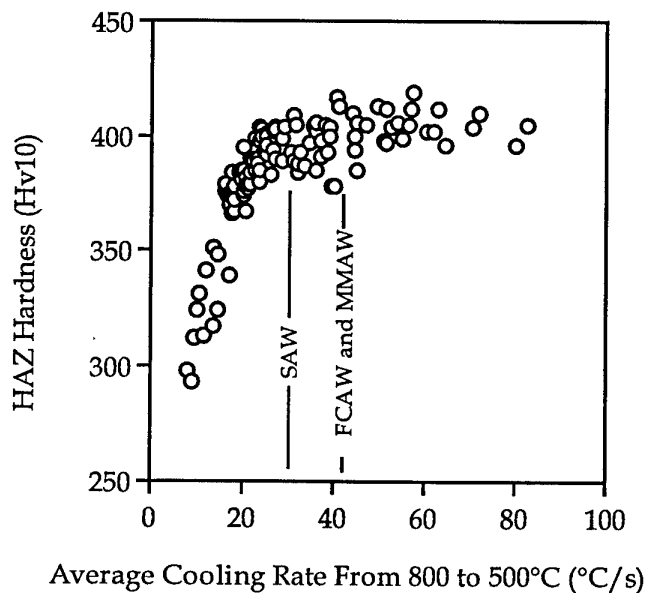


Figure 9. Measured HAZ hardness plotted as a function of the average cooling rate between 800 and 500 °C (calculated according to [10] using hardness data from [37]). The vertical lines indicate the HAZ cooling rates corresponding to SA, FCA and MMA welds deposited at 20 °C.

5. Conclusions

Using the GBOP test, the HIC resistance of a candidate FCAW consumable was compared with that of the MMAW and SAW consumables currently used in the construction of the COLLINS class submarines. As a result of these tests it is concluded that:

1. The hydrogen-induced cracking resistance of the flux cored arc welding consumable was found to be better than either the manual metal arc or the submerged arc consumables and the introduction of the flux cored arc welding consumable for use in the welding of the COLLINS pressure hull will not increase the risk of hydrogen induced cracking.
2. The lower yield strength of the FCA weld metal will contribute to its greater HIC resistance by reducing the residual stress in the weld metal and by improving its inherent resistance to hydrogen "embrittlement".
3. There is no fundamental reason for a one to one correlation between HIC resistance and toughness of a weld metal.
4. The hardness of the HAZs produced by each of the processes gives no information regarding the likelihood of weld metal HIC.

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19. ABSTRACT A MIL-121TM flux cored arc welding consumable, Alloy Rods Dualshield® II 120-M2, has recently been qualified for use in the fabrication of the pressure hull of the Royal Australian Navy COLLINS class submarines. Gapped bead on plate testing has been carried out to compare the hydrogen-induced cracking resistance of welds produced using the flux cored arc consumable with welds produced using the manual metal arc and submerged arc consumables currently used on the submarines. These tests will ensure that the introduction of the flux cored arc welding consumable does not increase the risk of hydrogen-induced cracking in the submarine pressure hull. The resistance of the flux cored arc welding consumable was found to be better than either the manual metal arc or the submerged arc consumables and the introduction of the flux cored arc welding consumable for use in the welding of the COLLINS pressure hull does not increase the risk of hydrogen induced cracking. The lower strength level of the							

deposited flux cored arc weld metal, contributed to its greater hydrogen-induced cracking resistance by reducing the residual stress state in the weld metal and by improving its inherent resistance to hydrogen "embrittlement". It is shown that there is no fundamental reason for a "one to one" relationship between hydrogen-induced cracking resistance and toughness in nominally 690 MPa yield stress weld metal. The hardness of the heat affected zones produced by each of the processes were not significantly different and gave no information regarding the likelihood of weld metal hydrogen induced cracking.

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