Effect of Welding Variables on Cracking in Cobalt-Based SMA Hardfacing Deposits

Investigation focuses on establishing guidelines for depositing crack-free hardfacing weld metal

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Introduction

Weld deposition of hardfacing alloys is commonly employed to increase the service life of components subject to abrasive wear. A number of alloys are commercially available, based largely on iron, nickel or cobalt matrices, and offering various properties in the deposit (Refs. 1, 2). In general, greater life is obtained for many applications by using deposits of higher hardness, this being obtained via the presence of hard second-phase particles, especially carbides, in the matrix. However, at high hardness levels, the tensile ductility of the hardfacing is reduced and cracking can occur as a result of welding contraction strain. Such cracking does not necessarily significantly reduce the service wear life of the component, and indeed is sometimes seen as an advantage in reducing residual stress levels (relief checking). Nonetheless, in many instances, cracking is undesirable, whether to obtain a sealing surface or to avoid fatigue failure, for example, and a requirement exists for deposition of crack-free, high-hardness surfacing.

Cracking can arise either in the solid state because of low tensile ductility or during solidification. The latter mechanism of cracking can normally be overcome by reducing travel speed, with attention to arc extinction procedure to avoid crater cracking, but the former crack type represents a rather more intractable problem in hardfacing alloys. Essentially, the incidence of cracking can be related to the tensile ductility of the deposit, and hence, to its composition and hardness. While cracking can be avoided by selection of an alternative consumable composition, this will generally involve a reduction in deposit hardness, which may be unacceptable in terms of service properties. Where such a material change is inapplicable, the most common preventative measure is to apply preheat (Refs. 1, 2), on the basis that the cooling rate after welding can be reduced with a concomitant reduction in the differential contraction strain between the cladding and the substrate. However, few quantitative data have been published on the effects of welding conditions on the cracking (or service) behavior of weld-deposited hardfacing (Ref. 3); thus, unless prior practical experience exists, substantial procedure development is commonly necessary to achieve crack-free hardfacing.

The present program was initiated to examine the effects of varying welding conditions on the cracking sensitivity of weld-deposited hardfacing. A particular objective was to explore the feasibility of producing nomograms affording general guidance on welding procedures to avoid cracking with different hardfacing alloys so that procedural development trials can be minimized. Shielded metal arc (SMA) welding was chosen to make deposits on steel using consumables meeting AWS A5.13 ECoCr-B specification, an excellent hardfacing alloy but one known for its tendency to crack on cooling. This combination of process and material is widely used in hardfacing applications, especially on site.

KEY WORDS

ECoCr-B
Hardfacing Alloys
Welding Variables
Cracking
Tensile Ductility
Nomograms
SMAW
Co-Based Consumable
Wear Resistance
Abrasive Wear

Approach

In essence, the risk of cracking in the hardfacing is governed by the tensile ductility of the deposit and by the applied shrinkage strain (Refs. 2, 3). The former is dependent on the material composition and microstructure, and the latter on composition and welding conditions, especially, from practical experience, on the preheat level (Ref. 3). Tests were therefore carried out varying deposit dilution and preheat temperature, changes in dilution and composition being achieved by altering the welding current with single- and double-layer deposition.

Test welds were deposited circumferential onto a steel bar of 100-mm (4-in.) diameter and 300-mm (12-in.) length. This geometry was selected as constituting a semi-infinite heat sink of fairly high restraint. Preliminary tests showed that differentiation could be made between the incidence of cracking in the ECoCr-B deposits produced with varying welding current.

Experimental Procedure

Materials

The hardfacing material used was in the form of SMA electrodes of 4-mm (0.16-in.) diameter, obtained to the AWS A5.13 ECoCrB specification. The substrate bar material was 0.4% carbon steel 080 A42 (Ref. 4) (Table 1), as representative of components for which such hardfacing might be employed in practice.

Deposition and Welding Conditions

Deposits were made manually by welding in the flat position onto the bar

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Table 1—Material Analysis

Material	Element (wt-%) Fe	C	S	P	Si	Mn	Ni	Cr	Mo	Cu	Co	W
ECoCr-B[a]	2.38	1.70	—[c]	—[c]	1.02	—[c]	2.47	31.5	0.20	—[c]	Bal	8.70
080 A42[b]	Bal	0.41	0.021	0.012	0.19	0.82	0.22	0.23	0.07	0.24	0.02	—[c]

(a) Manufacturer's analysis.
(b) TWI Ref. No. S/85/205.
(c) Not determined.

Table 2—Summary of Effects of Welding Conditions on Incidence of Cracking in the Deposited Hardfacing

Preheat (°C)	Current (A)	Nominal Heat Input (kJ/mm)	Cracking 1st Layer	2nd Layer
20	100	1	C[a], NC[b]	C, C
	120	1.2	NC	C
	150	1.6	NC	C
40	100	1	C	C
	150	1.6	NC	C
70	120	1.2	NC	C
	150	1.6	NC	C
100	100	1	NC, C	C, C
	120	1.2	NC	C
	150	1.6	NC	C
150	100	1	NC	C
	120	1.2	NC	C
	150	1.6	NC	C
200	100	1	NC	C, NC
	150	1.6	NC	C
250	120	1.2	NC	NC
	150	1.6	NC	NC
100	100	1	C	—[c]
250	100	1	—[c]	C
300	160	1.8	NC	—[c]
200	100	1	—[c]	NC

(a) Cracked.
(b) Not cracked.
(c) Not determined.

rotating at a surface (*i.e.*, travel) speed of 130 mm/min (5 in./min). Single-layer deposits consisted of three beads with approximately 25% overlap between each, and for double-layer samples, a further two beads were deposited on the original three beads.

Welds were made at preheats ranging from 20° to 300°C (68° to 572°F) for currents between 100 and 160 A at voltages between 21 and 24 V, as summarized in Table 2. Welding was performed using DC, electrode positive conditions. In most cases, the same nominal conditions were used for both layers. The preheat was applied by placing the test bar in a furnace at the required temperature, and it was maintained as an interpass temperature for deposition of adjacent beads. On completion of welding of each layer, the test piece was allowed to cool in air to room temperature. Some tests under specific conditions were repeated to clarify behavior as necessary.

Thermocouples were harpooned into the 3rd and 5th bead weld pools during deposition, and the cooling cycle down to about 100°C (212°F) was recorded. Cooling rates from 800° to 500°C (1472° to 932°F) were determined. This cooling parameter was taken as corresponding to a temperature range close to that at which cracking commences during cooling (Ref. 2), and because data exist (Refs. 5, 6) to predict the effect of changing welding conditions, joint heat sink, etc., on the deposit cooling cycle.

Examination

Cracks were detected both aurally as they occurred during initial cooling and by dye penetrant testing carried out after deposition of the 3rd bead and again after the final run. Sections were taken from deposits, mounted, ground and polished to a 1 μm finish. They were examined under an optical microscope to assess deposit microstructure and as a further check on the incidence of cracking.

To determine the extent of dilution, the amount of iron present in the deposit was measured by using energy dispersive x-ray analysis in conjunction with a scanning electron microscope. Analyses were taken in each of beads 3 and 5.

Hardness measurements on transverse sections were made using a Vickers pyramidal diamond indentor under a load of 5 kg.

Results

Material Microstructure and Cracking Behavior

Figures 1 and 2 show representative deposits and cracking. Both the first and second layers showed primary solidification to the metallic α-phase, with subsequent formation of interdendritic carbides, as in Fig. 3. The carbide content in the second layer was, however, very much higher than in the first beads.

The cracking observed was virtually all transverse to the weld bead, roughly perpendicular to the substrate, and had formed apparently randomly along the deposit (Figs. 1, 4). The cracking occurred with negligible plastic strain in the matrix, and developed along the carbide phase. No solidification cracking was observed.

In addition to cracking in the deposited bead, fusion line and heat-affected zone (HAZ), hydrogen cracking

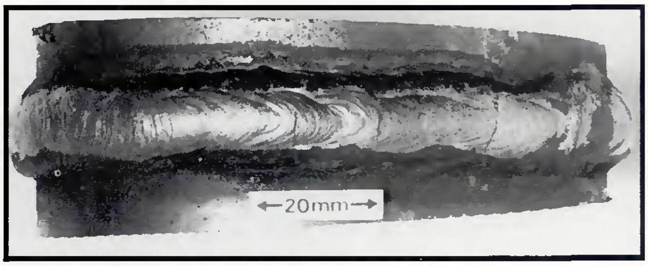

Fig. 1 — View of test weld: 1 kJ/mm and 100°C preheat.

Fig. 2 — Transverse section through test weld: 1.2 kJ/mm and 150°C preheat.

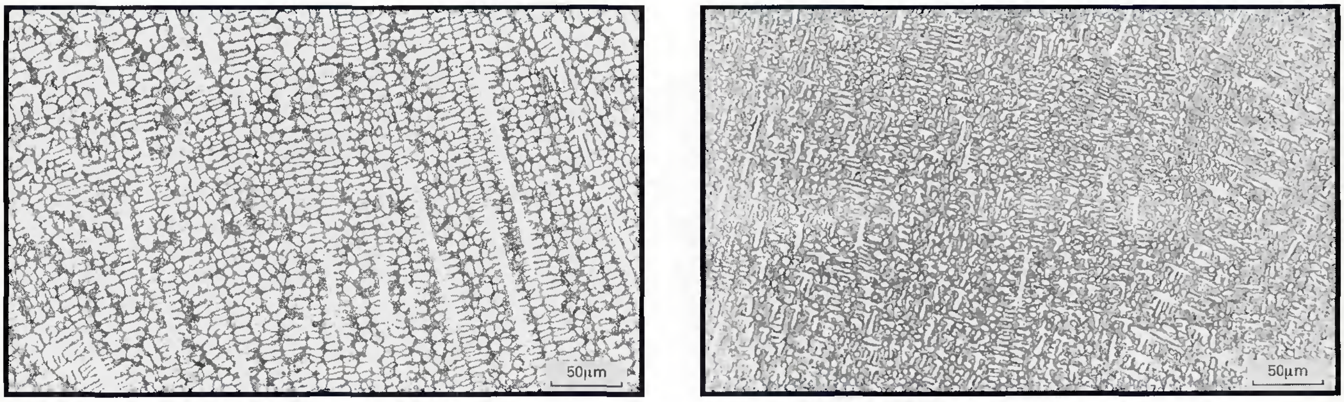

Fig. 3 — Microstructure of test weld in Fig. 1, 320X. A — Third run; B — fifth run.

was observed. Figure 5 shows cracking initiated at the deposit toe in a sample produced at 1 kJ/mm (25 kJ/in.) and 40°C (104°F) preheat. The HAZ cracks were of the toe and underbead type lying roughly parallel to the welding direction, the former sometimes being associated with cracking in the cladding material. Cracking appeared predominantly intergranular, and was largely confined to the transformed HAZ. The HAZ microstructure produced by the first layer varied from fully martensitic in the deposits with the most rapid cooling to mixed higher temperature transformation products at the longer cooling times. The HAZ cracks were observed only in the martensitic microstructures.

Effect of Welding Conditions

A summary of the effects of current and preheat on deposit cracking is given in Table 2. For both layers, cracking in the hardfacing was reduced by increasing current or preheat temperature, and cracking occurred mainly in the second layer, requiring a substantial increase in preheat level for its avoidance. As illustrated in Fig. 6, cooling times from 800° to 500°C (Δt_{8-5}) increased at higher preheat temperatures, especially with preheat to above 200°C (392°F). No particular difficulties were experienced with slag removal, and there were no indications that residual slag contributed to cracking in the second layer.

The HAZ cracking occurred only in deposits produced with preheat levels up to 100°C. Results of hardness measurements on typical cracked and uncracked HAZs are given in Table 3. The highest hardness was found with 20°C preheat, but, especially in the other welds, it must be presumed that some tempering and softening from the as-welded hardness had occurred during subsequent preheating and deposition of the second layer.

Dilution

Dilution as assessed by the iron content of the deposit increased with increasing current and, to a lesser extent, preheat — Fig. 7. The deposit hardness measurements are plotted against dilution in Fig. 8. Hardness fell at higher dilution levels. No evidence of deposit cracking around hardness indentations was seen.

Discussion

Effect of Welding Conditions

Cracking occurs in a deposit as a result of the strains set up not only by unequal cooling rates within the deposit, but also by any expansion mismatch between deposit and substrate. In the case of hardfacing, the problem is exacerbated by the high material strength over a range of temperatures that resists accommodation of shrinkage strain. It follows that any degree of dilution of a ECoCr-B or similar alloy by a steel sub-

Table 3—Representative Results of HAZ Hardness Measurements

Preheat (°C)	Current (A)	Nominal Heat Input (kJ/mm)	HAZ Cracking	Maximum HAZ Hardness HV5[(a)]
20	100	1	C[(b)]	473
100	100	1	C	349
150	100	1	NC[(c)]	289
300	160	1.8	NC	317

(a) Vickers hardness with 5 kg indenting load.
(b) Cracked.
(c) Not cracked.

Fig. 4 — Typical deposit cracking, 100X.

Fig. 5 — HAZ cracking from deposit toe, 50X.

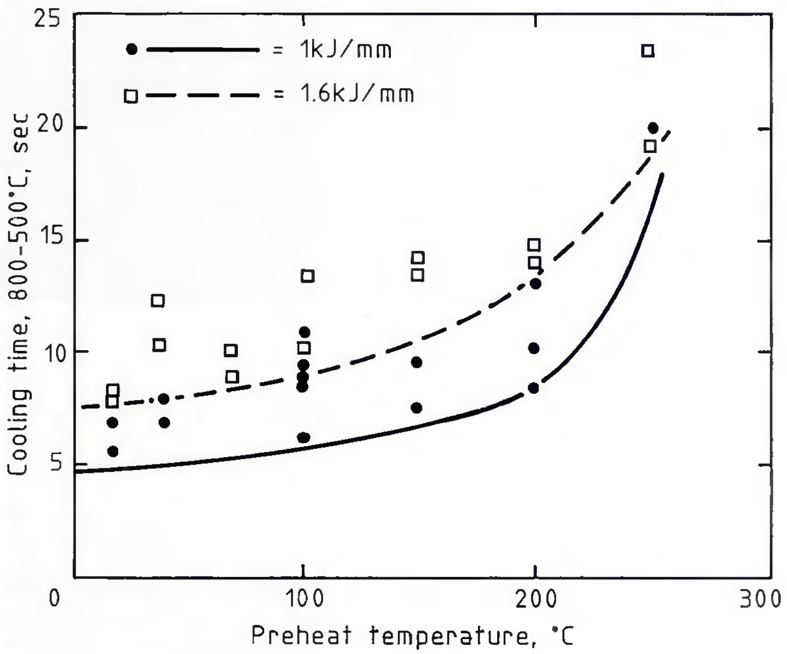

Fig. 6 — Effect of preheat temperature on deposit cooling times, 800°–500°C, for 1 and 1.6 kJ/mm, with bounding lines.

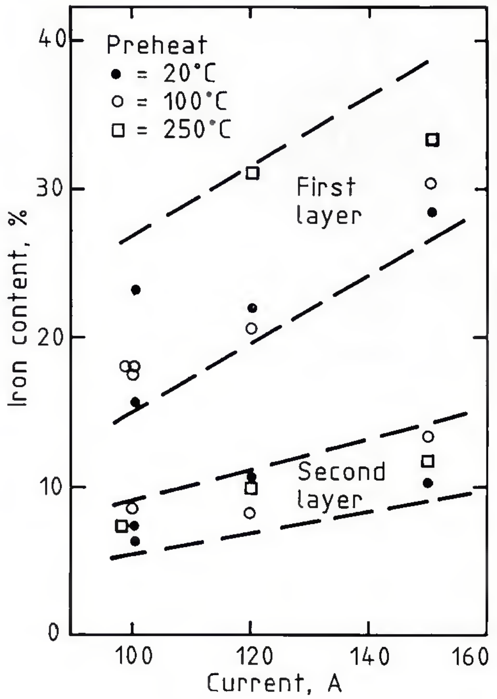

Fig. 7 — Effect of current on dilution, measured as iron content of the deposit.

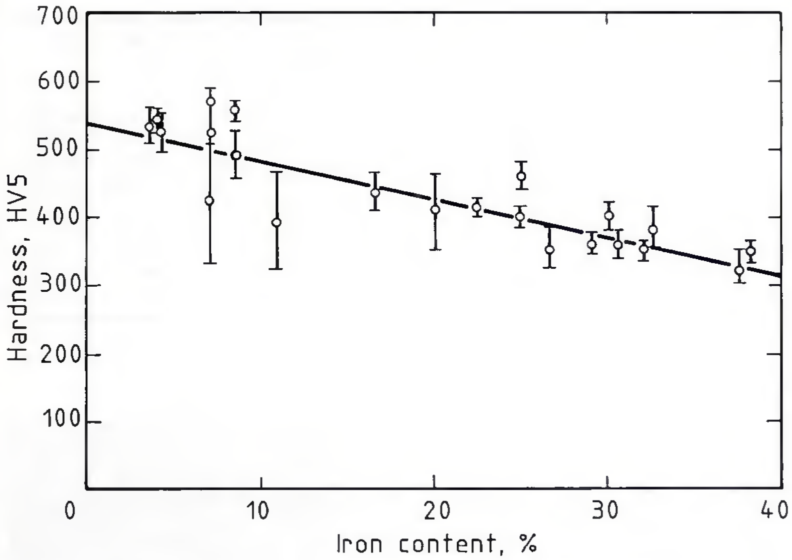

Fig. 8 — Effect of deposit dilution, measured as iron content, on hardness.

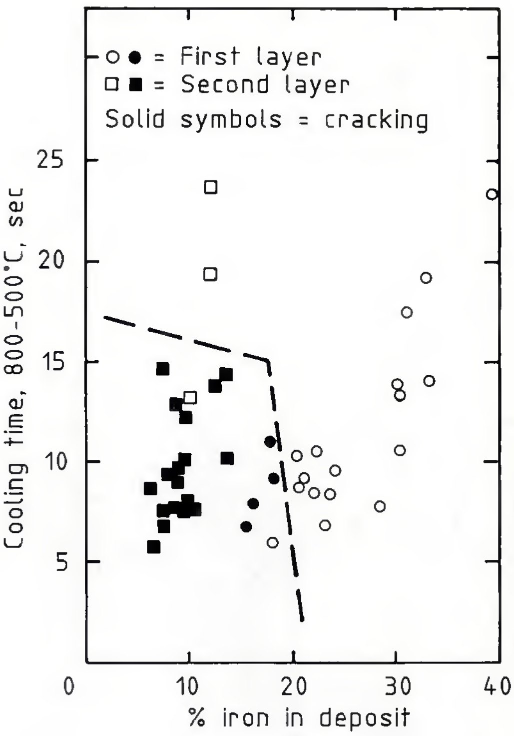

Fig. 9 — Effect of cooling time from 800° – 500°C on deposit cracking.

strate will be particularly important in avoiding cracking because dilution both reduces compositional (and to a large extent expansion) mismatch and also makes a softer and more accommodating deposit by decreasing the carbide content. In principle, the overall situation is made more complex when a second layer of deposit is added because compositional differences between the new layer and its substrate become much less, but so does the degree of accommodation required. In addition, mismatch strains generated in the first layer will, to some extent, be transmitted to the second layer, but the concept of tolerance to shrinkage strains being determined by dilution should remain valid.

It was further argued above that the effective applied strain and hence cracking risk is influenced by preheat level and subsequent cooling rate. Certainly this is indicated by practical experience with the proviso that a general preheat is applied to the substrate so that severe temperature gradients do not exist.

On this basis, the present data were plotted as in Fig. 9, considering dilution and cooling conditions. It can be seen that a bounding curve can be identified, dividing cracked and uncracked deposits. There would seem to be a distinction between the first and second layer results, as discussed further below, but it is considered that a diagram of this type could well form the basis for a nomogram system to define welding conditions giving crack-free deposits.

Such a nomogram should be applicable to hardfacing systems generally, and not only to the ECoCr-B consumables utilized in the present work. However, the practical usage of data in the form of Fig. 9 will require consideration of a number of aspects of behavior. In the first instance, the control of dilution is particularly important in hardfacing as it markedly affects the wear and corrosion resistance of the deposit (Refs. 3, 4, 7). Low levels of iron dilution may be acceptable (as is the case with ECoCr-B), but in general, the greater the percentage of iron in a hardfacing layer, the poorer its properties will be. The tendency for increased dilution to decrease the likelihood of cracking may lead to the use of welding conditions causing higher dilutions than are compatible with the service conditions of the component, when maximum performance will not be given by the hardfacing deposit. Hence, when employing Fig. 9, a decision must be taken as to the permissible dilution level. If, for example, deposit hardness can be taken as a useful guide to the resistance of a surface undergoing low-load abrasive wear (Ref. 3), data as in Fig. 8 may be applicable, but will need to be generated for particular hardfacing/substrate combinations.

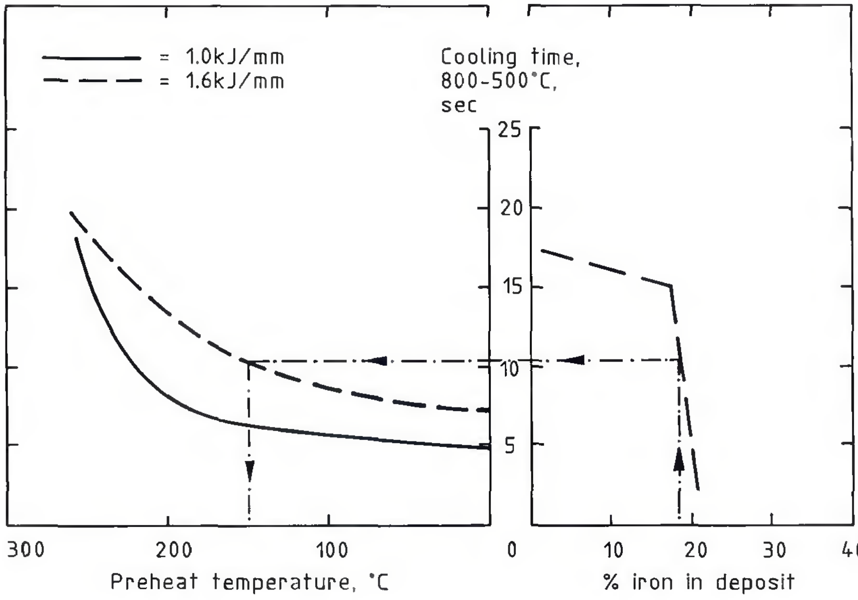

Fig. 10 — Proposed nomogram for derivation of SMA welding conditions giving a specified iron content in the deposit, with no brittle hardfacing cracking.

It will be necessary also to define the cooling rate and dilution experienced in any particular hardfacing operation. Some guidance on the latter is given in Fig. 7, but clearly more information is needed. In regard to cooling rate, a number of nomograms exist (Refs. 5, 6) for prediction of Δt_{8-5}. For infinite heat sinks, and using the bounding lines from Figs. 6 and 9, it is possible to develop a nomogram construction as in Fig. 10, whereby the cooling rate necessary to avoid cracking at the required dilution is directly related to welding conditions.

At present, the nomogram must be regarded as only a tentative proposal, and substantially more testing is essential to define its practical viability. Indeed, a number of reservations must be expressed. First, Fig. 9 was obtained for a specific material/welding process combination and for only one consumable batch; whereas, other hardfacing alloys must be expected to show a different cracking response with changes in dilution, depending on the composition involved. Further, the use of dilution as an input presupposes that process and welding conditions influence cracking only by determining the depth of substrate penetration. For practical purposes, this may be true but appropriate study of such welding variables is required. Moreover, it is assumed that the cracking test piece geometry used is relevant to service. Even if this is the case, Fig. 10 is limited to substrate geometries which act as a semi-infinite heat sink and allow no accommodation of residual stresses by distortion. In this last respect, the nomogram will err on the conservative side.

Hardfacing Cracking Behavior

The very high strength and low ductility of the deposit from ECoCr-B electrodes, and its great sensitivity to dilution, dominate the behavior of a deposit almost to the exclusion of other factors and are responsible for the somewhat different response of the first and second layers. It was difficult to generate cracks in the first layer, but the crack/no-crack boundary appears to approach a near-vertical line. This implies that dilution was the main factor influencing crack formation in the first layer, presumably via its effect on the carbide content, and that about 20% was the critical level, irrespective of cooling rate insofar as this was an independent variable. Thus, from Fig. 8, any first layer deposit with a hardness below say 450 HV would be expected to be crack-free almost without regard to its conditions of deposition, noting that only one consumable was tested and given that the present level of restraint was sufficient to represent a worst case. In principle, deposit cracking could be influenced also by the volume expansion associated with transformation to martensite in the substrate. However, since cooling rate had little effect on first layer cracking (Fig. 9), such an effect is unlikely to be of particular significance.

On the other hand, dilution of the second layer was restricted to low levels. Here, the crack/no-crack boundary would appear to be nearly horizontal, that is, cooling rate was the controlling factor with dilution having much less effect. It is difficult or impossible to achieve such low dilution in a single layer with the SMA process, but this is

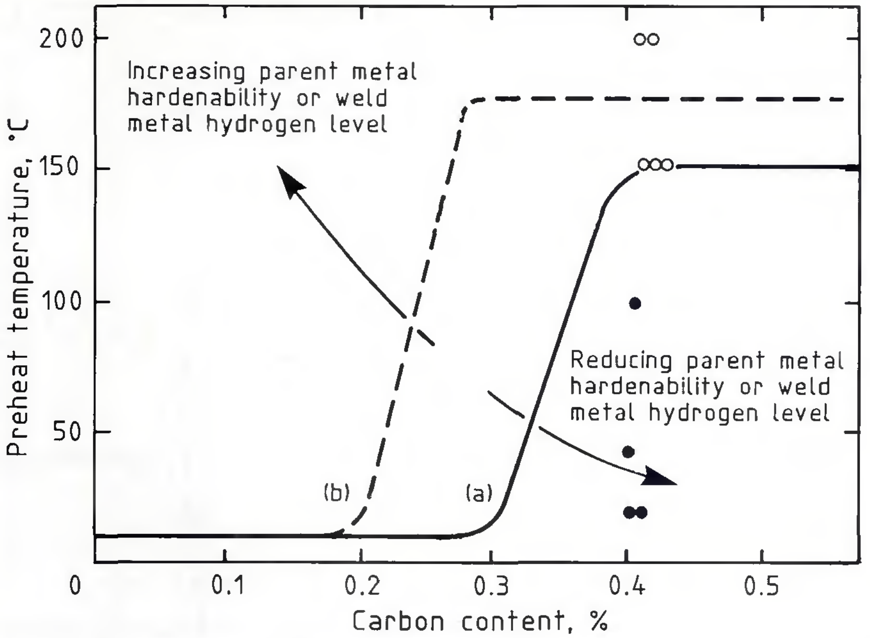

Fig. 11 — Guide to preheat temperatures using austenitic SMA electrodes at about 1 to 2 kJ/mm to avoid HAZ hydrogen cracking when welding ferritic steels (Ref. 10). A — Low restraint; B — high restraint. The present results are indicated as: = not cracked, = cracked.

readily obtained with alternative deposition methods, for example, by plasma transferred arc surfacing (Ref. 8), and a study of low-dilution first layers produced by other processes is necessary to indicate the general applicability (or otherwise) of Figs. 9 and 10.

It should also be noted that the present investigation has concentrated on brittle, solid-state cracking. Solidification cracking was not found to be a particular problem, but may require further attention in ECoCr-B deposits at very high dilution levels or in alternative hardfacing materials.

Heat-Affected Zone Cracking

The location and morphology of the HAZ cracking observed indicate this stems from hydrogen embrittlement. Clearly, this problem must be recognized in any scheme intended to give guidance on welding procedures for production of crack-free hardfaced components. Virtually no data exist on weld metal hydrogen levels associated with cobalt-based consumables, but in terms of hydrogen solubility and diffusion rate, a close parallel can be drawn with austenitic stainless steel and nickel alloy electrodes (Ref. 9). The risk of HAZ hydrogen cracking using stainless steel consumables was discussed by Gooch (Ref. 10) in terms of base metal transformation behavior and consumable hydrogen level, and it is considered probable that the guidelines proposed would be generally applicable to cobalt-based SMA electrodes as presently employed.

The maximum HAZ hardness in Table 3 of over 470 HV is high enough to engender significant sensitivity to hydrogen cracking. In plain carbon steels such as 080 A42, some control over HAZ microstructure and hardness can be achieved by varying welding conditions. However, in SMA hardfacing, the heat input employed may be restricted by the need for positional welding. In such cases, reliance must be placed on the use of preheat, especially to allow hydrogen diffusion away from the deposit while the material is at sufficiently high temperature for hydrogen embrittlement to be negligible. In this regard, reference can be made to Fig. 11, derived from welds in transformable ferritic steels made using austenitic SMA consumables. The present HAZ cracking results are shown, and it can be seen that the observed behavior is predicted well by the diagram.

Summary and Conclusions

Study has been carried out on the sensitivity of weld deposited hardfacing to cracking stemming from low tensile ductility of the deposit. Cobalt-based alloy to AWS A5.13 Grade ECoCr-B was deposited onto a 0.4% C steel by the shielded metal arc process, with varying current and preheat levels. The following conclusions were reached.

1) Cracking in the hardfacing was reduced by increasing current and preheat temperature for both single- and two-layer deposits.

2) Sensitivity to deposit cracking was substantially higher in the second layer than the first, as a result of lower dilution and higher deposit hardness.

3) Bounding conditions for deposit cracking were defined in terms of deposit dilution and cooling rate. The approach is proposed as a basis for a nomogram system to predict the risk of cracking in different hardfacing/substrate combinations.

4) For the particular electrodes and substrate steel studied, cracking of first layers was avoided by selection of welding conditions giving over 20% dilution: prevention of cracking in the second layer required a welding procedure such that a cooling time from 800° to 500°C above 20 s was obtained.

5) Heat-affected zone hydrogen cracking was observed. It is probable that existing guidelines for the avoidance of such cracking using austenitic stainless steel electrodes are applicable also to cobalt-based consumables.

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