Ferrite Vein Cracking in Electroslag Welds

Ferrite vein cracking may be avoided by using dried consumables and purging when necessary with a low humidity atmosphere, and by postweld heating to slow or interrupt cooling in the range below 200°C.

BY J. A. DAVENPORT, B.-N. QIAN, A. W. PENSE AND R. D. STOUT

ABSTRACT. An investigation of weld metal microcracking in electroslag welds indicated that it is a low temperature hydrogen-induced phenomenon which begins around 125°C (250°F) during postweld cooling. Sources of hydrogen were found to include moist flux, moist atmosphere and the filler wire. Commercially available wire containing 20 ppm hydrogen consistently produced microcracking when welded under a 53°C (128°F) saturated air atmosphere using neutral flux. The amount of external restraint on the weldment was a major factor in the extent of microcracking.

Charpy tests on electroslag weld metal indicated that microcracked and crack-free weld metals have comparable impact energy absorption at −18 and at 5°C (0 and 41°F). The tensile test yield and ultimate strengths were also unchanged by the presence of microcracks but the elongation was reduced.

Introduction

The electroslag welding process is a rapid deposition high heat input process in which the weld is made vertically while it is shielded by a blanket of molten slag. Despite the potential for excellent deposits, problems have been encountered with the integrity of welds in large structures such as bridges. One problem has been the occurrence of microcracks up to 4 mm (0.160 in.) long in the weld metal. The cracks are usually located in the proeutectoid ferrite outlining the prior austenite grain boundaries, and have been identified to be hydrogen-induced cracks (Ref. 1, 6).

The objective of this AISI-WRC project was to reproduce the occurrence of microcracks in electroslag welding in the laboratory and ascertain the influence of various base material, filler wire and flux combinations. A better understanding of the cracking mechanism was sought to reveal methods for alleviating microcracking defects in production weldments.

Experimental Program

Weldment Configuration

Shown in Fig. 1 is the size weldment used for most of the 39 welds made. The only major variation from this configuration was the use of smaller strongbacks (13 mm vs. 51 mm thick) for several "low" restraint weldments.

When a controlled atmosphere was to be injected above the molten slag during welding, a 4.8 mm (1/6 in.) diameter tube was strapped to the guide tube. The saturated atmosphere which the injection tube often carried was produced by bubbling air through a flask of water heated to a selected temperature. The same injection tube arrangement was used for furnishing an atmosphere above the weld of argon or hydrogen from compressed gas cylinders.

All welds were deposited at 575 amperes (A) and 38 volts (V) DCRP. The cooling water was normally left on for 1 to 2 hours (h) after a weld was completed by which time the weldment temperature was below 75°C (167°F). A postweld aging of at least three days at room conditions was observed before sectioning of the weld was begun.

Test Plates and Consumables

Three different grades of 51 mm (2 in.) thick plates were utilized for the experiments. Listed in Table 1 are chemical analyses of the A572-50, A588-B and A36 steel plates used. Table 2 shows the composition and basicity of the three commercial fluxes utilized. CaF₂ was treated as fully basic in the basicity calculations.

The composition of the undiluted deposits for the 2.5 mm (1/32 in.) diameter filler metals used are indicated in Table 3. All of them are commercially available. Filler metal "A" is a cored wire specifically developed for electroslag welding and is intended to give high impact toughness. Filler metal "B" is a solid carbon steel wire while filler metal "C" is a cored wire which closely matches the chemistry of...
Table 3—Undiluted Filler Metal Baseplate Compositions—Manufacturer's Data, Wt-%

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Table 2—Flux Compositions, Wt-%

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Notes:
- A36 = product analysis
- A588B = heat analysis
- A572-50 = heat analysis
- NT = no test either made and/or reported
Table 4—Relation Between Welding Conditions and Degree of Cracking

<table>
<thead>
<tr>
<th>Welding Conditions</th>
<th>Cracking</th>
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<tr>
<td>Saturated air @ 33°C (128°F), dry flux</td>
<td>Moderate, major or severe</td>
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<tr>
<td>Ambient atmosphere, flux 10% water</td>
<td>Moderate or major</td>
</tr>
<tr>
<td>Saturated air @ 40°C (104°F), dry flux</td>
<td>Slight</td>
</tr>
<tr>
<td>Ambient atmosphere, dry flux</td>
<td>Very slight or slight</td>
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<tr>
<td>Argon atmosphere, baked wire, dry flux</td>
<td>None</td>
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Table 5. Results and Discussion

Moisture Additions to the Atmosphere

The use of hydrogen or moisture-containing atmospheres above the molten slag during welding induced microcracking in the ferrite veins. Microcracking could also be induced by utilizing a wet flux (10% water by weight). Comparison of the cracking obtained in numerous welds, all of which utilized the high restraint configuration, neutral flux ("A") and the high Mn cored wire ("A"), relates the degree of cracking to the supply of hydrogen as indicated in Table 4. The results apply to both A572-50 and A588-B base materials. A limited number of tests indicated that welds in A36 steel follow these same cracking trends. Complete data on the test welds are listed in Table 5.

A hydrogen atmosphere imposed on a low restraint weldment in the early stages of the research produced a major amount of cracking, while a saturated or wet flux weldment cracked only slightly under the same low restraint conditions. The hydrogen atmosphere weld demonstrated that the form of the hydrogen (free or combined) was not critical to the formation of cracks.

Figure 2 shows weld metal microcracks typical of those induced by a moist atmosphere, wet flux or a hydrogen gas atmosphere. The moisture (i.e., hydrogen) induced cracking that resulted seems to be more crack susceptible than A588-B welds. The base metal influence on microcracking is less well defined than its influence on austenite grain size. There was a range in cracking of A572-50 welds (welds 2A28 and 2A31) from "moderate" to "severe" when welded under identical conditions. The "moderate" cracking of A588-B welds (welds 2A16 and 2A17) made under similar conditions (i.e., filler metal "A", dry flux "A", high restraint, saturated 53°C atmosphere) fell within the range of the A572-50 results. Hence the A588-B cracking level must be regarded as roughly comparable to that obtained in the A572-50, if the total number of cracks is the criterion for judging crack severity. Since the A588-B cracks were often shorter, the A572-50 tended to have a greater total crack length.

The results of this research indicated that the amount of microcracking obtained in high basicity "B" and "C" fluxes than in "A" flux weldments; it manifested itself as "wormhole" porosity rather than as cracking.

It was found that 10% moisture in "B" flux induced porosity with moist or dry atmospheres (welds 8B12 and 8B13). Flux "B" produced significant cracking with or without porosity with a 53°C (127°F) saturated atmosphere (welds 8B13 and 8B23) and flux "C" caused cracking, with some porosity, using a 40°C (104°F) saturated atmosphere (weld 2C24). Dried flux "C" with low restraint and an ambient atmosphere produced a crack-free and porosity-free weld. Neutral flux "A" did not produce porosity under any conditions.

Filler Metal

The three filler metals utilized in this work had the compositions noted in Table 3. The use of filler metal "B" with A36 plate and filler metal "C" with A588-B plate both under a 53°C saturated atmosphere using flux "A" produced no cracking (welds 6A3 and 6A4). If microalloying was a factor in the microcracking, these two welds should show different cracking propensities because of widely different base material and filler metal microalloying. When filler metal "A" was substituted in two additional welds on these steels, moderate cracking resulted.

The possibility that the filler metal could contribute to the weld metal hydrogen was suspected from these results. The results of hydrogen analyses on several sections of each filler metal (50 mm, i.e., 2 in. length of filler metal per analysis) are listed in Table 7. Filler metal "A" contained significantly more hydrogen (20 ppm) than filler metals "B" or "C".

Weld 2A23 was made with filler metal "A" which had been held in a 30-60°C (120-140°F) saturated atmosphere for two days prior to welding. No additional cracking was obtained beyond that induced by filler metal "A" as received from the manufacturer. The filler metal was apparently saturated with hydrogen-bearing compounds at normal atmospheric conditions.
cracking obtained was the same as that obtained in matching welds exposed to the laboratory atmosphere. The high hydrogen content of the filler metal explains why cracking could not be eliminated only by controlling the flux and atmosphere moisture levels. This residual cracking under ambient atmosphere and dry flux was not reported by L. S. Scollo in its work; here a different wire which was not cored was used (Ref. 1).

A second weld (weld 2A38) further verified that as-received wire "A" was able to induce cracking. The wire was baked at 425°C (800°F) for 1½ hours and then used within 25 minutes (min) of removal from the furnace for making a weld under an argon atmosphere. No cracking was obtained, and it was found that the hydrogen content was lowered from 20 to 2 ppm in the filler metal by the baking.

It was earlier determined that filler metal "A" would induce cracking using a
neutral flux and a 53°C (127°F) saturated atmosphere. Filler metals “B” and “C” did not induce cracking under the same circumstances. The microalloying effect of different filler metals on the hydrogen cracking susceptibility, if there are any, were overshadowed by the wire hydrogen content.

An additional experiment (welds 2A47 and 2A48) in which filler metal B was charged electrolytically with hydrogen before welding on A572 plate, produced major cracking, while the as-received filler metal produced none. This result proved that hydrogen carried in by the filler metal was especially effective in inducing vein cracking.

It is reasonable to assume that the combination of atmosphere hydrogen and wire hydrogen must reach a certain critical level, depending on other parameters (e.g., cooling rate) to cause cracking. Filler metals “B” and “C” could be expected to induce cracking with a hydrogen potential in the atmosphere greater than that required for “A” but less than that required for a baked filler metal having 2 ppm hydrogen.

Crack Morphology

With rare exception the cracks completely followed the proeutectoid ferrite along the prior austenite grain boundaries. The distribution of cracks from the top to the bottom of the weld was approximately uniform. All of the cracks apparently initiated in the ferrite with <5% running out into the pearlite matrix.

Cracks departing from the proeutectoid ferrite were more common with the fine grained (prior austenite) A588-B than with A572-50 or A36. The crack length varied from 0.13 mm to 3.8 mm (0.005 to 0.150 in.). The cracks were always located in the weld metal at least 4.8 mm (½ in.) inside of the fusion lines and cooling shoe surfaces. This is where a residual tensile stress would be anticipated. Flanagan’s work with hydrogen microcracking in shielded metal arc welding yielded a similar result with respect to crack location (Ref. 2).

In a marginal cracking situation with A572-50 base material, cracks were found in the coarse austenite grain boundaries rather than the fine grain boundaries which were also present. This trend was evident in all of the A572-50 welds which had “slight” or “very slight” cracking. The minimal amount of coarse grains in A588-B did not allow observation of this same preference. The marginal condition cracks, again mostly in A572-50 base, also tended to lie parallel to the fusion lines, as viewed in a horizontal section.

Acoustic Emission

The purpose of acoustic emission monitoring was to determine when cracking occurs as the weldment cools during the postweld period. The results of the first pair of monitored welds indicated that the “severely” cracked weld exhibited a higher emission rate than the “good” (very slightly cracked) weld only in the first 20 min after the weld was completed but showed a lower rate thereafter.

Figure 3 shows the results of the second pair of monitored welds (2A30 and 2A31). The results are very similar in that the “cracked” weld again had a comparable or higher emission rate only during the initial postweld period. The total emissions after 36 h were higher for the good weld by 15–45% in both pairs. The results, which are consistent for both pairs of welds, are apparently indicative of a noise source other than weld metal cracking. Slag cracking is the most likely candidate even for the first pair of welds in which the majority of slag was manually removed with a chipping hammer before postweld monitoring was begun. Another possible noise source was plastic flow in the metal near stress raisers such as the strong-back filler welds. The use of higher frequency transducers with the second pair was apparently unsuccessful in eliminating pickup of slag cracking noise.

Initiation of Cracking

The inability of the acoustic emission work to provide information on the start of vein cracking in the weld metal led to another approach involving the interruption of cooling of a series of welds. The welds were made with A572-50 plate, flux “A” (dry), filler metal “A” and a 53°C (127°F) saturated air atmosphere. Earlier welds made under these conditions had shown “moderate” to “severe” microcracking.

Each succeeding weld in the series was allowed to cool to a lower postweld temperature before cooling was interrupted by placing the weld in a furnace for at least 12 h to reheat or to maintain the selected minimum temperature. Earlier work by others had shown that it was possible to eliminate cracking by immediately heating to 300°C (572°F) after welding (Ref. 1, 3). The last weld of the series was one which was allowed to cool to a low enough temperature to allow cracking to take place. It is known that maintaining an elevated postweld temperature allows the hydrogen to diffuse out of the weld metal and thereby prevents cracking.

In the first test (weld 2A32), weld

<table>
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<td></td>
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<td>2A04</td>
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</table>

*Charpy test values shown are averages of three tests after excluding high and low values from a set of 5 tests; standard deviation of the impact values is 3–8 joules (2–6 ft-lb); cracking in 2A04 induced by a hydrogen atmosphere.*
cooling was interrupted at 195°C (383°F) and the weld was held at 300°C (572°F) for 12 h. No cracking occurred. Maintaining a minimum weldment temperature of 145°C (293°F) by holding in a furnace for 60 h also eliminated cracking (weld 2A33). Figure 4 shows the thermal history of this weldment measured 25 mm (1 in.) from the inside of the plate edge by chromel-alumel thermocouples located at mid-thickness. The temperature was monitored at mid-height and 25 mm (1 in.) from the top of the plate.

A similar time-temperature record was obtained for weld 2A34 except it was held at a minimum temperature of 105°C (221°F) instead of 145°C (293°F). It showed “slight” cracking.

The results indicate that cracking did not occur until the weld cooled below 145°C (293°F). An estimate of the cracking threshold for this weldment configuration (i.e., restraint), hydrogen level and cooling rate is in the vicinity of 125°C (257°F), approximately 25 min after welding was completed.

Since the hydrogen diffusion rate is both time and temperature dependent, a different cooling cycle would change the threshold temperature. An increase in hydrogen or restraint might be expected to shift the temperature to a higher value.

A difference in one or more of these parameters may account for Kunihiro and Inoue on electroslag weldments (Ref. 5); it indicated that some microcracking may continue for several days after the weld is completed.

The cracking observed in the short-aged weld was “moderate.” This indicates that welds subjected to the thermal history used here undergo a significant percentage (> 25%) of their cracking within a few hours of the weld finish but by no means all of it. This result is consistent with work on thick plate shielded metal arc welds (Ref. 8) or other electroslag weldments (Ref. 5); it indicated that some microcracking may continue for several days after the weld is completed.

**Weldment Thermal Cycle**

The test welds discussed so far were all produced in 152 mm (6 in.) wide plates. Cooling water was left running for 1 to 2 h after the weld was completed. How the thermal history of these welds differed from a production weld in a very wide plate, which would not have the cooling shoes left on, was of interest. Accordingly, two 610 mm (24 in.) wide plates were welded under conditions which had been shown to give some microcracking.

**Mechanical Tests**

In Table 8 are the results of Charpy V-notch tests on cracked and uncracked weld metal made with A572-50 base material, filler metal “A” and flux “A.” There is essentially no difference in the impact energies measured. However, examination of the specimen fracture surfaces indicated that the microcracked weld metal developed a much more faceted fracture surface. The facets may originate from vein cracks which opened when the specimen was broken. The failure of the cracks to decrease the energy absorbed suggests that the crack size was too small and the orientation was ineffective. The results indicate that microcracking of the severity present in this study cannot be revealed by the standard Charpy test.

Hydrogen-induced microcracking was observed to reduce the specimen elongation in a tensile test without substantially effecting the yield and tensile strengths. In tests of cracked (induced by weld flux) weld metal in A588-B steel the elongation decreased from 29% to 22% in a 51 mm (2 in.) gage length compared to crack-free weld metal. The cracked tensile specimens also clearly showed “fish eyes” on the fracture surface and cracks which opened up on the surface in the necking region. These tensile test results are in agreement with work done by Nishio, Hiromoto and Inoue on electroslag weld metal (Ref. 5).
reaches the cracking threshold of 125°C (257°F) slightly quicker, but then cools more slowly from 100°C (212°F) it would be difficult to know whether it is more or less susceptible to microcracking. In any event, the cracking in wide plate weld 2A35 and a similar narrow plate weld was the same ("slight"). Hence it is concluded that the results of this research work would likely have been similar if 610 mm (24 in.) wide plates had been utilized throughout.

Summary

The results of this investigation clearly indicate that hydrogen-induced cracking can be caused by a combination of high restraint and moisture. Moisture from a saturated air atmosphere between 40 and 53°C (104 and 127°F) was found to cause significant cracking when used with neutral flux, high restraint and a filler wire containing 20 ppm hydrogen. The filler metal apparently supplemented the hydrogen in the austenite and its inevitable rejection to discontinuities upon austenite transformation, but the mechanism of cracking in these welds is as inexplicable as similar cracking found in shielded metal arc weld metal.

This research has demonstrated that control over vein cracking is possible by observing certain precautions. These include utilizing:

1. Filler metal with a limited hydrogen content.
2. Reduced cooling by preheat/postheat to maintain the weldment temperature > 125°C (257°F) for several hours after welding.
3. Atmosphere control, perhaps with an inert shield, when extreme humidity is encountered.

Acknowledgment

The authors are grateful to the American Iron and Steel Institute for sponsoring the investigation reported here and to members of the Weldability Committee of the Welding Research Council for technical guidance.

References