Heat Treatment Effect on 2CrMo Joints Welded with a Nickel-Base Electrode

Study shows that high weld metal strength, rather than heat treat effects, promotes interfacial cracking, and, therefore, a new nickel-base electrode needs to be developed

BY W. K. C. JONES

ABSTRACT. This note describes an investigation into the effect of fusion boundary microstructure on low ductility cracking at the interface between a ferritic creep resisting steel, 2CrMo, and a weld made with a nickel-base electrode conforming to AWS ENiCrFe-2.*

The complex microstructures observed at the fusion boundary in this type of joint are shown to be sensitive to preheat conditions and postweld heat treatment. The results of cross-weld tensile and stress rupture tests are presented to show that the changes in microstructure which result from variations in heat treatment do not affect significantly the failure mode or the properties of the welds.

It is suggested that the low ductility interface failure is caused by the large disparity in strength between the 2CrMo steel and the ENiCrFe-2 deposited weld metal which produces shear stresses at the interface when the joint is deformed. Cracking initiates at the specimen surface where these shear stresses are maximum.

It is proposed that a nickel-base filler metal should be developed which will produce a weld that is weaker under creep conditions than existing commercial products.

*Inco-weld A

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Introduction

Nickel-base electrodes of the Inconel type are used widely to weld austenitic ferritic steel joints and are employed in current power plant designs for the welding of both pressure containing parts and structural members. The low solubility and low diffusion coefficient of carbon in nickel-base weld metal minimizes carbon diffusion across the ferritic steel/weld metal interface and thereby reduces the tendency to form a soft decarburized zone in the ferritic steel. Circumferential cracks are frequently observed in this region when austenitic stainless steel filler metals are employed. A further attraction of Inconel filler metal is that its composition can be chosen to give a thermal expansion coefficient which is intermediate between that of the ferritic and austenitic base metals so minimizing thermally induced stresses (Ref. 1).

In the Central Electricity Generating Board, in England, Inconel electrodes have been used successfully for many years in the welding of thin walled tubes of dissimilar steels (Refs. 2,3); but, when thick walled steam pipe** was welded with an ENiCrFe-3 electrode,** failures occurred at the weld interface in less than 1000 h (Ref. 4). Furthermore, although a 9CrMo/316 joint welded with the same electrode showed no damage after being thermally cycled 300 times in a manner designed to simulate service life (Ref. 5), uniaxial stress rupture tests promoted low ductility failure at the interface between ferritic creep resisting steel and the nickel-base weld metal (Ref. 6).

It is generally acknowledged that the low ductility interface failures in the Rex 500/Inconel 182 joints were caused by poor mechanical properties in the region of the interface (Refs. 7,8), although the precise mechanism is still not clear. Similar structural features occur at the fusion boundaries of other ferritic steels welded with nickel-base electrodes (Ref. 9) but the extent to which the properties of these joints are dependent upon the microstructure at the interface has not been established. In this report previous work relating the microstructure at the interface to the properties of joints welded with nickel base electrodes is briefly reviewed and some of the variables which control this microstructure are examined. The

**Rex 500 and Inconel 182 respectively. Typical chemical analyses are given in Table 1.

Table 1 — Typical Chemical Analyses of Rex 500 and Inconel 182 (wt %)

	C	Fe	Mn	Si	Ni	Cu	Cr	Mo	V	W	Nb + Ta	Ti	Co	S	P
Rex 500	0.08	B	1.0	0.5	0.5	—	6.0	1.0	0.2	0.5	—	0.25	—	0.01	0.002
Inconel 182	0.07	8.0	6.5	0.8	B	0.3	16	—	—	—	2.0	0.8	0.1	0.003	—

Table 2 — Chemical Analyses of Experimental Materials

	C	Fe	Mn	Si	Ni	Cu	Cr	Mo	V	Nb + Ta	Ti	S	P
2CrMo	0.12	B	0.56	0.21	0.22	—	1.9	1.1	0.09	—	—	0.002	0.006
Inco-Weld A	0.09	6.1	1.1	0.46	B	0.1	15.1	2.4	—	0.92	0.06	0.011	0.005
Esshette 1250	0.09	B	6.9	0.56	9.8	—	15.8	0.95	0.29	<0.01	2.0	0.012	0.026

contribution which the structure makes to the high temperature strength of the joint is also investigated.

Review of Previous Work

It has been appreciated for many years that ferritic steels welded with nickel base electrodes can be susceptible to low ductility failure at the fusion boundary (Refs. 10,11) and this has led to their rejection for some applications (Ref. 12) and to premature failure in others (Ref. 4). The main cause for concern was the apparently brittle nature of these failures which were associated with the weld interfaces, where the structures were complex. These interfacial structures have been studied by many workers (Refs. 7,8,13,14) and Eaton and Glossop (Ref. 9) have shown that they occur commonly when ferritic steels are welded with Inconel electrodes. Generally a lamellar structure is observed at the interface which is light in the as-welded condition and dark etching after heat treatment. Barford et al (Ref. 7) have reported that, at room temperature, the interfacial structure is sometimes harder and sometimes softer than the HAZ and weld metal immediately adjacent but all other workers have found it to be significantly harder than the surrounding material (Refs. 8,9,15).

The exact nature of the interfacial structure has not been definitely established although suggestions as to its identity are numerous. Eaton and Glossop (Ref. 9) suggested that it was a lamellar intermetallic precipitate, Barford et al (Ref. 7) contended that the composition gradient at the weld/base metal interface must pass through a stable $\alpha + \gamma$ field and Wood (Ref. 14) proposed that the dark etching areas were tempered martensites similar to those observed by Slater and Winn (Ref. 16) at the fusion boundary between ferritic steel and austenitic weld metal.

The cause of the low ductility failure is also not definitely established. Kent (Ref. 4) considered that the Rex 500/Inconel 182 failures were initiated by sharp cracks located at or near the fusion boundary. He suggested that the initial cracks occurred during production of the joint, probably because cracking was initiated by a defect at or near the fusion boundary during welding or postweld heat treatment. During service, propagation of cracks from existing defects would be encouraged by the imposition of operational stresses on residual welding stresses.

Barford et al (Ref. 7,13) showed that low ductility failure similar to that experienced in service could be reproduced by testing in uniaxial tension if the weld interface was inclined to the applied stress. They demonstrated that, although the failures were macroscopically brittle, cracking occurred in a ductile region of the weldment and attributed the failure to the presence of a narrow soft layer in the Rex 500 immediately adjacent to the fusion boundary. They considered that this soft zone was very thin and was caused either by decarburization or by the presence of a micro-duplex structure which had superplastic properties.

Boniszewski (Ref. 8) argued that the lamellar structure might be either weak under creep conditions or strong enough to restrict the spread of deformation which was consequently localized in a weak zone adjacent to the interface. Eaton and Glossop (Ref. 9) also suggested that the structure may have a low creep ductility which would effect service performance. However, Barford (Ref. 13) considered that the duplex structure did not have a deleterious effect on the properties of the interface and Wood (Ref. 14) suggested that the path of cracking, although close to this region, was outside it and so the structure would not effect the strength of the interface. However, he did not rule out the presence of a low creep strength zone close to the lamellar region.

There is clearly some confusion in published work on the importance of the various structural and mechanical differences of the component materials that lead to poor performance and there appears to be no systematic investigation reported of these parameters with any material combination. The program of work described below was undertaken in an attempt to elucidate the extent to which the microstructure at the fusion boundary controls the mechanical properties of the interface between a ferritic creep-resisting steel and a nickel-base weld metal. Some of the variables which determine the microstructure at the weld interface are examined and the effect of the variations in interfacial microstructure on the mechanical properties of the composite joints investigated.

Experimental Investigation

A 2CrMo-Esshete 1250 joint welded with Inco-weld A was chosen for investigation as it was known to exhibit the microstructural features required and to be prone to interfacial failure when tested under appropriate conditions. Analyses of the materials used are given in Table 2.

To determine the effects of preheat and postweld heat treatment on the microstructure at the interface, Inco-weld A beads were deposited on 2CrMo plate. In one series, different preheats were used and in each weld the preheat was allowed to fall immediately after welding when a slice was removed. The remainder of the weld was subsequently heat treated at 730 C for 1 h. On a further series of bead-on-plate deposits the preheat was maintained until stress relief. Details are given in Table 3. Subsequently, all welds were examined metallographically.

Butt welds were fabricated in 20

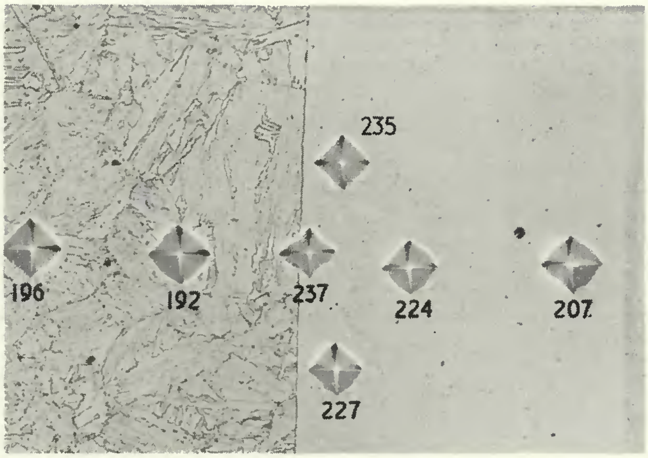

Fig. 1 — Weld interface between 2CrMo and Inco-weld A with no visible interface phase. X475, reduced 25%

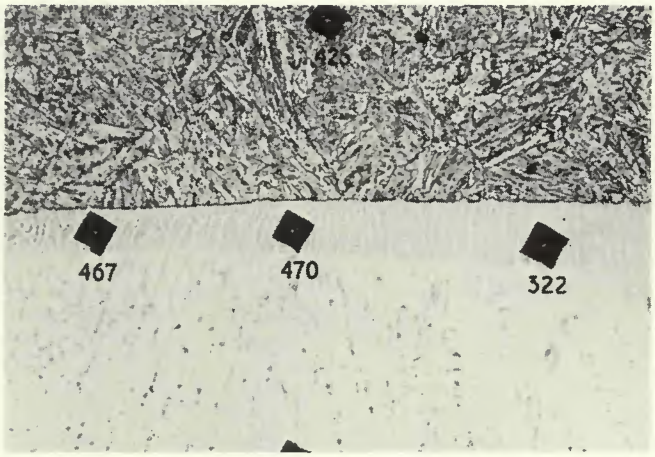

Fig. 2 — Light etching microstructure at interface in as-welded specimen. X475, reduced 25%

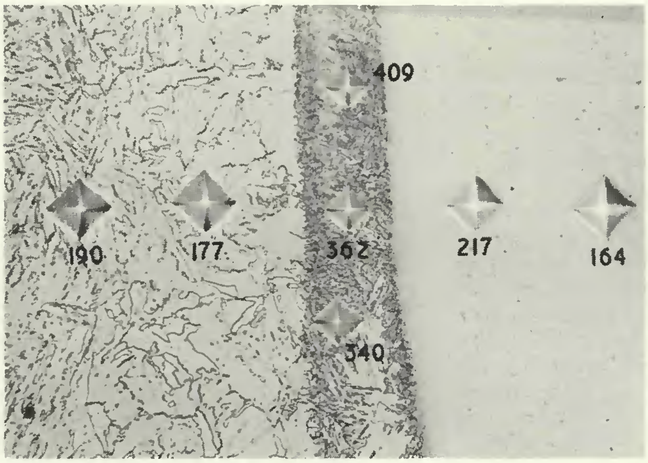

Fig. 3 — Dark etching microstructure at interface in specimen heat treated after welding at 730 C for 1 h. X475, reduced 25%

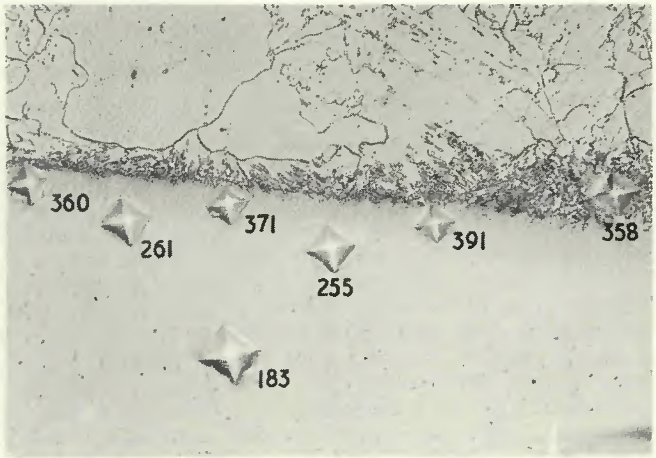

Fig. 4 — Weld heat treated at 730 C for 1 h showing both types of microstructure. X475, reduced 25%

mm (0.79 in.) thick plate and the preparations used were such that in uniaxially tested, cross-weld specimens the weld interface was inclined at either 45 or 75 deg to the tensile axis. Specimens for hot tensile and stress rupture tests and for metallographic examination were machined from the joints.

Results

Weld Interface Structures

In the as-welded condition the microstructure of the interface between 2CrMo and Inco-weld A appeared in the light microscope either as a sharply defined line (Fig. 1) or as a light etching region in the weld metal immediately adjacent to the 2CrMo (Fig. 2). This structure was most dense at fusion coves.

A dark etching structure (Fig. 3) similar to that reported by previous workers was observed in all specimens which had been heat treated after welding with the exception of those samples in which a preheat of 350 C or higher had been maintained prior to stress relief (Table 3). The light etching structure was also observed in welds where the dark etching areas occurred (Fig. 4). In those specimens in which a high preheat (>350 C) was maintained prior to stress relief only a light etching structure similar to that observed in the as-welded condition was found, but if this preheat was allowed to drop prior to heat treatment the dark etching structure was again observed at the weld interface of the stress relieved specimen.

At room temperature both structures were found to be significantly harder than the 2CrMo HAZ or the Inco-weld A weld metal, the hardness of the light etching structure usually lying in the range 400-500 HV (Fig. 2) and the dark etching structure in the range 300-400 HV (Fig. 3). The hardness of the interfacial structure was never found to be lower than that of the surrounding material which contrasts with the observations of Barford et al (Ref. 7). It is interesting to note that the latter are the only workers who report carrying out hardness measurements on sections tapered to give a lateral magnification. It is likely that this method produced inaccurate results as the zones of interest are very narrow and, in a taper section, the soft matrix beneath the interface may give rise to low hardness values.

The reaction of the interface structures to preheat and postweld heat treatment suggests that the light etching structure formed at a relatively low temperature (<350 C) and nucleated the darker etching phase. It seems probable that the light etching areas observed in the as-welded condition were martensitic (or bainitic) and the high hardnesses tend to support this hypothesis.

If the structure was martensitic a solution treatment above Ac_3 should result in its transformation to austenite, and martensite would only be reformed when the temperature of the joint fell below the M_s of the interface region. Therefore, bead-on-plate welds which had been deposited at a preheat of 50 C and then heat treated at 730 C for 1 h to produce

Table 3 — Relationship Between Preheat Temperature and Fusion Boundary Microstructure on 2CrMo/Inco-weld A Bead-on-Plate Welds Stress Relieved for 1 h at 730 C

Preheat, deg C	Preheat maintained to stress relief	Structure at fusion boundary
20	Yes	D.e.s[a]
	No	D.e.s
50	Yes	D.e.s
	No	D.e.s
100	Yes	D.e.s
	No	D.e.s
150	Yes	D.e.s
	No	D.e.s
200	Yes	D.e.s
	No	D.e.s
250	Yes	D.e.s
	No	D.e.s
300	Yes	L.e.s[b]
	No	D.e.s
350	Yes	L.e.s
	No	D.e.s
400	Yes	L.e.s
	No	D.e.s

(a) Dark etching structure (e.g., Fig. 5)
(b) Light etching structure (e.g., Fig. 6)

Table 4 — Testing Conditions[a] and Results for Cross-Weld Tensile Tests on 2CrMo/Inco-weld A/Esshete 1250 Weldments

Speciman condition	1% proof stress N/mm²	UTS N/mm²[b]	R of A, %	Elong., %	Failure location
As welded	281	355	84	19.8	2CrMo
As welded	333	362	87	25.0	2CrMo
Stress relieved 1h at 730 C	273	364	88	19.5	2CrMo
Stress relieved 1h at 730 C	277	364	89	17.5	2CrMo

(a) Temperature of test, 580 C; strain rate, 1×10^{-4}/min; angle of 2CrMo/Inco-weld A interface to tension axis, 45 deg
(b) 10 N/mm² = 1.4503 ksi

the dark etching structure at the weld interface were normalized at 960-1300 C. The plates were then air-cooled to between 450-550 C to produce a bainite or ferrite-bainite matrix in the 2CrMo and finally tempered at 730 C for 1 h. Subsequent metallographic examination revealed only light etching areas indistinguishable from those observed in as-welded samples.

This behavior strongly suggests that the light etching structure was martensitic as proposed by Wood (Ref. 14) and indicates that, if the final joint temperature was maintained above the M_s of the mixed region, complex structures would not be formed at the interface. It also follows that the structures observed at the fusion boundary after heat treatment were tempered martensite and, although extraction replicas of the interface region did not reveal many of the large carbide particles frequently observed in tempered structures, there was evidence of fine precipitation in the matrix of this mixed region (Fig. 5). No diffraction patterns could be obtained from these fine particles however. A slight softening of the interface region was observed on heat treatment, and this is consistent with a tempering reaction.

Those areas of white etching structure which did not appear to be affected by heat treatment (Fig. 4) may be of such a composition that the tempering reaction occurs very slowly. Alternatively, the diffusion of austenite stabilizing elements such as carbon and nickel away from the interface during the stress relief heat treatment may have been sufficient to raise the M_s of some regions above room temperature so that martensitic areas formed on subsequent cooling.

Tensile and Creep Rupture Tests

Room temperature hardness measurements are extremely useful for detecting the multiphase region close to the fusion boundary but they do not give any guide to the behavior of the interface under creep conditions. Barford et al (Ref. 7) used slow strain rate, cross-weld, tensile tests to promote interfacial failure in Rex 500/Inconel 182 joints and, in the present investigation, similar tests were carried out on specimens taken from the butt welds of 2CrMo/Esshete 1250 welded with Inco-weld A so that the weld interface was at 45 deg to the tensile axis. An experimentally convenient test temperature of 580 C was selected. The testing conditions and results are summarized in Table 4.

No significant differences were observed between those results obtained from the as-welded and stress relieved specimens, and in both series failure always occurred in the 2CrMo base metal well away from the weld interface or HAZ.

Metallographic examination of failed specimens showed that, although final failure was always located in the 2CrMo, some damage did occur at the weld interface between 2CrMo/Inco-weld A. Figure 6 shows the profile of a crack which was initiated at the specimen surface and followed the fusion boundary very closely until finally terminating in the 2CrMo HAZ. This behavior was observed in both as-welded and stress relieved specimens and suggests that the interface may initiate cracks which subsequently propagate through the HAZ of the 2CrMo.

Since complete failure at the interface was not obtained in this material combination when tested in tension under these conditions, the effect of the microstructure at the interface on crack initiation and propagation could not be determined readily. However, Mellor (Ref. 17) had shown that interface failure can occur in a stress relieved 2CrMo/Inco-weld A/Esshete 1250 combination when tested under creep conditions. Therefore, stress rupture tests were carried out in an attempt to determine the contribution of the microstructure at the fusion boundary to interface failure. The object of the investigation was to establish whether postweld heat treatment and the consequent change in the initial microstructure of the interface has any observable effect on the mode of failure and time to rupture of the 2CrMo/Esshete 1250 combination.

In the discussion on weld interface structures, it was shown that the structure at the interface changed from a completely light etching structure in the as-welded condition to a predominantly dark etching condition after stress relief and so heat treatment after welding clearly provides a simple means of changing the fusion boundary microstructure. A butt weld was made at a preheat of 50 C and half the joint was stress relieved for 1 h at 730 C. Cross-weld, stress rupture specimens were machined with the interface at 75 deg to the tensile axis from both the as-welded and stress relieved joints and were subsequently tested at 580 C over a range of stresses. The results are summarized in Fig. 7. The results for both welds lie about the same line which suggests that the postweld heat treatment had no significant effect on the rupture properties.

Oxidation of the ferritic parts of the cross-weld specimens made measurement of reduction in area at failure very difficult. However, in every case failure occurred at very low ductility and the weld metal was undeformed. The concentration of deformation in the HAZ of the 2CrMo could be seen clearly in specimens tested at high stresses such as 154 N/mm² (22.3 ksi) as shown in Fig. 8.

Macroscopically all the fractured specimens appeared to have failed at

Fig. 5 — Carbon replica of interface structure in stress relieved joint. X2000, reduced 25%

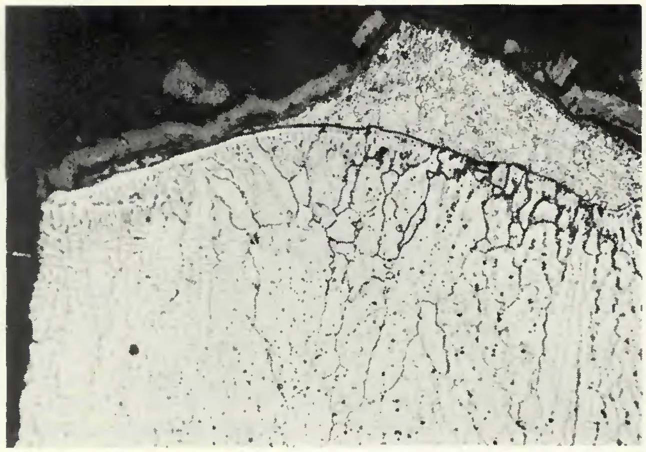

Fig. 6 — Profile of crack at weld interface in tensile test specimen which failed in 2CrMo. X240, reduced 25%

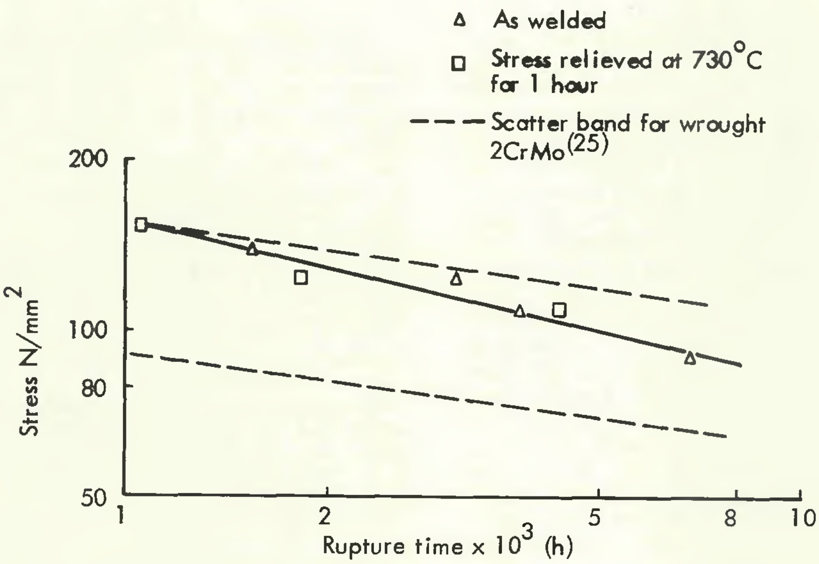

Fig. 7 — Stress rupture results on cross-weld specimens at 2CrMo/Inco-weld A/Esshete 1250, tested at 580 C

or very close to the weld interface irrespective of their heat treatment condition but microscopic examination showed that the cracking, which was intergranular, had propagated almost entirely in the HAZ of the 2CrMo steel. Grain boundary cavitation associated with the fracture was typical of creep failure. In both as-welded and stress relieved specimens the crack profile followed the interface very closely at the specimen surface (Fig. 9), although the interface structure present at the start of the test was different in the two cases. However, examination of the surface of stress rupture specimens just before failure showed no evidence of cracking.

Discussion

The stress rupture and cross-weld tensile tests both indicated that the low ductility interface cracking observed at the ferritic steel/nickel-base weld metal fusion boundary initiated at the interface but propagated in the HAZ of the 2CrMo. The changes in microstructure which resulted from the heat treatment used in this investigation did not significantly effect the nature of or time to failure. In particular, the complex structures observed at the interface did not have any noticeable effect on the initiation or propagation of cracking, contrary to some of the opinions expressed in the literature (Refs. 8,9,18). This insensitivity to microstructure suggests that there is a controlling factor in the initiation of cracking in joints of this type which dominates the behavior so that effects caused by changes in microstructure are not significant.

Although no clear measure of specimen deformation was obtained under stress rupture conditions, in no instance was any reduction in area of the Inco-weld A weld metal observed and, in the shorter term tests, the majority of the deformation of the 2CrMo occurred in the region of the interface (Fig. 8).

Furthermore, slow strain-rate, cross-weld, tensile tests carried out in vacuum by Barford and Probert (Ref. 19) also showed that the majority of the deformation took place in the 2CrMo while the Inco-weld A was hardly affected. The concentration of deformation in the 2CrMo demonstrates the wide disparity in the strengths of the 2CrMo and the weld metal under creep conditions. This mismatch in strength will produce shear stresses along the interface between the weld metal and the weaker 2CrMo which will be maximum at the specimen surface and consequently any cracking is liable to initiate in this region. Microstructural observations support this contention.

Specimens which were examined just before failure did not show any evidence of cracking which suggests that, once a crack is initiated, it propagates rapidly. This, in turn, indicates that the HAZ is a region of low ductility. However, although in this investigation those specimens tested under stress rupture conditions all failed in the region of the interface, Barford & Probert (Ref. 19) and Mellor (Ref. 17) have shown that, at the higher stresses, failure may occur in the 2CrMo base metal well away from the HAZ. This change in the location of failure with stress or time to rupture is not confined to joints welded with nickel-base filler metals.

Similar observations have been made in both tube bursting experiments (Ref. 20) and uniaxial tests (Ref. 21) when 2CrMo joints were welded with matching electrodes. Hopkins et al (Ref. 20) considered that the ductility of the HAZ decreases with increasing time of testing to below that of the pipe material and caused the location of failure to change with time. However, Townsend's work (Ref. 22) suggests that the location of failure is controlled by

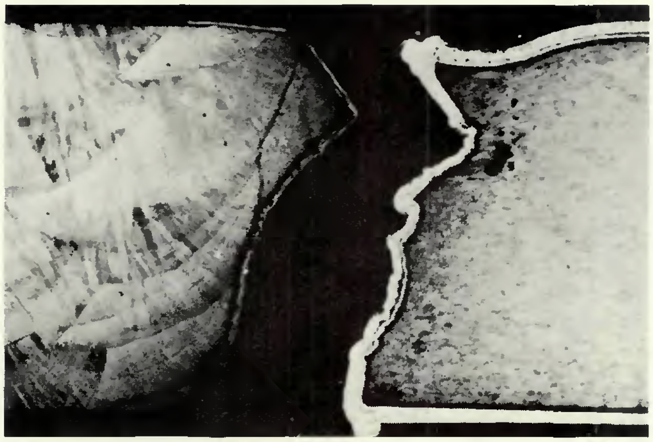

Fig. 8 — Broken stress rupture specimen (580 C, 154 N/mm^2). X6, reduced 25%

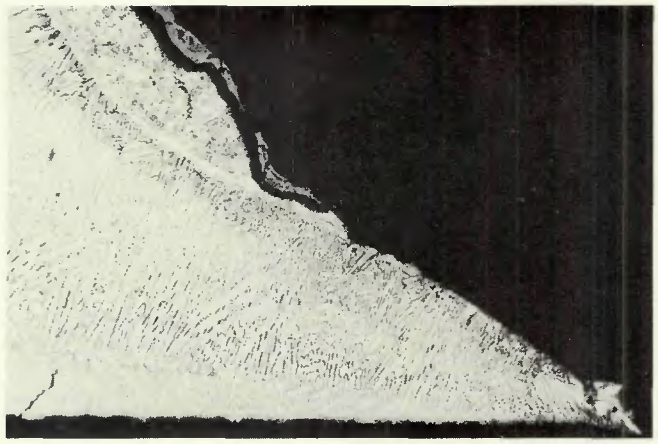

Fig. 9 — Profile of crack at specimen surface. X100, reduced 25%

the relative contributions of grain boundary and matrix deformation which change with applied stress. Under plant operating conditions the stresses applied are usually small and so the uniaxial stress rupture test results suggest that if failure occurs it is likely to be located at the 2CrMo/Inco-weld A interface. Under these circumstances any measure which improves the ductility of the HAZ will also improve the performance of the joint.

At low stresses, variations in creep behavior brought about by microstructural differences are small, due in part to structural deterioration and also to the tendency for grain boundary deformation to control the behavior so that only differences in grain size are important (Ref. 21). Therefore, refinement of the prior austenite grains should produce a reduction in the creep resistance of the HAZ and an increase in the ductility of the region. The effect of refinement of the HAZ, which can be achieved readily by re-normalizing the welded joint, is being investigated at the Central Electricity Research Laboratories but work on crack propagation in 2CrMo suggests that creep strength is the dominant feature in determining propagation rates and grain size has only a secondary effect (Ref. 22). Thus, if a significant improvement in the crack resistance of the HAZ cannot be achieved by heat treatment, the performance of the joint will only be improved if the initiation of the crack can be avoided or delayed.

The initial interface cracking has been attributed to the high shear stresses developed at the weld interface and the most straight forward way to reduce these is to minimize the disparity in creep strengths between the component materials. If it is assumed that the composition of the 2CrMo cannot be altered, the shear stresses at the interface will be most readily reduced by using a filler metal with a strength more closely matching that of the base metal.

Ideally, a matching electrode should be used but although this may be feasible for certain applications, such as the welding of Rex 500 steam piping (Ref. 24), nickel-base electrodes possess many advantages for the welding of dissimilar steel joints (Ref. 1). A possible solution would be the development of a nickel-base filler metal possessing lower creep strength and greater ductility than the existing commercial products, so that when used to weld low alloy, creep resisting, ferritic steels, the shear stresses set up across the weld interface, in service, will be too low to initiate cracking and subsequently cause failure.

Conclusions

In the as-welded condition the lamellar structure observed at the interface between wrought 2CrMo and Inco-weld A weld metal is martensitic. This martensite tempers on stress relief to produce a dark etching structure at the fusion boundary. However, if a high preheat or an appropriate heat treatment is used, the nucleation of the martensite can be delayed until the postweld heat treatment is completed. In this case the dark etching tempered structures are not observed.

Changes in the initial microstructure of the weld interface produced by a postweld heat treatment did not significantly affect the stress rupture properties of cross-weld specimens although failure always occurred close to the weld interface.

In no instance was any deformation of the weld metal detected and the majority of the deformation of the 2CrMo occurred close to the interface. It is suggested that the disparity in strengths between the component materials sets up high shear stresses at the weld interface which initiate cracking at the specimen surface where they are maximum. The crack propagates through the HAZ of the 2CrMo.

A nickel-base filler metal which produces a weld of lower creep resistance and higher ductility than existing products should reduce the shear stresses along the weld interface and lower the incidence of crack initiation at the interface.

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References

1. Payne, B. E. "Nickel-base Welding Consumables for Dissimilar Metal Welding Applications," Proceedings of Conference on Welding Dissimilar Metals, *Metal Construction, 1,* 12s, p.79-87, 1969.

2. Wright, V. S. "Service Experience with Austenitic/Ferritic Superheater Transition Welds within the Central Electricity Generating Board (Midlands Region)," Proceedings of Conference on Welding Dissimilar Metals, *Metal Construction, 1,* 12s, p.1-5, 1969.

3. Wyatt, L. M. "Dissimilar Metal Joints used in the Steam Circuit of Electrical General Plant," Proceedings of Conference on Welding Dissimilar Metals, *Metal Construction, 1,* 12s, p.66-71, 1969.

4. Kent, R. P. "Investigations into Rex 500/Inconel Weld Failures," Proceedings of Conference on Welding Creep-resistant Steels. The Welding Institute, Abington, p.1-17, 1970.

5. Jones, W. K. C. "Metallographic Examination of a Transition Test Piece for Hartlepool Boiler Support Spine," CEGB Research Note RD/M/N582 p.1-5, July 1971.

6. Barford, J. Discussion in Proceedings of Conference on Welding of Creep-resistant Steels. The Welding Institute, Abington, p.191-192, 1970.

7. Barford, J., Willoughby, G., Mellor, H.

G. and Teare, P. W., "Weldment Cracking Involving a Complex Creep-resistant Steel," Proceedings of Conference on Welding Creep-resistant Steels. The Welding Institute, Abington, p. 54-65, 1970.

8. Boniszewski, T. Discussion in Proceedings of Conference on Welding Creep-resistant Steel. The Welding Institute, Abington, p. 185-186, 1970.

9. Eaton, N. F. and Glossop, B. A. "The Welding of Dissimilar Creep-resisting Ferritic Steels," Proceedings of Conference on Welding Dissimilar Metals, *Metal Construction, 1,* 12s, p. 6-10, 1969.

10. Emerson, R. W. and Hutchinson, W. R. "Welded Joints between Dissimilar Metals in High Temperature Service," *Welding Journal, 31,* (3), Research Suppl., p. 127-s to 141-s, 1952.

11. Gotalskii, Yu. N. "The Problem of Welding Dissimilar Steels in Structures Intended for Use for Long Periods at High Temperatures," *Automatic Welding, 17,* (12). p.34-40, 1964.

12. Mullery, F. Discussion in Proceedings of Conference on Welding Dissimilar Metals, *Metal Construction, 1,* (12s), p.143, 1969.

13. Barford, J. Discussion in Proceedings of Conference on Welding Dissimilar Metals, *Metal Construction, 1,* (12s), p.136-137, 1969.

14. Wood, D. Discussion in Proceedings of Conference on Welding Dissimilar Metals, *Metal Construction, 1,* (12s), p.134-136, 1969.

15. Clarke, S. F. Unpublished Work, 1969.

16. Slater, D. and Winn, J. M. "The Fusion Boundary between Mild Steel and High Alloy Steel Weld Metal," Proceedings of Second Commonwealth Welding Conference, Institute of Welding, London, Paper M12, p.180-198, 1965.

17. Mellor, H. G. "Mechanical Properties of 2¼Cr1Mo-Inco-Weld A — Esshette 1250 Welds," CEGB Research Report RD/L/N153/72 p.1-14, October, 1972.

18. Glossop, B. A. Hall, D. and Irvine, J. "The Welding of High Pressure Pipework for the CEGB," Proceedings of Conference on Pipe Welding, Welding Institute, Abington, p.125-130, 1969.

19. Barford, J. and Probert, K. S. "Interfacial Effects in Dissimilar Steel Joints," Proceedings of Conference on Welding Research Related to Power Plant, CEGB, London, Paper 4, p. 1-12, Sept. 1972.

20. Hopkins, L. M. T., Murray, D. and Deval, D. "Creep Properties of Some Butt Welds in Steampipes," *Jnl. Iron and Steel Institute 203* p.819-826, 1965.

21. Lister, E., Mickleraith, J. and Higginbottom, A. "High Temperature Properties of Steampipe Welds," Proceedings of Conference on High Strength Steels, ISI London, p.473-481, 1966.

22. Townsend, R. "The Effect of Structure on the Creep Properties of Low Alloy Ferritic Steels," CEGB Research Report RD/L/R1740, 1971.

23. Neate, G. J. Private Communication, 1972.

24. Boniszewski, T., Kent, R. P. and Kirkby, H. W. "Feasibility of Welding Rex 500 Pipe with Matching Electrodes," Proceedings of Conference on Welding of Creep-resistant Steels, Welding Institute, Abington, p.32-53, 1970.

25. ISO/TC17/WG10/ETP-SG. Secretariat 46-53.

WRC Bulletin
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"Submerged-Arc-Weld Hardness and Cracking in Wet Sulfide Service"

by D. J. Kotecki and D. G. Howden

This study was undertaken to determine:

(1) The causes of higher-than-normal hardness in submerged-arc welds in plain-carbon steels

(2) The levels of strength or hardness which will not be susceptible to sulfide-corrosion cracking

(3) Welding procedures which will assure that nonsusceptible welds will be produced.

Concentration is primarily on weld metal, though some consideration to the weld heat-affected zone is given. The study covered a two-year period. The first year was concerned with a macroscopic view of the weldments. In that first-year study, some inhomogeneities were observed in weldments which are not obvious in a macroscopic view of the weldment. It appeared likely that these inhomogeneities could affect the behavior of the weldment in aqueous hydrogen-sulfide service. Accordingly, their presence and effects were investigated during the second year.

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WRC Bulletin
No. 185
July 1973

"Improved Discontinuity Detection Using Computer-Aided Ultrasonic Pulse-Echo Techniques"

by J. R. Frederick and J. A. Seydel

The purpose of this project, sponsored by the Pressure Vessel Research Committee of the Welding Research Council, was to investigate means for obtaining improved characterization of the size, shape and location of subsurface discontinuities in metals. This objective was met by applying computerized data-processing techniques to the signal obtained in conventional ultrasonic pulse-echo systems. The principal benefits were improved signal-to-noise ratio and resolution.

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