FRACTURE BEHAVIOUR OF LOW ALLOY STEEL WELDMENTS

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ABSTRACT

Type IV cracking is an extensive in-service problem with low alloy steel pipe weldments operating under creep conditions. This paper describes part of an ongoing research programme studying the factors affecting this form of damage in which cross-weld, uniaxial creep tests were conducted on a weldment removed from service. These tests, conducted over a range of stress and temperature combinations, invariably resulted in failures within the heat affected zone. However, damage was observed in the coarse grained region near the fusion boundary, in the intercritical region adjacent to the base metal, or in some cases in both areas. The variation in fracture location is rationalised in terms of local differences in weldment structure.

INTRODUCTION

Components in petrochemical plant and power stations operate under conditions of high temperature and pressure such that failures are frequently usually associated with the initiation and propagation of creep damage. In low alloy steel piping systems service experience suggests that damage usually occurs at weldments[1], as a consequence of the specific operating conditions and the heterogenous microstructures produced by the welding process. A particular form of creep cracking, designated Type IV[2], has been identified within the intercritical region of weldment heat affected zones (HAZ). Initial experience suggested that damage of this form occurred in ½Cr½Mo¼V steels between about 60,000 and 80,000 hours of operation[3]. However, it now appears that a wide range of low alloy steels are susceptible and that details of the factors affecting damage development have not been established.

To understand the processes that control the formation of type IV damage, it is necessary to reproduce typical service failures in relatively small scale, accelerated laboratory tests. Indeed, a number of programmes considering the long term creep behaviour of cross-weld specimens tested in uni-axial tension have been reported[4]. However, even with long term testing on joints of apparently similar steels details of the failure modes vary considerably.

The present paper describes part of an on-going research programme studying the high temperature behaviour of low alloy steel weldments. The microstructure of a typical service weld is characterised. Cross weld testing of large geometry specimens is conducted and the changes that occur during creep are studied. Details of the high temperature fracture properties of these specimens are presented and the implications to in-service behaviour discussed.

MATERIAL AND EXPERIMENTAL METHODS

A series of butt welds, manufactured in ½Cr½Mo¼V pipe using 2¼Cr1Mo weld metal, were removed from the steam piping system of a petrochemical plant. These weldments had been in service for approximately 130,000 hours at a temperature of about 550°C. The piping sections had an outside diameter of 356mm and a wall thickness of 38mm. The weldments had been manufactured to a standard procedure, using a 'J' preparation, with a tungsten inert gas root pass, and manual metal arc filler and capping passes. Full thickness sections were removed from each weldment, revealing a cross-section of the weld preparation. The sections were metallographically prepared and studied under the optical microscope, to characterise the microstructures present.

The composition of the weld metal and both sections of base metal are given in Table 1. Cross weld samples were prepared with a gauge length of 100mm and a 14mm square gauge-section, with the weld situated at the centre of the specimen, Figure 1. Constant load testing was performed over a range of stresses between 35MPa and 82MPa at temperatures between 630°C and 700°C.

TABLE 1
Composition of Ex-service Weldment as Tested

MATERIAL	C	Si	S	P	Mn	Ni	Cr	Mo	V
Base Metal A	0.12	0.25	0.01	0.01	0.69	0.13	0.34	0.62	0.22
Weld Metal	0.05	0.39	0.01	0.02	0.79	0.05	1.97	1.11	0.03
Base Metal B	0.12	0.14	0.01	0.01	0.58	0.14	0.35	0.62	0.28

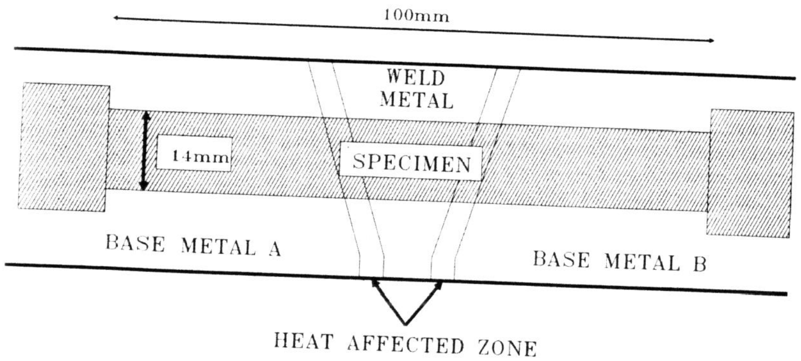

FIGURE 1:
Schematic showing the position of cross-weld specimens removed from the ex-service weldment.

For each test the displacement time behaviour was monitored using specialist, high temperature extensometry, and capacitance transducers, connected to a computer controlled logging system. The temperature during testing was held constant across the entire gauge length. After testing all specimens were mounted and metallographically prepared. These samples were then examined both macroscopically and microscopically to document details of the fracture locations and mechanisms.

RESULTS

Characterisation of Weldment Microstructures

A cross-section of the ex-service weldment selected for high temperature testing is shown in Figure 2. The weldment has been sectioned and metallographically prepared to reveal the base metal, HAZ and weld metal structures present.

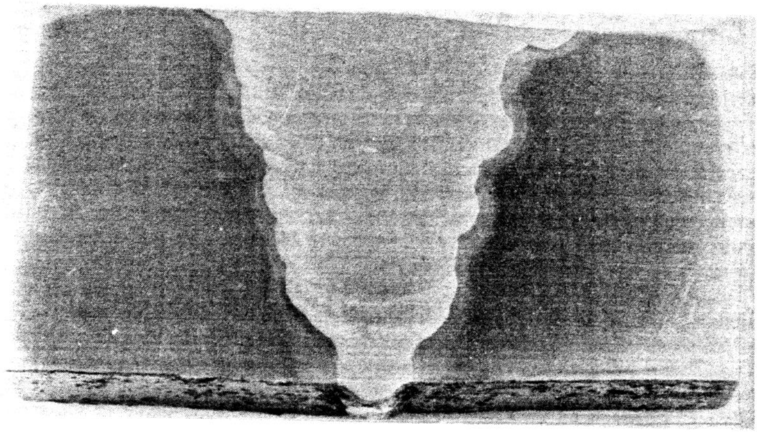

FIGURE 2:
Micrograph showing the macrostructure of the ex-service weldment. (x2)

Both base metal sections contained predominantly ferrite with a grain size of approximately 30µm with about 10% transformation product. Average macro-hardness values of 158DPH and 162DPH were noted for sections A and B respectively. The weld metal had been laid down using a weaved welding technique. This resulted in significant refining of the microstructure within the bulk of the weld. However, a columnar grain structure was observed in the cap pass. In all weld metal locations transformation to bainite has occurred. The weaving process allows more variation in heat input than welding using stringered passes. This variation is particularly pronounced in the HAZ adjacent to the fusion boundary, so that the microstructures, and the distribution of microstructures, in this region are complex. Moreover, the size of the HAZ in general and the distribution of HAZ microstructures in particular was found to be very variable. In some locations it can be seen that significant heat input resulted in grain growth. Thus, in these regions a relatively coarse prior austenite grain structure was produced, with a prior austenite grain size of approximately 60µm. In other regions where significant overlap of thermal cycles had occurred a refined grain structure was developed, with a grain size of approximately 10µm. Both regions had undergone complete transformation to bainite. In the intercritical region of the HAZ a mixed grain structure of ferrite and bainite, with a grain size of approximately 6.5µm was present. Despite the large variations in structures through the wall thickness there was no significant variation in structures between sides A and B of the weldment.

It was also noted that no indications of creep damage were found in any regions of the weldment when studied under the optical microscope.

Creep Testing

The 14mm square gauge section of the testpieces was selected to provide a specimen with microstructural features representative of those found in the bulk of the ex-service weldment. Clearly then, since the weldment contained a predominantly refined structure samples were produced from these regions of the weld, ie both the root area and the capping pass were machined off during testpiece manufacture.

As the test temperature decreased for a given stress, or the test stress decreased for a given temperature, time to rupture for the specimen was found to increase. This increase varied linearly with the log of rupture time, Figure 3a & 3b. Failures were produced over a range of times between 144 hours and 1725 hours. On loading at temperature there was an initial elastic strain in the specimen, before displacement rate decreased during primary creep. After approximately 20 hours there was a short period of apparently constant displacement rate, but for the majority of each test an accelerating displacement rate was observed. All tests showed a similar creep curve shape, so that primary creep occupied only about 12% of the total specimen life, whilst tertiary behaviour accounting for over 80%, of the test.

Examination of the failed test pieces revealed that damage was restricted predominantly to the failed HAZ in all tests conducted. Failures did occur preferentially on one particular side of the weldment. Although all failures were of low ductility, two distinct types were observed. Firstly, a typical type IV failure, ie. cracking through the intercritical region of the HAZ, Figure 4. Damage was found to occur throughout the intercritical region of the HAZ. Secondly, a mixed failure mode where cracking occurred in both the intercritical region and the coarse grained region of the HAZ, Figure 5. Damage in the coarse grained bainitic region of the HAZ appeared to be associated primarily with regions where weld beads intersected. Total elongation at failure and the failure type and location are given in Table 2.

TABLE 2
Characterisation of Failed Test Pieces

STRESS (MPa)	TEMP (°C)	FAILURE TYPE AND LOCATION	ELONGATION (%)
55	650	MIXED HAZ	-
82	650	MIXED HAZ	7.78
55	630	MIXED HAZ	3.32
82	630	MIXED HAZ	3.95
55	680	TYPE IV IN HAZ	5.27
35	650	TYPE IV IN HAZ	4.78
35	680	MIXED HAZ	4.38

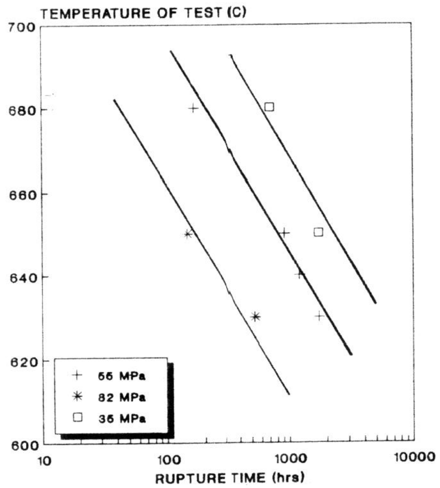

FIGURE 3a:
Time to rupture vs. temperature for tests conducted on ½Cr½Mo¼V/2¼Cr1Mo weldments.

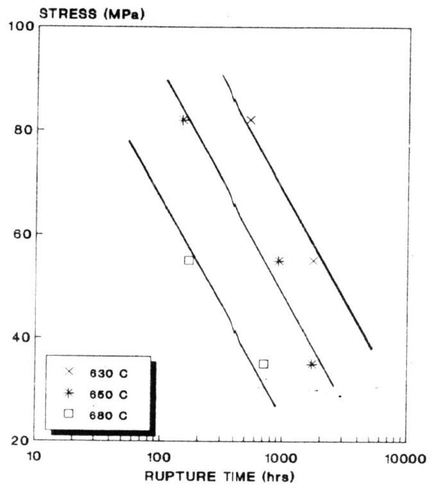

FIGURE 3b:
Time to rupture vs. stress for tests conducted on ½Cr½Mo¼V/2¼Cr1Mo weldments.

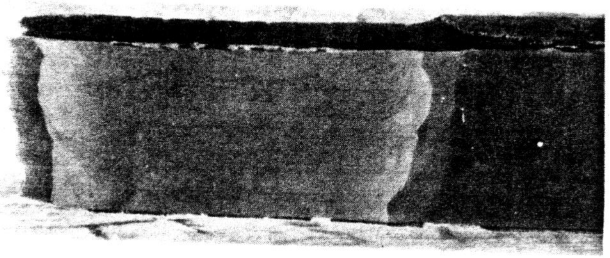

FIGURE 4:
Micrograph of the failed specimen from the test at 55 MPa & 680°C. (x3.6)

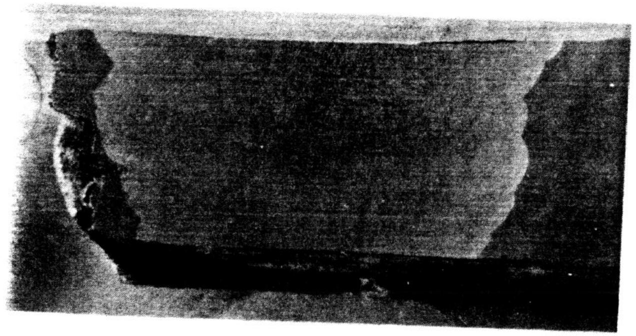

FIGURE 5:
Micrograph of the failed specimen from the test at 82 MPa & 650°C. (x3.6)

DISCUSSION

Weaved pass welding produces a much more uneven heat input than stringered pass. This means that the microstructures produced by the welding process are much more varied. Work by Coleman[5] has discussed microstructural development for idealised stringered pass weldments. However with weaved pass there is little control of the electrode during welding, especially at the edge of the preparation where the welder is effectively reversing the direction of travel of the electrode. Any variation in 'turning speed' produces large variations in heat input, and thus weldment microstructure. It can be seen that although, superficially the weld gives the appearance of being made up of, albeit irregular, beads and cusps, the microstructures in these regions are not necessarily those predicted for the stringered pass technique.

In a similar way to stringered pass welding, regions within both the weld metal and the heat affected zone can be made up of either coarse or fine grains. The coarse grained regions arise in areas which have seen the heat input from only one weld pass. Whilst, in the fine grained regions subsequent weld passes have refined the previous coarse structures. In the HAZ, unlike stringered pass welding, these regions are not associated specifically with beads or cusps.

The creep rupture lives of each test were found to vary linearly with the log of both stress and temperature, Figure 3. The rupture times and the rupture time / temperature relationship found for the tests conducted at 82 Mpa, were found to be in excellent agreement with published uniaxial cross-weld test data[6].

Failure, under all conditions tested, occurred in the HAZ of the failed testpiece, by low ductility intergranular cracking. No damage was noted in the weld metal at the surface of the specimen, but some damage was found in a specimen sectioned to reveal the centre of the testpiece. It was also noted that failures were associated with one particular side of the weldment. Thus, although one HAZ has accumulated damage to the point of failure, there is no sign of damage in the intact HAZ, which has been subjected to the same conditions. This implies that there was damage associated with this HAZ from previous service. However, examination of the ex-service weldment showed no indications of this. Therefore, once damage has been nucleated in the more susceptible HAZ, its multiplication and propagation is so rapid that failure occurs before damage has initiated in the stronger HAZ. A specimen was sectioned to find if damage had accumulated in the centre of the specimen, at any locations other than the failed HAZ. Some intergranular cracking had occurred in the weld metal, but there was no damage associated with the intact HAZ. Therefore, after an initial period, damage accumulation and fracture occur quickly in the susceptible intercritical region. Work is now in progress to define the features that make one interface more susceptible to cracking.

Failure types fell into two distinct categories, firstly a typical Type IV failure, and secondly a mixed HAZ failure. The Type IV failures occurred in specimens with rupture times of 165 hours and 1707 hours, showing that these failure types can be reproduced in short term tests. Damage was concentrated in the fine grained intercritical region of the weldment. Damage occurs throughout the region before crack propagation leads to fracture. In the mixed failure types, again there was extensive Type IV damage and crack propagation through the intercritical region. However, some cracking was also present in the coarse grained bainitic region adjacent to the fusion boundary. This only occurred where a coarse grained bainitic structure was associated with a cusp like feature on the fusion boundary. In such regions stress redistribution occurs with the relatively ductile weld metal, off-loading stress onto the brittle coarse grained bainite, resulting in intergranular cracking. The crack grows across a cusp like region where it is arrested by the weld metal at each side. Thus in fine grained cusp regions no cracking occurs, as the material is more ductile, whilst in coarse grained bainite regions associated with bead like features no stress redistribution occurs to promote cracking. Mixed failures, therefore, occur when simultaneous damage growth in the intercritical region and the coarse bainitic HAZ links to form the mixed failure type observed.

CONCLUDING REMARKS

It is possible to reproduce type IV failures typical of those in service, using cross-weld uniaxial test pieces. In weldments manufactured using a weaved pass technique constrained regions of brittle microstructure are produced adjacent to the fusion boundary that can promote cracking. However these constrained regions do not occur in stringered pass welding. One side of the weldment tested is particularly susceptible to cracking. Work is continuing to define the features that increase the susceptibility of this particular interface.

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