THE MECHANISM OF STRESS RELIEF CRACKING IN Cr-Mo-V STEELS

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INTRODUCTION

Stress relief cracking (SRC) in the weld heat affected zones (HAZ's) of low alloy creep resisting steels is commonly understood as a problem of low creep ductility involving unstable intergranular crack growth at low strains [1]. Although the previous work on SRC has failed to elucidate precisely the conditions under which crack growth occurs, it is usually assumed to be promoted by the creep strengthening of the grains as a result of the formation of a fine dispersion of alloy carbides during the stress relief heat treatment [2]. The work described here was performed on two commercial samples of 1/2 Cr, 1/2 Mo, 1/4 V steel. The details of the composition of these samples are described elsewhere [3]. One of the samples labelled A, had been taken from steam pipe which was welded, stress relieved and put into service without any sign of premature failure. The other, labelled B, was taken from steam pipe which had been found to suffer from SRC during fabrication.

EXPERIMENTAL

Smooth tensile specimens having an overall length of 120 mm, a gauge diameter of 12.7 mm and a parallel gauge length of 55 mm, and Hounsfield No. 13 double shouldered tensile specimens were machined from the samples of steam pipe. All specimens were solution treated in a vacuum furnace for one hour, at temperatures of 1273K and 1473K, after which they were water quenched. The prior austenite grain sizes after these heat treatments were estimated to be 25 μm and 130 μm respectively for sample A, and 110 μm and 225 μm respectively for sample B. The Hounsfield No. 13 double shouldered tensile specimens were tempered for one hour at temperatures in the range 393K to 973K, and then broken in tension at 77K. The smooth tensile specimens were subjected to anisothermal stress relaxation tests using a MAND electro-servo-hydraulic testing machine in closed loop strain control mode. The details of furnace control, extensometry and data collection used in this system are described elsewhere [5]. A block diagram of the system is shown in Figure 1. Prior to heating, the specimens were loaded at a nominal strain rate of 2 x 10^{-4} s^{-1}. Initial stresses in the range 200 to 850 MPa were employed. During testing, the specimens were held at a constant total imposed strain while being heated from ambient to 973K at a rate of 50K h^{-1}.

When a tensile specimen is loaded under conditions of constant total imposed strain it flows plastically, under the action of the unrelieved elastic stress, at a rate which may be estimated [4] according to:

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\[ \dot{\sigma} = \frac{\dot{\varepsilon}}{N} \]  

where \( \dot{\sigma} \) is the observed rate of stress relaxation and \( N \) is the appropriate elastic coefficient for the specimen and the machine. For the presently used loading conditions, it was found [5] that \( N \) could be closely approximated by \( E \) (the value of Young's Modulus for 1/2% Cr, 1/2% Mo, 1/4% V) over the range of temperatures investigated [5]. The method used to calculate strain rate during anisothermal stress relaxation is summarized in Figure 2. The results are presented in the form of graphs of stress versus temperature and strain-rate versus temperature.

RESULTS

Figures 3 and 4 illustrate the stress-temperature and corresponding strain-rate-temperature response for sample A during anisothermal stress relaxation after solution treatment at 1273K for one hour. The numbers in parentheses identify the specimen used in a particular test. It is clear from the form of the stress-temperature curves that:

(a) Stress relaxation occurred smoothly; and
(b) a significant reduction of stress occurs during the heating cycle of a stress relief heat treatment.

From Figure 4, it would appear that the strain-rate during anisothermal stress relaxation of 1/2% Cr, 1/2% Mo, 1/4% V steel passes through two minima. Metallographic examination of these latter specimens revealed that cavitation of the form shown in Figure 5 had occurred during testing.

The stress- and strain-rate-temperature responses for sample A, after solution treatment at 1473K, and sample B, after solution treatment at 1273K, were observed to be essentially similar to that shown in Figures 3 and 4. Metallographic examination of these latter specimens revealed that cavitation of the form shown in Figure 5 had occurred during testing.

The stress-temperature response for Sample B after solution treatment at 1473K is shown in Figure 6. It is clear that stress relaxation of sample B, in this condition of prior solution treatment did not always occur smoothly. In particular, for specimens no. 4 and no. 15, abrupt reduction of stress was observed to occur at about 623K. Metallographic examination revealed the presence of gross intergranular cracking in both of these specimens, as shown in Figure 7.

Observations of the failure mode of tensile specimens broken at 77K revealed:

(a) that those specimens which had been solution treated at 1273K failed in a mixed ductile/cleavage manner;
(b) that those specimens which had been solution treated at 1473K failed predominantly by cleavage;
(c) that specimens of this latter group which had been tempered in the range 545K to 693K showed evidence of intergranular failure, as shown in Figure 8.

DISCUSSION

The extremely brittle nature of the cracking, illustrated in Figure 7, suggests that the cracks, once nucleated, have grown rapidly under the action of the unrelied stress. The model of Smith and Barnby [6] for the nucleation of grain boundary cavities may be described by an equation of the form:

\[ \frac{\partial^2}{\partial t^2} = \frac{2}{3} \frac{\eta}{E} \frac{d^2}{(1-v^2)} \]

where \( d \) = barrier thickness; \( 2f \) = distance along boundary between barriers.

McLean [7] showed that a cavity nucleus would grow once it achieved the size:

\[ r_c = \frac{2Y_d}{\eta} \]

If it is assumed that:

(a) the barrier thickness equals the critical nucleus size (i.e., \( r_c = h \));
(b) the distance between the barriers, \( 2f = 0.34 \mu m \) [5];
(c) \( v = 0.3 \);
(d) \( E_{623K} = 18 \times 10^5 \text{ MPa} \) [5];

it is then possible to solve equations (2) and (3) simultaneously to estimate values of \( r_c \) and \( Y_c \). For the measured stresses of 410 MPa and 350 M observed at the beginning of the stress drops in specimens no. 15 and no. the results are:

(i) for specimen no. 15

\[ r_c = 87 \times 10^{-19} \text{ m} \]

\[ Y_c = 1.8 \text{ J m}^{-2} \]

(ii) for specimen no. 4

\[ r_c = 51 \times 10^{-19} \text{ m} \]

\[ Y_c = 1.4 \text{ J m}^{-2} \]

These estimations of critical nucleus size and surface energy are consistent with boundary particle size observations from fractography of crack Cr-Mo-V welds [11], and surface energy values in the Fe-P system [9].

It is significant that the cracking described occurred at about 623K for three major reasons:

(i) This temperature is close to that of the observed lower temperature strain-rate minima.
(ii) Tensile specimens tempered at this temperature and then broken at 77K showed evidence of intergranular failure.
(iii) Tensile specimens deformed at this temperature (after equivalent solution treatment) do not fail in a brittle intergranular manner [5].

For the first, after comparison of the apparent activation energies for the strain-rate minima with published data [8], it would appear reasonable to attribute the lower temperature minima to a carbon-dislocation interaction, and the upper temperature minima to a vanadium-carbon-dislocation interaction. The cracking is therefore not associated with a creep strengthening effect due to the formation of a fine dispersion of alloy carbides. For the second, it is reasonable to attribute the boundary
weakening observed after tempering at 623K to an impurity segregation effect, most probably involving phosphorus [9]. It is tempting to relate this effect to the observed ease of nucleation of cavities during stress relaxation at 623K. For the last, it is necessary to understand that the conditions of loading in the anisothermal stress relaxation test allow crack propagation to occur in association with stress reduction, but that this is not so for the rising load tensile test. Furthermore it is observed [10] that both the strain-rate sensitivity, and the work hardening exponent of 1/3 %, 1/2 %, 1/4 % are low for deformation at 623K. It seems likely, therefore, that even when cavity nucleation occurs in a rising load tensile test at 623K, subsequent brittle growth is inhibited because the growing cavity rapidly becomes blunted as a result of strain localization. For the anisothermal stress relaxation test, the conditions of loading allow such crack tip blunting to be inhibited until significant crack growth has occurred. Essentially these are the conditions which exist in a weld HAZ prior to stress relief.

It is further noted that the prior austenite grain size of sample B is about twice that of sample A after solution treatment at 1473K. This observation is significant for two reasons:

(i) For any impurity boundary segregation phenomena it is recognized [9] that a larger grain size will allow more intense boundary coverage for a given bulk concentration of impurity, because of a reduced grain boundary area.

(ii) Triple junctions are recognized [10] as providing barriers to cavity growth such that a fine grain size will promote cavity blunting and inhibit crack propagation.

It appears, therefore, that while both samples showed evidence of boundary weakness during failure at 77K, the conditions of microstructure and loading during stress relaxation at 623K were sufficient to cause nucleation controlled intergranular failure of sample B, but not of sample A, consistent with the observed susceptibilities to SRC.

CONCLUSIONS

(i) SRC occurs in the first instance under loading conditions which allow rapid propagation of nucleation controlled intergranular failure.

(ii) In Cr-Mo-V steels SRC is not associated with the creep strengthening of the grains as a result of the formation of a fine dispersion of alloy carbides.

(iii) The susceptibility of commercial Cr-Mo-V steels to SRC depends significantly upon the extent of austenite grain coarsening within a weld HAZ, and to some degree upon the segregation of impurities to the prior austenite grain boundaries.

ACKNOWLEDGEMENTS

The authors wish to thank Drs. G. T. Jones and A. T. Price for the supply of material, and Professor R. W. K. Honeycombe for the provision of laboratory facilities. One of the authors (RAT) was supported by the Science Research Council during the course of this work.

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Figure 1 Block Diagram of System Used for Anisothermal Stress Relaxation Tests
Figure 2 Method Used to Estimate Strain Rate During Anisothermal Stress Relaxation

Figure 3 Stress-Temperature Response of Sample A During Anisothermal Stress Relaxation After Solution Treatment at 1273K

Figure 4 Strain-Rate-Temperature Response of Sample A During Anisothermal Stress Relaxation after Solution Treatment at 1273K

Figure 5 Intergranular Cavitation of Sample A Formed During Anisothermal Stress Relaxation after Solution Treatment at 1473K

Figure 6 Stress-Temperature Response of Sample B During Anisothermal Relaxation, after Solution Treatment at 1473K

Figure 7 Intergranular Failure of Sample B During Anisothermal Stress Relaxation continue
Figure 7 Intergranular Failure of Sample B During Anisothermal Stress Relaxation

Figure 8 Intergranular Failure in Sample B, Broken at 77K after Solution Treatment at 1473K and Tempering at 623K