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 sam-
ples 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 dia-
meter 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 treat-
ments 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 [3]. A block dia-
gram of the system is shown in Figure 1. Prior to heating, the specimens
were loaded at a nominal strain rate of $2 \times 10^{-3} s^{-1}$. 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 im-
posed 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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$$\dot{\epsilon}_p = -\frac{\dot{\sigma}}{N} \tag{1}$$

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 [3] 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 the specimens in this series of tests failed to reveal any evidence of intergranular cavitation.

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 ostensibly 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 543K 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 unrelieved stress. The model of Smith and Barnby [6] fo the nucleation of grain boundary cavities may be described by an equat of the form:

$$\sigma^2 = \frac{2\gamma_p E\pi h}{f^2(1-\nu^2)}$$

where 2h = barrier thickness; 2f = distance along boundary between barriers.

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

$$r_c = \frac{2\gamma_s}{\sigma}.$$

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 μm [3];
(c) $\nu = 0.3$;
(d) $E_{623K} = 18 \times 10^4$ MPa [5];

it is then possible to solve equations (2) and (3) simultaneously to est mate values of r_c and γ. 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

$r_c = 87 \times 10^{-10}$ m

$\gamma_s = 1.8$ J m^{-2}

(ii) for specimen no. 4

$r_c = 81 \times 10^{-10}$ m

$\gamma_s = 1.4$ J m^{-2}

These estimations of critical nucleus size and surface energy are consistent with boundary particle size observations from fractography of cracke 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 [3].

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 [3] that both the strain-rate sensitivity, and the work hardening exponent of 1/2% Cr, 1/2% Mo, 1/4% V 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

REFERENCES

1. JOY, G. D. and SAGE, A. M., "Stress Relief Heat Treatment of Vanadium Steels", Highveld Publication, 1972.
2. NICHOLS, R. W., Welding in the World, 7, 1969, 244.
3. TAIT, R. A., Ph. D. Thesis, Cambridge, 1976.
4. HART, E. W., Acta Met., 18, 1970, 599.
5. DATE, E. H. F., J.I.S.I., 207, 1969.
6. SMITH, E. and BARNBY, J. T., Met. Sci. J., 1, 1967, 1.
7. McLEAN, D., J. Inst. Met., 85, 1956, 468.
8. GEMMILL, M. G., "The Merits and Demerits of Strengthening Mechanisms in Low-Alloy Steels", Climax Moly. Publ., Conference on Steel Strengthening Mechanisms, Zurich, 1969.
9. CLAYTON, J. Q., Ph. D. Thesis, Cambridge University, 1976.
10. FLECK, R. G., BEEVERS, C. J. and TAPLIN, D. M. R., Met. Sci. J., 9, 1975, 49.
11. BONISZEWSKI, T. and EATON, N. F., Met. Sci. J. 3, 1969, 103.
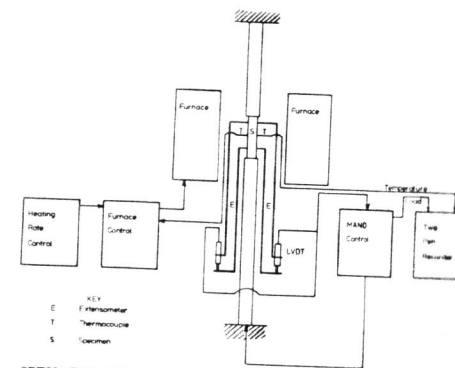
SPECIMENS LOADED AT ROOM TEMPERATURE, UNDER CLOSED LOOP STRAIN CONTROL, SIMULATING A RESIDUAL STRESS SITUATION AS THE TEMPERATURE IS INCREASED, STRESS RELAXATION IS MONITORED.

Figure 1   Block Diagram of System Used for Anisothermal Stress Relaxation Tests
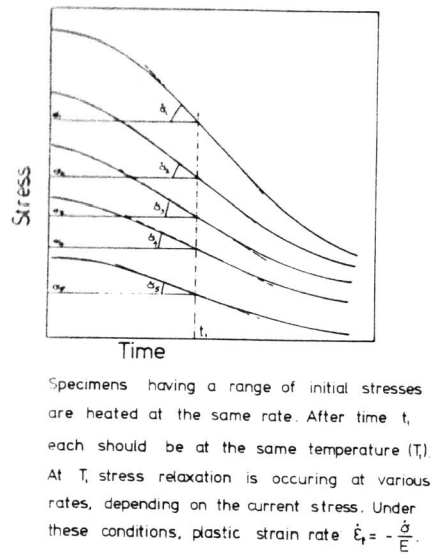
Specimens having a range of initial stresses are heated at the same rate. After time t, each should be at the same temperature (T_1). At T_1 stress relaxation is occuring at various rates, depending on the current stress. Under these conditions, plastic strain rate $\dot{\epsilon}_t = -\frac{\dot{\sigma}}{E}$.

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

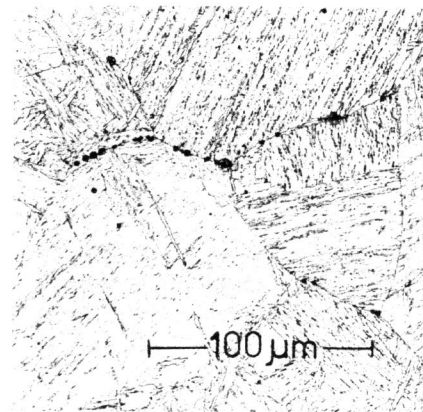Figure 6   Stress-Temperature Response of Sample B During Anisothermal Relaxation, after Solution Treatment at 1473K

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
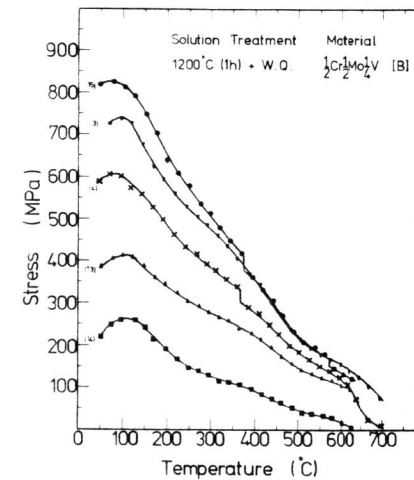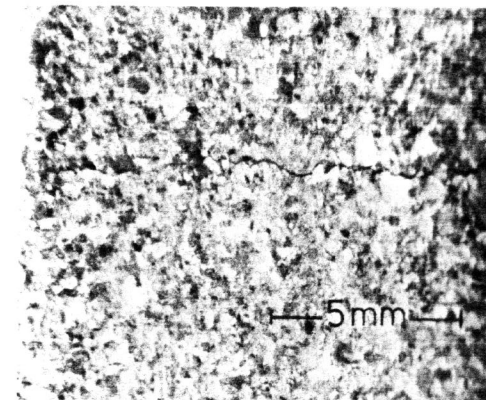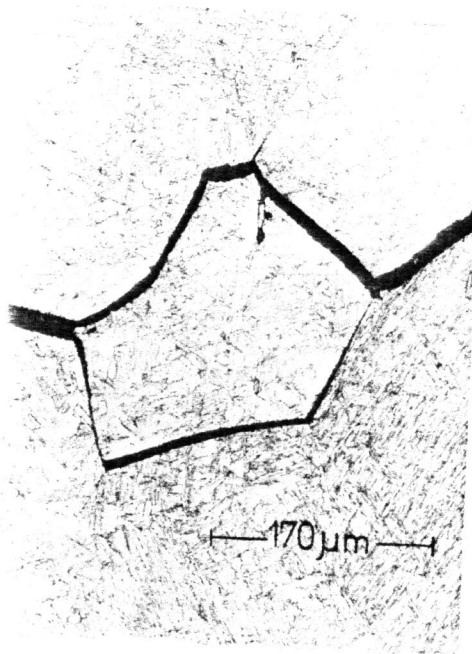

(a)

Figure 7   Intergranular Failure of Sample B During Anisothermal Stress Relaxation

continued

(b)

Figure 7   Intergranular Failure of Sample B During Anisothermal Stress
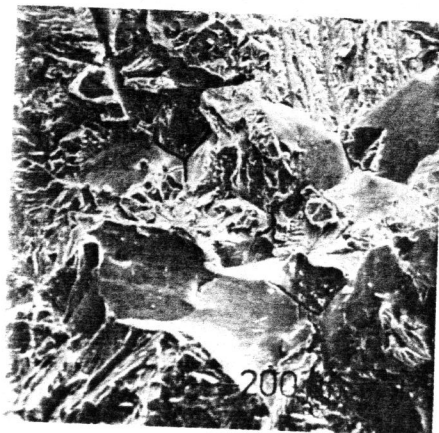Relaxation

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