Fracture 1977, Volume 2, ICF4, Waterloo, Canada, June 19 - 24, 1977

THE MECHANISM OF A FATIGUE/CREEP INTERACTION IN A LOW ALLOY STEEL

R. A. Bartlett, W. J. Plumbridge,
T. E. Chung and E. G. Ellison*

INTRODUCTION

Low alloy steels of the Cr Mo V type are utilized in steam power generating plant under conditions in which fatigue/creep interactions may occur. Such interactions may be damaging and reduce component life to significantly less than that calculated from single mechanical property data. Although substantial effort has been devoted to the correlation between microstructure and mechanical properties in isolation [1 - 4], little attention has been paid to this more complex type of stressing.

The present paper describes a metallographic investigation of the nature of a sequential interaction between high strain fatigue and subsequent creep at low stress levels.

EXPERIMENTAL DETAILS

The material investigated was a 1 Cr Mo V steel supplied by the C.E.G.B. from a single cast as part of their collaborative research programme on this alloy [5]. The composition and heat treatment are given elsewhere [6], and shown in Table 1. Static load creep tests were conducted to various fractions of the rupture life, recording strain at all times. Cyclic strains were applied in a servohydraulic machine operating under strain control. Conventional techniques were used for the preparation of specimens for optical microscopy and carbon extraction replicas. Thin foils for transmission electron microscopy were prepared from 3 mm disc specimens by jet electropolishing in an electrolyte of 7% perchloric-acetic acid mixture at an applied voltage of 60 V at 15° C.

RESULTS

(1) *Creep* - Creep tests were carried out at 565°C at an initial stress of 220 MN/m^2, in order to produce an intergranular "brittle" type of failure. One specimen was allowed to creep to failure at this stress. Six other specimens were then loaded similarly, but the tests were stopped at different fractions of the creep life, (Figure 1).

(2) *Response to Cyclic Strain* - Specimens were cycled at both ambient temperature and 565°C at a total strain range of ± 0.5%, either to failure, or to estimated fractions of that life, (Figure 2). The material cyclically softened at both temperatures. Initial softening occurred rapidly, whereas in the life range 10-80%, conditions appeared more stable until cracks became sufficiently large to cause failure. The fatigue life at 565°C was shorter than that at room temperature, though the softening characteristics were very similar, (Figure 2).

* Department of Mechanical Engineering, University of Bristol, England

MS274

(3) *Prior Fatigue on Subsequent Creep* - A specimen was first subjected to a nominal 50% of the total number of cycles to failure at ± 0.5% total strain at 565°C. After cooling, it was transferred to a static creep machine and reloaded again at 565°C. The subsequent creep rupture curve for a stress of 230 MN/m^2 is shown in Figure 1. Final creep failure was by a brittle intergranular mechanism, and the specimen exhibited a faster creep rate and shorter life than that observed for the uncycled material.

(4) *Metallography* - The initial microstructure is shown in Figure 3. It reveals an uneven distribution of coarse carbides (M_7C_3) formed within the tempered bainite and preferentially along prior austenite grain boundaries, and sub-boundaries within grains. The matrix contains a fine dispersion of vanadium carbide (M_4C_3) which provides the enhanced creep resistance in these alloys, (Figure 4).

Observations made on interrupted static creep specimens reveal the changes in structure accompanying deformation. Whilst still in the primary regime, a substructure begins to form, (Figure 5) and dislocations become pinned by carbides within the grains. During secondary creep, coarsening of the M_4C_3 and M_7C_3 precipitates occurs, especially the latter in grain boundaries, (Figure 6), and the dislocation substructure becomes more extensive. Cavities are visible during secondary creep and are associated with the large grain boundary particles (Figure 7). Final rupture is by the linkage of such cavities to produce a crack (Figure 8).

During cyclic straining, a dislocation substructure starts to form within 2% of life. Deformation appears to be concentrated initially in regions essentially free of M_7C_3 clusters, but containing long particles. With further cycling, the substructure becomes more extensive, and there is evidence of break up and redistribution of M_4C_3 which has been transported into the subcell walls (Figures 9 and 10) leaving the matrix relatively free of carbides.

The microstructural changes occurring during cyclic straining were very similar at ambient and elevated temperature.

In the case of sequential fatigue/creep interaction, the microstructure prior to creep testing would be similar to that in Figures 9 and 10. The changes induced by subsequent exposure to creep are shown in Figures 11 and 12. A cell structure exists throughout the specimen, and is smaller and more well defined than that observed in creep or fatigue alone. The areas of dense dislocation tangles that were present in the aligned regions of the fatigue specimens have been replaced by subcells (recovery has taken place). There has been growth and nucleation of M_7C_3 particles in and near grain boundaries, and also growth of M_4C_3. As in simple creep, final failure was intergranular and associated with large grain boundary carbides.

DISCUSSION

In agreement with other work on low alloy steels [1 - 3, 6, 7], creep failure at low stress levels has been shown to involve nucleation, growth and linkage of voids which are usually associated with large grain boundary precipitates. Both grain boundary and matrix diffusion are involved [8], and the creep ductility is low (~ 5%). At high stress levels (> 260 MN/m^2) [6, 9] the mechanism of creep changes, with deformation occurring within the grains, and involving dislocation motion. Fracture strains are

generally substantial (~ 50%).

The changes in the subsequent creep rate and fracture strain depend on the amount of prior fatigue involved. This may be attributed to cyclic softening, and subsequent recovery processes which occur within the grains. However, in the low stress brittle creep case the creep rupture process occurs in the grain boundaries and the effect of the cyclic softening is not as great as in the ductile creep case. It could be argued that there is a relative independence of processes taking place in the grains (in fatigue) and in the grain boundary regions (in brittle creep) which accounts for this moderate effect.

It would also be expected that prior fatigue would have a larger effect on high stress creep behaviour, since both processes involve deformation within the grains. Previous work [6, 9] on the same alloy supports this hypothesis when order of magnitude reductions in creep life are observed.

Since subcell formation, degradation of carbide morphology, and grain alignment occur in both fatigue and creep, it may be that similar softening processes are taking place in both cases. This would account for the observed absence of an effect of prior fatigue on the creep ductile-brittle transition stress [6].

Finally, it has been observed [6] that the creep rupture ductility subsequent to about 20% prior fatigue is approximately twice as high as that for creep of the virgin material. However, for greater amounts of prior fatigue, the rupture ductility subsequently decreases drastically, probably as the result of actual fatigue damage during the prior straining. Furthermore, preliminary observations indicate that the application of 10-20% prior fatigue reduces the void density at rupture relative to that for the zero and 50% prior fatigue cases.

CONCLUSIONS

(1) In low stress creep of this low alloy steel, dislocation substructures are formed and there is growth of M_4C_3 and M_7C_3 precipitates, especially the latter in grain boundaries. Failure involves nucleation, growth and linkage of voids at grain boundary carbides.

(2) Cyclic softening involves the formation of dislocation subcells, the destruction of vanadium carbide precipitates, and microstructural alignment in certain grains.

(3) A sequential fatigue/low stress creep interaction causes little change since the fracture process in the creep case occurs in grain boundaries. At high creep stress levels the influence of prior fatigue is substantial since both creep deformation and fracture processes occur within the grains. In many design situations, the former case applies, so that this type of interaction should not be too serious.

REFERENCES

1. MURPHY, M. C. and BRANCH, G. D., J.I.S.I., 207, 1969, 1347.
2. TIPLER, H. R. et al, Metals Tech., 2, 1975, 206.
3. TIPLER, H. R. and HOPKINS, B. E., Metal Science J., 10, 1976, 47.
4. DUNLOP, G. L. and HONEYCOMBE, R. W. K., Metal Sc. J., 10, 1976, 124.

5. DARLASTON, B. J. L. et al, Int. Conf. on "Creep and Fatigue in Elevated Temperature Applications", 1973 Philadelphia and 1974 Sheffield.
6. ELLISON, E. G. and PATERSON, A. J. F., Proc. Inst. Mech. Eng., 190, 1976, 321.
7. CANE, B. J., Metal Sci. J., 10, 1976, 29.
8. FROST, H. J. and ASHBY, M. F., Research Report, Dept. of Mech. Eng., University of Cambridge, "Deformation - Mechanism Maps for Pure Iron, Two Stainless Steels and a Low Alloy Ferritic Steel", July, 1975.
9. ELLISON, E. G. and PLUMBRIDGE, W. J., Second Int. Conf. on "Mechanical Behaviour of Materials", August, 1976, Boston, Massachusetts.

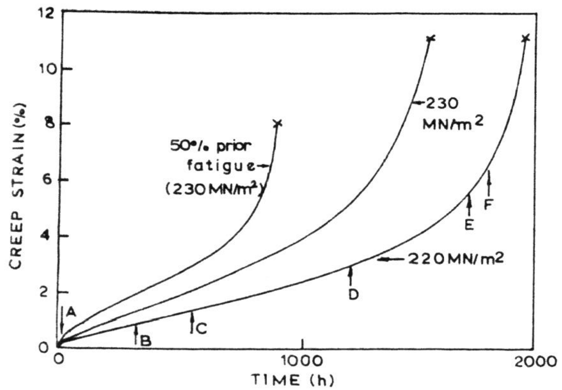

Figure 1 Strain-Time Curves for Creep Specimens. Interrupted Tests are as Indicated

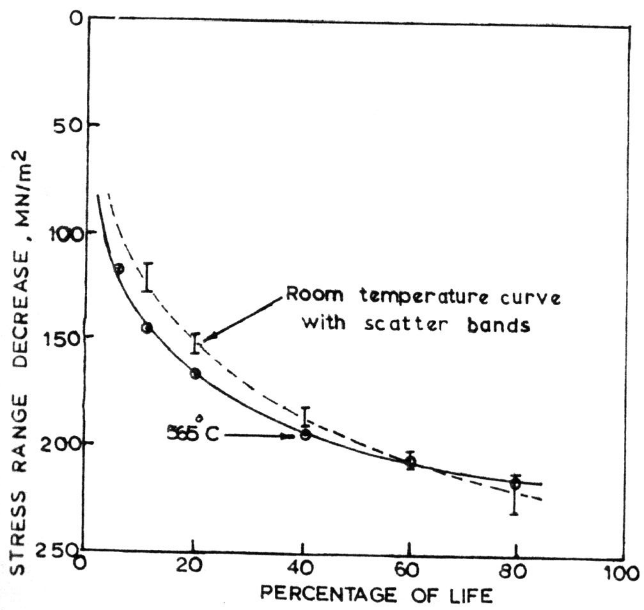

Figure 2 General Softening Characteristics at 565°C. Interrupted Tests are as Indicated

Figure 3 Thin Foil. Initial Structure

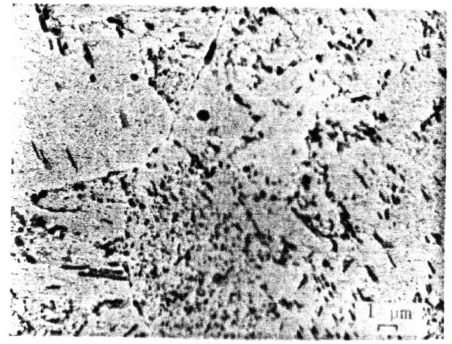

Figure 4 Carbon Extraction Replica. Initial Structure

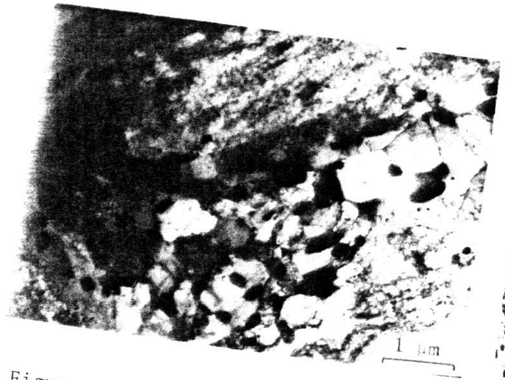

Figure 5 T.F. Specimen A

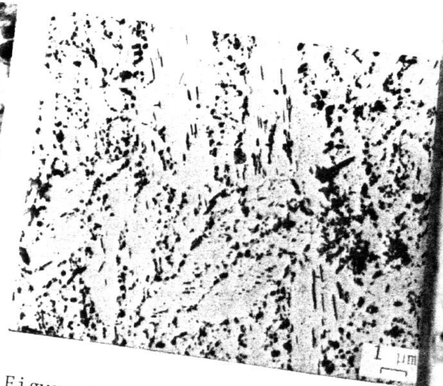

Figure 6 C.E.R. Specimen C

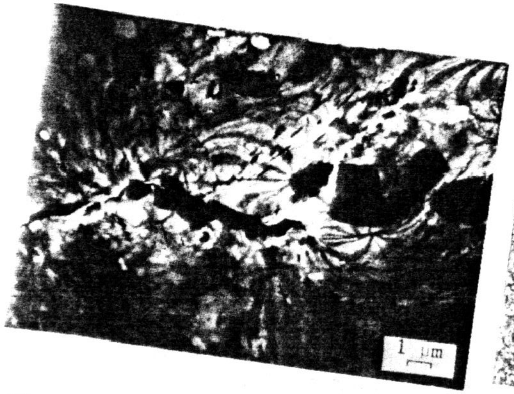

Figure 7 T.F. Specimen D

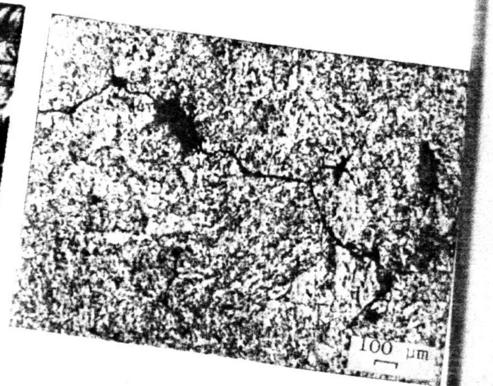

Figure 8 Optical Micrograph Specimen F

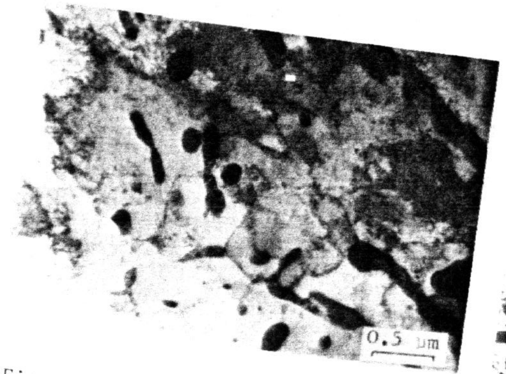

Figure 9 T.F. Specimen After 50% Fatigue

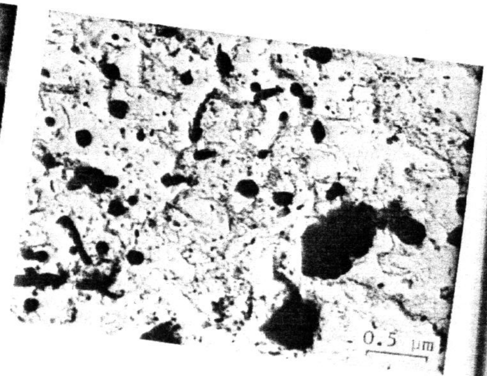

Figure 10 C.E.R. Specimen After 50% Fatigue

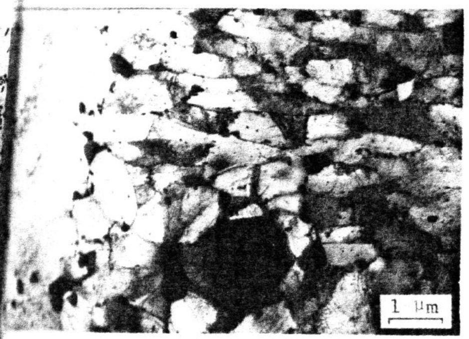

Figure 11 T.F. Failed Sequential Specimen

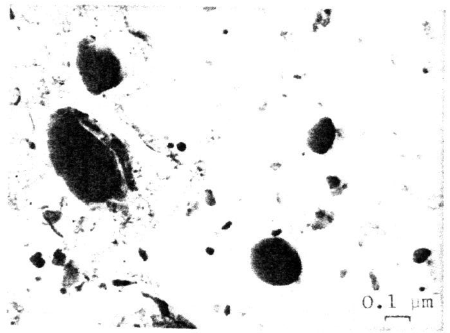

Figure 12 C.E.R. Failed Sequential Specimen

Table 1

Composition %								
C	Si	Mn	Ni	Cr	Mo	V	S	P
0.24	0.29	0.64	0.21	1.02	0.57	0.29	0.01	0.016
Heat treatment								
Austenitized 1000 °C, cooled to 675 °C at 50 °C/h. Held for 70 h, air cooled.								
Re-austenitized in salt bath at 975 °C. Transferred to another salt bath at 475 °C, held for 5 h, cooled to 300 °C max. Air cooled. This produced a fine grained bainitic structure.								
Reheat in a salt bath at 700 °C for 20 h after rough machining. This serves as a tempering treatment and stress relief H_v 243 - 255.								