THE INTERRELATION BETWEEN PRIOR CREEP DAMAGE AND CREEP CRACK PROPAGATION IN STEELS

D. Worswick** and R. Pilkington*

*Metallurgy Department, Manchester University, Grosvenor Street, Manchester M1 7HS, U.K.
**Now at: Research Department, Wilkinson Sword Ltd., 287 Acton Lane, London W4 5LE.

ABSTRACT

Controlled amounts of cavitation damage were introduced at 939K into smooth test pieces of $\frac{1}{2}$%Cr$\frac{1}{2}$%Mo$\frac{1}{4}$%V low alloy steel. These test pieces were then notched, pre-cracked at ambient temperature, then tested under constant load conditions at 823K in three point bending to monitor the creep crack growth characteristics. Crack growth rates were substantially increased compared to those for undamaged material and fracture lives were reduced by up to 60%. The parameters K_I and C* were found to correlate with crack growth rates in both undamaged and pre-cavitated specimens. The sizes and spacings of cavities were measured for both pre-damaged and virgin material and the results used to assess the validity of a model for creep crack growth. Metallographic evidence suggests that in the later stages of crack propagation diffusion effects may have a minimal part to play and that crack growth will be controlled by the accumulation of displacement at the crack tip.

KEYWORDS

Creep fracture, crack growth, cavitation, low alloy steel, rupture life.

INTRODUCTION

A substantial amount of evidence is now available to show that macroscopic creep crack growth may be described by an empirical law of the form: $da/dt = AK^n$, where a is crack length, K is stress intensity factor and A and n are constants (Siverns and Price, 1973; Ellison and Harper, 1978; Pilkington, 1979). This relation is however very dependent on the microstructural nature of the material (particularly for steels) such that under certain conditions there may be a better correlation between net section stress and creep crack growth rate (Harrison and Sandor, 1971). This is also observed for materials which exhibit substantial ductility under creep conditions such as austenitic stainless steels (Nicholson and Formby, 1975). It is necessary to have accurate knowledge of the state of a material which is entering service or which has seen extended life, to ensure that components are not allowed to enter or remain in service in an unsafe condition, but equally, that they are not rejected unnecessarily when a substantial amount of residual life is still available. Such an accurate assessment requires a detailed knowledge of the mechanisms of creep crack growth, but most of the data available to date has been generated on material in the "virgin" condition; that is, by allowing crack propagation through material which is exposed to stress at elevated temperature at the same time as crack propagation commences. In practical terms however, a component may fail after, say, 100,000 hours with the macroscopic crack propagating through material for a relatively small part of that time, when cavitation damage has already been generated in the material over a much longer time scale. Under these conditions it would be expected that creep crack growth behaviour would be different when compared to that for virgin material and it might reasonably be expected to be substantially faster.

An alternative way of considering creep crack growth on a macroscopic basis is to consider the accumulation of displacement at the crack tip. Following the initial work by Wells and McBride (1967); Pilkington et al. (1974) showed that a linear relationship exists between increasing crack length and increasing crack opening displacement and that this relation was both dependent on material and microstructure (Gooch, 1977; Jones and Pilkington, 1978). This concept was developed by Haigh (1975) to suggest that the crack growth rate could be related to the crack opening displacement rate and there is some experimental similarity in this approach to that based on the J contour integral known as $\dot{J}$ or C*.

The $\dot{J}$ or C* approach has been developed by Webster (1975) and by Harper and Ellison (1978) and is analogous to: $J = -dU/da$ where U is a function of potential energy and a is the crack length; thus J is a measure of the rate of release of energy with increasing crack length. However, $\dot{J}$ can be defined as $-dU^*/da$; where $U^* = \int_0^u P.d\dot{u}$ (P: applied load, u: displacement on the contour around the crack tip). $\dot{J}$ is therefore not equal to dJ/dt and will be referred to as C*. It has been suggested by Webster that this latter parameter could encompass both the initial stress intensity factor and net section stress descriptions of creep crack growth, but recent work (Riedel and Rice, 1979) shows that K and C* may characterise opposite situations dependent on the relationship between the creep zone size and the specimen geometry. It is true therefore that, at present, there is no single parameter which can be used to predict creep crack growth under widely different circumstances. Equally none of the macroscopic parameters take account of the development of background cavitation damage during long term service.

The purpose of the present paper is to show that prior cavitation damage increases creep crack growth rates and lowers rupture times. Previous work (Pilkington and co-workers, 1974, 1977, 1978) has already characterised the creep crack growth and smooth bar characteristics of this material in the virgin condition.

EXPERIMENTAL

A commercial $\frac{1}{2}$%Cr$\frac{1}{2}$%Mo$\frac{1}{4}$%V steel of composition: C, 0.11wt%; Cr, 0.42wt%; Mo, 0.51wt%; V, 0.25wt%; Mn, 0.51wt%; Ni, 0.13wt%; Si, 0.18wt%; P, 0.012wt%; S, 0.017wt%:, Sn, 0.01wt%; Ti, 0.002wt%; W, 0.02wt%; Al, <0.001wt%; Fe, balance; was given a full solution treatment of $\frac{1}{2}$h at 1533K with an air cool, followed by a tempering treatment of 24h at 943K. This gave a prior austenite grain size of 155 μm and a mixed bainite/ferrite structure containing approximately 97% bainite and 3% ferrite. The mechanical properties, both short term and long term, of this material were similar to those values documented in earlier work (Jones and Pilkington, 1978).

Specimens of section 18.75 mm wide x 12 mm thick, with a gauge length of 152 mm, were machined with screwed heads 24 mm diameter. In this form two series of specimens were pre-strained at 939K, the first to 0.26% plastic strain and the second to 0.58% plastic strain. The specimens were then unloaded and cooled to ambient temperature and the threaded ends of the specimens cut off, producing rectangular section specimens suitable for three point bend testing. An additional series of bend specimens was machined without screwed heads for testing in the virgin condition. All specimens were notched and fatigue cracked to approximately 1/3 of the specimen width at ambient temperature, in accordance with the requirements of the relevant ASTM specification (A.S.T.M., S.T.P. 410, 1967). Three point bend tests under constant load conditions were carried out at 823K (±2K) under vacuum. Crack growth was monitored by a D.C. potential difference technique, accuracy of measurement (±30 μm) being checked by visual examination of the fracture surfaces. Loading point displacements and crack opening displacements were measured by transducer during the course of the bend tests. This enabled the function $C^*(\dot{\Delta}) = P \times \dot{\Delta}/Ba$ to be calculated where P = specimen load, $\dot{\Delta}$ displacement rate at the loading point, B = specimen thickness, a = crack length. Further, a parameter $C^*(\dot{\delta})$ was calculated, where this parameter was similar to the $C^*(\dot{\Delta})$ with the modification that the crack opening displacement rate ($\dot{\delta}$) was used instead of beam displacement rate ($\dot{\Delta}$).

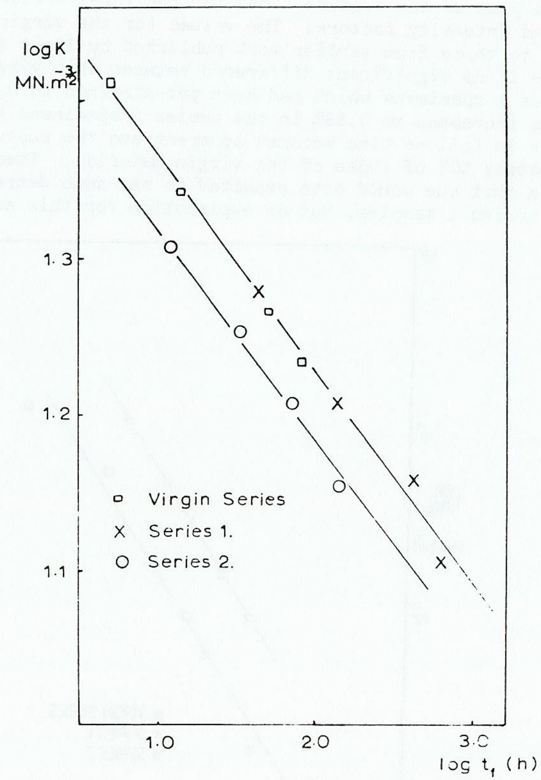

Fig. 1. Variation of time to rupture with initial stress intensity factor.

After test, metallographic examination of specimens was carried out using both the optical and scanning electron microscope. Two types of examination were carried out. The first involved cleaving the specimens open at 77K, well away from the creep crack growth region to facilitate an assessment of the amount of damage introduced into the material during pre-straining. The second involved fracturing the specimen at 77K, in the sense of continuing the plane of propagation of the macroscopic creep crack (since high temperature failure in the present work did not involve the separation of the fracture surfaces at the elevated temperature). It was thus possible to determine the cavitation distribution of both the macroscopic creep crack, and the distribution in the cleavage region ahead of this crack. These studies enabled the extent of cavitation to be determined, as well as the sites and sizes of individual cavities. The use of analytical X-ray facilities with the scanning electron microscope also permitted the identification of carbides and a quantitative assessment of their size and distribution was also made.

RESULTS AND DISCUSSION

Macroscopic Data

Figure 1 is a plot of the rupture times for the three series as a function of the initial stress intensity factors. The values for the virgin specimens correspond very closely to those from earlier work published by Jones and Pilkington (1978). However there is no significant difference between this virgin series of specimens and the series 1 specimens which had been pre-strained to 0.26% strain. When the pre-strain is increased to 0.58% in the series 2 specimens however, then a substantial decrease in failure time becomes apparent and the rupture times of these tests are approximately 60% of those of the virgin material. These results are somewhat surprising in that one would have expected to see some decrease in the fracture life of the series 1 samples, but an explanation for this effect exists within the

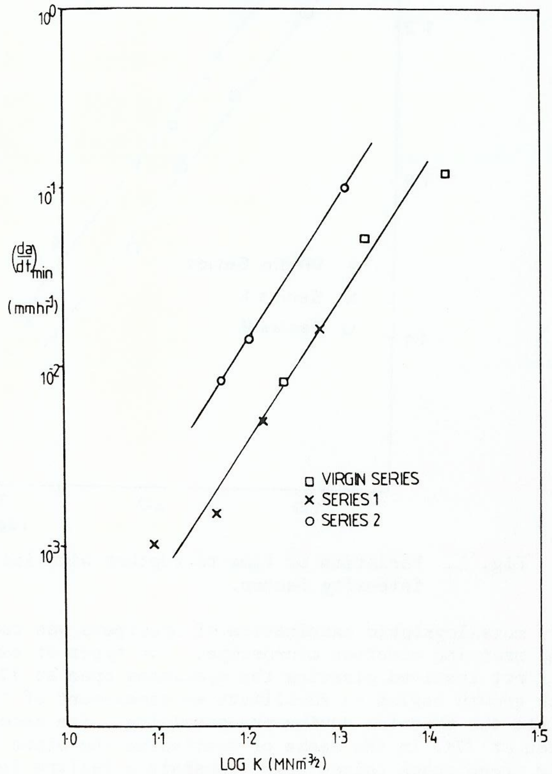

Fig. 2. Variation of minimum crack growth rates against initial stress intensity factor.

detailed metallography of the cavitation behaviour and has been presented elsewhere (Worswick and Pilkington, 1979). Figure 2 shows the variation of the minimum crack growth rates against stress intensity factor, with the accelerated rates which occur after prior damage in the series 2 specimens being clearly detectable. Similar behaviour was also obtained when the values for the crack opening displacement rates were plotted as a function of stress intensity factor. Thus a linear relationship was observed when plotting da/dt against $d\delta/dt$ (Haigh, 1975) and a derived variation of this observation is shown in figure 3 which is a plot of da/dt against $C^*(\dot{\delta})$. It is clear that the presence of prior damage does not influence the subsequent derived value of $C^*(\dot{\delta})$.

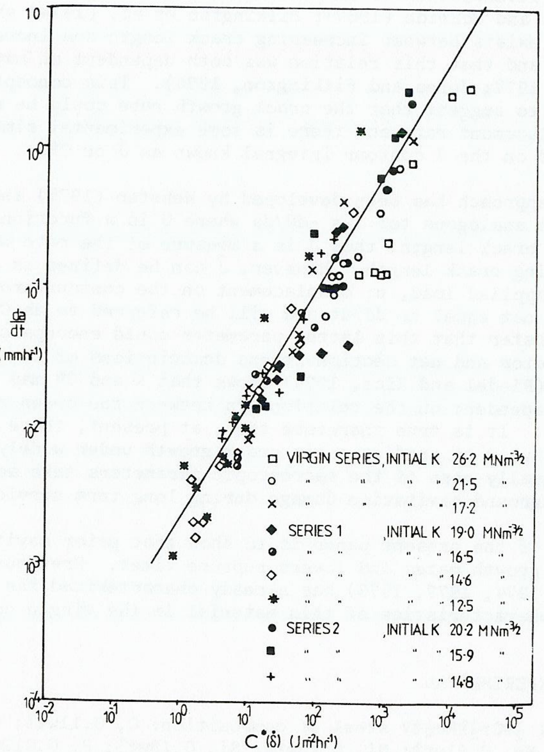

Fig. 3. Variation of crack growth rates against parameter $C^*(\dot{\delta})$.

Microscopic Examination

Transmission electron microscope examination of carbon extraction replicas revealed the presence of coarse M_3C carbides within the matrix of the bainitic regions with a fine distribution of V_4C_3 in the ferritic regions and coarse V_4C_3 at prior austenite grain boundaries. Carbide sizes and spacings are detailed in Table 1. After failure the quantitative information regarding cavity sizes and spacings, both after pre-strain, and also after fracture by crack growth was obtained and these results are also summarised in Table 1. It will be seen that the average cavity diameter of the first series of specimens was 1.285 μm after pre-strain and 1.85 μm

TABLE 1 Summary of Size and Spacing Measurements of Carbides and Cavities

Carbides	Mean diameter (μm)	Mean spacing (μm)
VC (matrix)	0.041	0.11
VC (grain boundary)	0.21	0.28
M_3C (grain boundary)	1.08	5.14
Predamage cavities		
Series 1	1.29	1.91
Series 2	1.85	2.35
Creep crack cavities		
Virgin	0.62	0.63
Series 1	1.04	1.05
Series 2	1.01	1.02

for the second series. After creep crack growth these average sizes were reduced to 1.04 μm and 1.01 μm, indicating that new cavities had been nucleated during creep crack growth. It can also be inferred from these observations (Worswick and Pilkington, 1979) that in the second series some cavity coalescence has occurred. In addition the relation between the carbide spacings and cavity spacings fits the established trend for these steels (Cane, 1976; Miller and Pilkington, 1978a, 1978b) that cavity nucleation is extremely inhomogeneous and that it probably occurs selectively at larger carbides.

Theoretical Assessment

To attempt to correlate the present experimental results with a theoretical model the results of da/dt against K are plotted as mean lines in figure 4. On this same figure are plotted the results of a theoretical calculation based on the model proposed by Miller and Pilkington (1980). The basis of this model is that in the final stages of creep crack propagation, crack growth occurs by the nucleation and growth of cavities with cavity growth being controlled by the power law creep of the surrounding matrix with little, if any, assistance from surface diffusion and/or grain boundary diffusion. It will be seen that the prediction gives a marginally higher suggested value for the crack growth rate at a given value of K, but it should be noted that the model assumes that an elastic stress concentration is maintained at the tip of the crack. Clearly under creep conditions there may be some limited relaxation of this constraint even though the material is in a creep brittle condition. (It should also be pointed out that the model takes no account of the continuous nucleation of cavities which is clearly observed in the present work.) Thus in practical terms, one would expect to find the experimental values being somewhat less than those predicted theoretically.

It is however necessary to provide supporting evidence for the suggestion that diffusional effects play a minimal role in this context. Recent work (Beere and Speight, 1978) has pointed out that for cavity growth to be of a crack like nature the surface diffusion coefficient should be less than the grain boundary diffusion coefficient. Although substantial variation exists in experimental data, there is reason to believe that at 823K the surface diffusion coefficients are in practice lower than the grain boundary diffusion coefficients for this material. However figure 5 is a micrograph of the cavitation damage associated with the creep crack growth of the present work and it will be seen that rather than having a crack like or finger like nature (Fields and Ashby, 1976; Miller and Pilkington, 1978) on the contrary, the cavities are almost entirely spherical, whether or not coalescence has occurred. It is felt that this is conclusive evidence to suggest in the

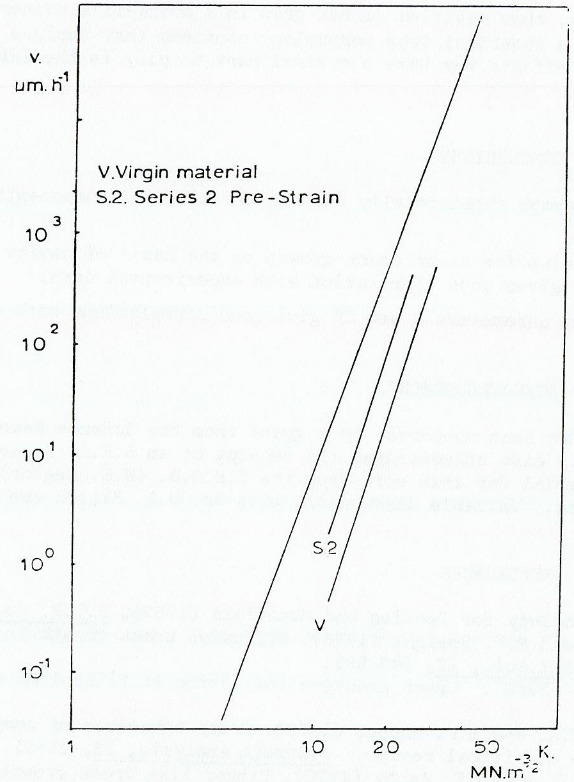

Fig. 4. Comparison of experimental and theoretical crack growth rates (after Miller and Pilkington, 1980).

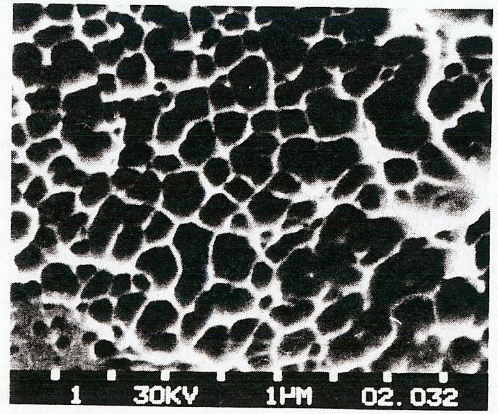

Fig. 5. Scanning electron micrograph showing spherical nature of cavities.

present work, that cavities do not grow in a crack-like manner and that the presence of a spherical type morphology confirms that surface and grain boundary diffusional effects may have a minimal part to play in the later stages of crack growth.

CONCLUSIONS

1. Prior damage substantially reduces the lives of components containing creep cracks.

2. A mechanism for creep crack growth on the basis of cavity growth controlled by deformation gives good correlation with experimental data.

3. Both the parameters K and C* give good correlations with empirical data.

ACKNOWLEDGEMENTS

This work has been supported by a grant from the Science Research Council and one of us (D.W.) also acknowledges the receipt of an S.R.C. Studentship. The provision of the material for this work from the C.E.G.B. (N.E. Region) is gratefully acknowledged. Valuable discussions with Dr. D.A. Miller are also acknowledged.

REFERENCES

American Society for Testing and Materials (1967), S.T.P. No. 410.
Beere, W. and M.V. Speight (1978). Diffusive crack growth in plastically deforming solid. Met.Sci., 12, 593-599.
Cane, B.J. (1976). Creep fracture initiation in $2\frac{1}{4}$%Cr-1%Mo steel. Met.Sci., 10, 29-34.
Ellison, E.G. and M.P. Harper (1978). Creep behaviour of components containing cracks - a critical review. J.Strain Analysis, 13, 35-51.
Fields, R.J. and M.F. Ashby (1976). Finger like crack growth in solids and liquids. Phil.Mag., 33, 33-46.
Gooch, D.J. (1977). The effect of microstructure on creep crack growth in notched bend tests on 0.5Cr-0.5Mo-0.25V steel. Mat.Sci.Eng., 27, 57-68.
Haigh, J.R. (1975). The mechanisms of high temperature crack growth. Mat.Sci.Eng., 20, 213-223.
Harper, M.P. and Ellison, E.G. (1977). The use of the C* parameter in predicting creep crack propagation rates. J.Strain Analysis, 12, 167-179.
Harrison, C.B. and G.N. Sandor (1971). High temperature crack growth in low cycle fatigue. Eng.Fract.Mech., 3, 403-420.
Jones, C.L. and R. Pilkington (1977). The effect of temperature on creep crack growth in a 1%Cr$\frac{1}{2}$%Mo$\frac{1}{4}$%V steel. Proc.4th Int.Conf.on Fracture, University of Waterloo, Vol. 2, 641.647.
Jones, C.L. and R. Pilkington (1978). The influence of microstructure on creep crack growth in 0.5%Cr, 0.5%Mo, 0.25%V steel. Met.Trans., 9A, 865-871.
Miller, D.A. and R. Pilkington (1978). The effect of temperature and microstructure on the creep damage found in low alloy ferritic steels. Met.Trans., 9A, 1221-27.
Miller, D.A. and R. Pilkington (1980). Diffusion and deformation controlled creep crack growth. Met.Trans., 11A, 177-180.
Nicholson, R.D. and C.L. Formby (1975). The influence of various fracture mechanics methods at creep temperatures. Int.J.of Fracture, 11, 595-602.
Pilkington, R., D. Hutchinson and C.L. Jones (1974). High temperature crack opening displacement measurements in a ferritic steel. Met.Sci., 8, 237-241.
Pilkington, R. (1979). Creep crack growth in low alloy steels. Met.Sci.,13,555-564.
Riedel, H. and J.R. Rice (1979). Tensile cracks in creeping solids. A.S.T.M., S.T.P. No. 700, to be published.
Siverns, M.J. and A.T. Price (1973). Crack propagation under creep conditions in a quenched $2\frac{1}{4}$ chromium 1 molybdenum steel. Int.J.of Fracture, 9, 199-207.
Webster, G.A. (1975). The application of fracture mechanics to creep cracking. Proc.Conf.on Mech.and Phys.Fracture, Cambridge, Paper 18.
Wells, A.A. and F.H. McBride (1967). Application of fracture mechanics to high temperature creep rupture. Can.Met.Q., 6, 347-368.
Worswick, D. and R. Pilkington (1979). The effect of prior damage on creep crack propagation in ferritic steels. Proc.5th Int.Conf.on Strength of Metals and Alloys, Aachen, 1, 409.416.