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THE EFFECT OF TEMPERATURE ON CREEP CRACK GROWTH IN A 1/2% Cr 1/2% Mo 1/4% V STEEL

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INTRODUCTION

Considerable attention has been given to the problem of creep crack propagation in steels over the past few years following the initial work of Siverns and Price [1]. It is now fairly well documented that an empirical relationship of the form $da/dt = AK^n$ exists, for some materials, between the rate of creep crack propagation, da/dt, and the stress intensity factor, K, as though the material were behaving in a linear elastic manner [2 - 4]. On the other hand evidence is available that there is a much better correlation between da/dt and net section stress [5, 6] and also, in some cases, between da/dt and an arbitrary reference stress [7, 8]. Little definitive work has been carried out to determine the activation energies which may be required for crack growth and crack opening displacement (C.O.D.) in these temperature ranges. Evidence which is now available [9, 10] suggests that creep crack growth may be displacement controlled and this displacement control will be determined by the nature of the creep behaviour of the material around the crack tip. Activation energies may therefore give some insight into the metallurgical processes which will be controlling this deformation. Understanding of these processes may then allow more accurate prediction of creep crack growth rates and possibly the times to failure of service components containing crack like defects.

EXPERIMENTAL

1/2% Cr, 1/2% Mo, 1/4% V steel, vacuum melted from Swedish iron to a composition given in Table 1, was given a full solution treatment of 1/2 h. at 1533 K with an air cool, followed by a tempering treatment of 24 h. at 953 K. This gave a prior austenite grain size of 93 μm and a mixed bainite/ferrite structure containing approximately 95% bainite and 5% ferrite. Specimens 18.75 mm wide by 12 mm thick by 152 mm long were machined, notched and fatigue cracked to approximately 1/3 of the specimen width. Three-point bend tests under constant load conditions were carried out at 803 K, 823 K and 843 K (all ± 2 K) 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. The C.O.D. was measured either by using a paddle technique [4] or by a modified version of the high temperature strain gauge developed by Noltingk et al [11]. Corrections were made to all readings to compensate for increasing crack lengths, so that all measurements were of *crack tip* C.O.D.'s. Comparability of the mechanical properties of this material to commercial material was ensured by carrying out short term hot tensile tests and creep tests.

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RESULTS

The crack lengths and C.O.D.'s increase with time in the same manner as that observed in earlier work [4], with a constant rate of increase for both parameters for most of each test, followed by an accelerating stage leading to final failure. Variations of accelerating da/dt with K for the material at the 3 different test temperatures are plotted in Figure 1. The substantial difference in these rates will be clearly apparent, with an order of magnitude rise in da/dt when the temperature is increased by 40 K. It will also be clear, although there is a linear relationship between da/dt and K for each individual test, it is not possible to make such a correlation from test to test. This is particularly true for the results of tests carried out at 843 K. The gradients ('n') vary from $\sim$12 to $\sim$30. These are somewhat greater than values obtained in previous work [1 - 4] and hence the variations of constant crack growth rates, $(da/dt)_s$, with initial K have also been studied. Figure 2 shows the results for this material, where the slopes are now of the order of 7.5. Similar plots are obtained for the variation of da/dt with net section stress (Figures 3, 4). (Net section stress in this instance being defined as the crack tip fibre stress). Again the gradients are of similar magnitudes for the accelerating crack growth stages ($\sim$12 to $\sim$30) and $\sim$7.5 for the steady state stage. Variations of C.O.D. with time and the variations of the rates of change of C.O.D. ($d\delta/dt$) with K and net section stress respectively have also been obtained for these specimens. The gradients for the accelerating parts of the curves vary between $\sim$9 and 30 whereas when the constant $(d\delta/dt)_s$ values are plotted against initial K and net section stress, gradients $\sim$6.9 are obtained.

It is found that, except in the initial stages of each test, there is a linear relationship between the crack length and the C.O.D. and Figure 5 shows typical plots for specimens tested at 843 K. (Only one set of points is given for the sake of clarity). Similar curves are obtained for tests at 823 and 803 K and in all cases the gradients vary from $\sim$41 to $\sim$51. These gradients are sometimes referred to in terms of the crack geometry as "aspect ratios", *i.e.*, the ratio of increase in crack length to increase in C.O.D. When the values of da/dt are plotted with $d\delta/dt$ on a logarithmic scale, linear relationships are obtained over both the linear and accelerating regions of each test. Figure 6 shows that these results can all be contained with a scatter band with only small deviations from an average gradient of unity. Figures 7 and 8 show, respectively, the variation of $(da/dt)_s$ and $(d\delta/dt)_s$ with temperature. Values of apparent activation energies for crack growth and change in C.O.D. have been determined from these curves and are found to be 252 kJ mol^{-1} and 268 kJ mol^{-1} respectively.

DISCUSSION

The values of the index 'n' obtained in the present work for the accelerating stages of crack growth are rather high, but are comparable with results obtained by Gooch [12]. When consideration is made of this index for the steady state crack growth behaviour however, much lower values of the order of 7.5 are obtained. Recent papers by Williams and Price [13], and by Ellison and Neate [14] have suggested that, for materials which are very brittle under creep conditions, a correlation may be expected between da/dt and K for 'n' < $\sim$5 and that a reference stress description would be more appropriate for 'n' > $\sim$10. The present material is extremely brittle ($\varepsilon_f \sim 0.3\%$ after 10,000 h. at 823 K), and one might expect to see a correlation from test to test between da/dt and K. Further, attention has been drawn [13, 14] to the possibility that a K correlation might be expected

if highly localized creep damage is observed. Metallographic evidence [4, 15] shows clearly that cavitation damage is closely confined to the crack tip, in the present work being no more than 350 µm in extent ahead of the crack. (Similar observations have also been reported by Gooch [12]). Thus, even with conditions which are favourable for the observation of a correlation between da/dt and K, such a correlation is only obtained for each separate test and *not* from test to test. The effect of specimen geometry is also important but in the present work, straight crack fronts were obtained, except for slight curvature over a distance of $\sim$1 mm from the free surfaces of the specimen. This implies that constraint was retained at the crack tip, again enhancing the potential "brittleness" of the test conditions [14]. Thus the present results suggest that K is inadequate to describe creep crack growth in the present work and that its use as a parameter in assessing creep crack growth behaviour should be treated with extreme caution. It is thus necessary to consider an alternative approach.

One of the most important features of the present work is the evidence of a substantial increase in crack opening displacement for a relatively small increase in crack length. This is followed by an approximately linear relationship between crack length and C.O.D. (Figure 5). It is felt that this may be evidence for the control of creep crack growth in this material by the local displacement of material around the crack tip. It is suggested that it is necessary for displacement to build up at the crack tip before substantial crack growth occurs on a macroscopic scale. After this critical displacement has been achieved, if the overall C.O.D. is the controlling parameter for crack growth, a linear relation between crack length and C.O.D. could be obtained.

It is also necessary to consider the absence of change in aspect ratios of the cracks with change in temperature. Increases in temperature will lead to a general loss of strength over the whole of a specimen, but it might be expected that if the stress dependence of the material changed with temperature, there would be a greater loss of strength in the local region around the crack tip with this increase in temperature, thus leading to a substantial decrease in the aspect ratio, i.e., more deformation of the material adjacent to the crack tip with increasing temperature. This is not observed, nor is there any detectable change in the relationship between da/dt and $d\delta/dt$.

The general slope of the curves relating ℓn da/dt to ℓn $d\delta/dt$ is approximately unity. It has been observed by Haigh and Richards [7] that for a more creep ductile material, a gradient of the order of 0.8 may exist and this is attributed to weakening of the material around the notch root. It is felt that the slope of approximately unity in the present work is further support for the absence of any substantial weakening in the material around the notch root relative to the remainder of the specimen. Clearly creep deformation must be taking place but the high values of aspect ratios implies that little weakening occurs. The absence of any relaxation of the stress concentration around the crack tip would lead to very rapid increases in crack length for small crack opening displacements. It will also be seen that there are substantial increases in the absolute values of the crack growth rates with increasing temperature. These are up to an order of magnitude for a temperature increase of 40K. Clearly the reason for these increases is that the deformation process in both the matrix and the grain boundaries will be thermally activated and thus the rate of all deformation must increase, resulting in an increased rate of crack opening displacement and hence an increase in the crack growth rates.

Examination of the values of apparent activation energies for both creep crack growth and creep crack opening displacement suggests that there is a close relation between these two processes. The values of 252 kJ mol^{-1} for da/dt and 268 kJ mol^{-1} for $d\delta/dt$ are lower than other values which have been reported in the literature for the creep of low alloy ferritic steels, although it must be re-emphasized that the present results, since derived from individual tests at separate temperatures, are in the creep sense, 'apparent' activation energies [16]. However it is clear [16] that for this type of material, a comparison can be made with the 'true' activation energies reported in the literature for the temperature and test conditions under consideration in the present work. The results of Myers at el [17] suggest that, in a high stress situation, values of $\sim$410 kJ mol^{-1} would be applicable and only at relatively low stresses, such as those which exist in the components of plant in service conditions, would values of $\sim$300 kJ mol^{-1} be obtained. Furthermore, the work of Collins [18] on artificial ferritic steels suggests, on the basis of strain rates rather than stress levels, that at high strain rates and high 'n' values ($n \sim 10$), values $\sim$620 kJ mol^{-1} would be applicable and that only at low strain rates, with 'n' $\sim$ 3, would values of $\sim$300 kJ mol^{-1} apply. The higher value is applicable when the deformation is controlled by the strength of the matrix, but the low activation energy values would be obtained when, it was suggested, grain boundary deformation was the controlling parameter. The values quoted are, of course, applicable to plain bar creep tests and the application of these values to either notched bar, or to creep crack growth studies, must be treated with caution. It might be inferred from these values that grain boundary deformation was the significant deformation mode controlling crack growth. However, the values obtained in the present work are also very similar to those for the self-diffusion of Fe [19], and this information, together with the 'n' values of $\sim$7 which are obtained (the intermediate range in the sense of Collins' results), suggest that no single mechanism will be controlling the build-up of displacement at the crack tip. Nevertheless, this aspect of the work is further evidence for the concept of displacement control over crack growth. This concept is now well supported for creep ductile materials [6], but its extension to such relatively brittle situations is only now becoming apparent. It is suggested therefore, that in this material in this heat treatment, creep crack growth is controlled by the accumulation of displacement of material in the local region around a crack tip.

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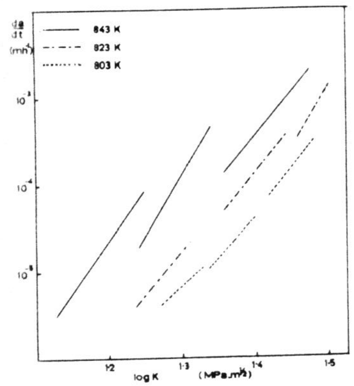

Figure 1 Variations of ℓn(da/dt) with ℓn K for Three Test Temperatures

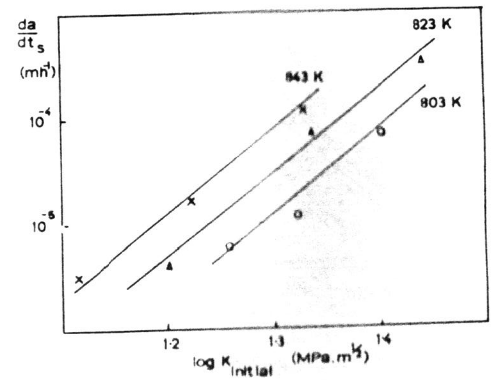

Figure 2 Variations of $\ell n(da/dt)_s$ with ℓn K for Three Test Temperatures

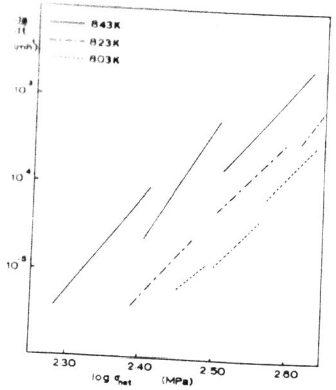

Figure 3 Variations of $\ell n(da/dt)$ with $\ell n\ \sigma_{net}$ for Three Test Temperatures

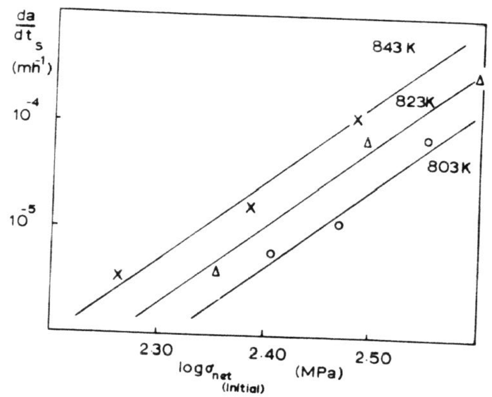

Figure 4 Variations of $\ell n(da/dt)_s$ with $\ell n(\sigma_{net})$ for Three Test Temperatures

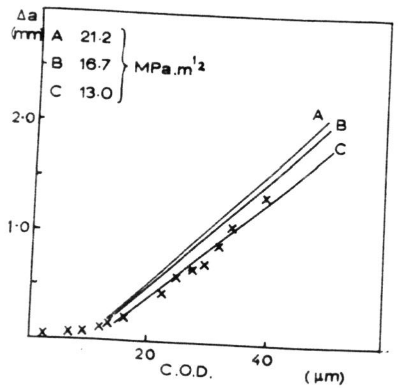

Figure 5 Variations of Crack Length (Δa) Against C.O.D. for Specimens Tested at 843 K

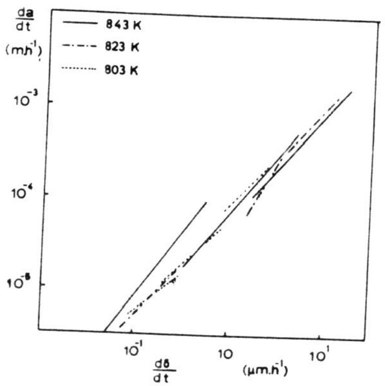

Figure 6 Variations of ℓn da/dt with ℓn dδ/dt for Three Test Temperatures

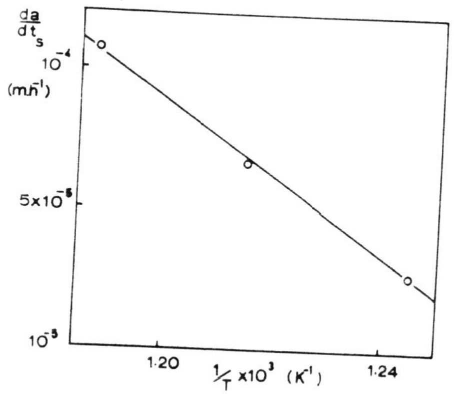

Figure 7 Variation of $\ell n(da/dt)_s$ with Temperature

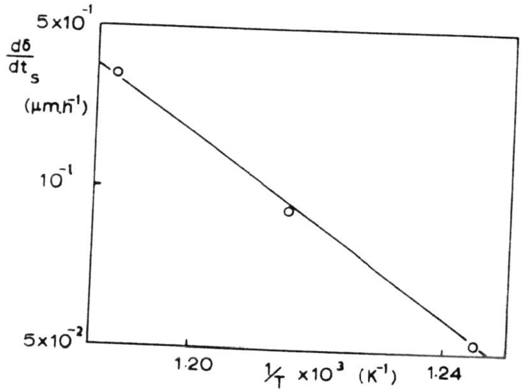

Figure 8 Variation of $\ell n(d\delta/dt)_s$ with Temperature

Table 1 Analysis of Material

	Al	As	B	C	Cr	Co	Cu	Mn	Mo
Sample 1	<0.01	<0.01	<0.01	0.16	0.56	$\sim$0.01	<0.01	0.63	0.50
Sample 2	<0.01	<0.01	<0.01	0.14	0.53	<0.01	<0.01	0.59	0.48
	Ni	**Nb**	**P**	**Si**	**S**	**Sn**	**Ti**	**W**	**V**
Sample 1	<0.01	$\sim$0.01	<0.005	0.02	0.009	<0.01	<0.01	$\sim$0.01	0.26
Sample 2	<0.01	$\sim$0.01	<0.005	0.02	0.008	<0.01	<0.01	$\sim$0.01	0.23