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MICROSTRUCTURAL EFFECTS ON THE CREEP FRACTURE OF 1/2 Cr Mo V STEEL

G. L. Dunlop* and R. W. K. Honeycombe**

INTRODUCTION

Considerable attention has recently been paid to the elevated temperature fracture of low alloy steels of the Cr Mo V and 2 1/4 Cr Mo types which are used extensively in steam plant for electricity generation. A number of workers have studied the propagation of single cracks in these materials at high temperatures using various fracture mechanics parameters such as the stress intensity factor, K, or the crack opening displacement, COD, in order to characterize the rate of crack propagation under dead loading conditions or during high temperature fatigue [1 - 4]. Others have considered the fracture of smooth bar tensile creep specimens as a function of both stress and microstructure, which in turn is directly attributable to the heat treatment path which the material has undergone [5, 6, 7]. The linking factor in all of this work is the fact that fracture is intergranular in nature and is highly dependent upon the microstructural condition of the steel in question.

An inherent property of creep resistant low alloy steels is that they are transformable and their detailed microstructures depend strongly upon the manner in which transformation from austenite occurs. In practise, heavy sections such as the rotors or casings of turbines are normalized by air cooling or "gashing" (water spraying) from the austenitizing temperature and are subsequently tempered at 973 K. Depending upon the cooling rate, which varies continuously through a thick section, the resultant primary transformation product may be ferrite, bainite or martensite. Because of the economic importance of this type of steel, considerable efforts have been made to determine which transformation product provides the best creep properties in a given creep situation [5 - 11]. Recently, detailed electron microscopy has defined the range of microstructure which it is possible to achieve in a 1/2 Cr Mo V steel [12]. This was done by isothermally transforming at various sub-critical temperatures and subsequently tempering. The creep strengths of these microstructures have been determined and have been shown to be mainly controlled by the state of the VC precipitate dispersion which results from precipitation during either direct transformation to ferrite or tempering of bainite and martensite [13]. This is in accordance with previous work which has shown that, in normalized and tempered microstructures of similar steels, there is a continuous increase in creep strength with decreasing interparticle spacings. In this paper the important effect which microstructure is found to have upon the creep fracture process and ductility in smooth bar specimens of 1/2 Cr Mo V steel is reported. The experimental details have been reported elsewhere [13]. Prior to creep testing at 823 K the specimens were solution treated at 1423 K and isothermally transformed at various sub-critical temperatures for 15 minutes. Specimens with bainitic and martensitic microstructures, were tempered for three hours at 973 K. The room temperature hardnesses

* Chalmers University of Technology, Gothenburg, Sweden
** University of Cambridge, Cambridge, England

MS289

obtained through these heat treatments are given in Table 1.

INFLUENCE OF HEAT TREATMENT ON CREEP STRENGTH

A typical set of results for the variation of secondary creep rate with heat treatment condition are shown in Figure 1. The lowest creep rates, or highest strengths, were obtained for structures which had either been quenched or which had been transformed at temperatures which were close to the bay between the ferrite and bainite transformation curves in the TTT diagram. These primary transformation temperatures correspond to the highest room temperature hardnesses in each generic microstructure (Table 1) and also correspond to the finest dispersion of VC precipitate obtained in either equiaxed ferrite or granular bainite [13].

INFLUENCE OF HEAT TREATMENT ON CREEP DUCTILITY

Elongation and reduction-in-area results for two sets of tests are shown in Figure 2. There was a general trend for increasing elongation with increasing transformation temperature but the reduction-in-area results (which are more likely to reflect small differences in the fracture process) show a considerable variation from this trend. For ferritic specimens, the reduction-in-area decreased with decreasing primary transformation temperature but a sharp increase occurred as the transformation moved into the bainitic range. A continuous decrease then occurred as the transformation temperature was further lowered. Tempered martensite exhibited an intermediate value of reduction-in-area.

METALLOGRAPHY OF FRACTURE

The tests at the stress level of 340 MPa were conducted in vacuum to enable the fracture surfaces to be examined by scanning electron microscopy.

Except for the specimen transformed at the highest temperature of 1023 K all of the fracture surfaces contained two macroscopically distinct regions: (i) a brittle intergranular region where the fracture had followed a grain boundary path which was, on average, approximately perpendicular to the stress axis; (ii) a ductile lip where the fracture was transgranular and inclined to the stress axis at an angle of approximately 45°. More detailed examination showed that for martensitic and bainitic specimens (except for that transformed at 873 K) the intergranular fracture path was along prior-austenite grain boundaries, while in ferrite specimens the intergranular fracture path was along ferrite grain boundaries.

Optical metallography revealed that intergranular cavitation occurred in all of the specimens but with the essential difference that cavitation was associated with ferrite grain boundaries in directly transformed specimens while in bainitic and martensitic specimens cavitation was associated with prior-austenite grain boundaries (Figure 3). In all cases cavities were preferentially situated on grain boundaries which were approximately perpendicular to the stress axis. Closer to the fracture surface, cavities were seen to have interlinked to form cracks which often occupied whole ferrite or prior-austenite grain boundary facets. These cracks appeared to be barred from further propagation at triple junctions or, especially in the directly transformed ferrite microstructures, at irregularities in the grain boundaries.

For directly transformed ferrite specimens, the decrease in fracture ductility with decreasing transformation temperature was associated with a change in the appearance of the fracture surface. As shown in Figure 4 the intergranular fracture surfaces of specimens transformed at the highest temperatures had a more ductile appearance than did the fracture surfaces of specimens transformed at lower temperatures in the ferrite range.

In bainitic specimens the amount of cavitation in the gauge length away from the fracture surface decreased with decreasing primary transformation temperature. In specimens where cavities were visible at some distance from the fracture surface the cavities were small and rounded and often lay in "strings" along the grain boundaries (Figure 3b). A large number of cavities can be seen on a prior-austenite grain boundary in the final fracture surface in Figure 5.

In specimens transformed at the lower temperatures in the bainite range, cracks at the prior-austenite grain boundaries close to the fracture surface were straight and generally occupied one or two prior-austenite grain facets. Granular bainite specimens transformed at the higher transformation temperatures (873 K) showed improved ductility and contained less well defined cracks along the austenite grain boundaries. Close examination of such specimens after creep failure showed that the fracture surfaces were highly irregular on a fine scale and that the fracture path was displaced a little from the position of the prior-austenite grain boundaries (Figure 6). Instead, the crack path followed the interface between a narrow rim of directly transformed ferrite and the tempered bainite.

DISCUSSION

From the results it is apparent that the fracture ductility of this steel, as exemplified by reduction-in-area, is not solely related to its microstructural condition through the dependence of ductility upon overall creep strength. If this were the case, then tempered bainite which was initially transformed at 873 K should show a lower reduction-in-area than that initially transformed at lower temperatures. There is clearly a strong effect of the morphology of the intergranular crack path which sometimes may be overriding. An example of this may be seen by comparing the ductility of tempered bainite which had been initially transformed at 673 K and ferrite which had been directly transformed at 1023 K. The two microstructures exhibited similar secondary creep strengths (Figure 1) but the directly transformed ferrite exhibited a considerably higher fracture ductility (Figure 2). An important point comes to light here. The austenite grains from which directly transformed ferrite grains derive are about five times as large as the ferrite grains. Thus, the effective grain size for intergranular fracture is considerably larger for tempered bainite and martensite. This permits more rapid fracture in these microstructures since the ease of crack interlinkage and propagation to fracture increases with increasing grain size [16].

When directly transformed ferrite alone is considered, then the overall strength of the microstructure is over-riding. The fracture ductility decreases as the creep strength is raised, i.e., when the transformation temperature is lowered from 1048 K to 923 K. The influence of matrix strength can be seen by comparing the two fracture surfaces shown in Figure 4. Although in both cases fracture resulted from intergranular cavitation it is clear that considerably more local plastic deformation was involved in the fracture process of the weaker specimen transformed at 1023 K.

The relatively high ductility of the high strength granular bainite (transformed at 773 - 873 K) would appear to be due to the irregular fracture path along a thin uneven rim of ferrite at the prior-austenite boundaries. This leads to a decrease in the ease of brittle intergranular crack propagation and therefore the amount of plastic deformation prior to fracture is increased.

Although the results described here were obtained on smooth bar creep specimens it is considered that they could also have direct relevance to the influence of microstructure upon the ease of propagation of a single large crack at high temperatures. This is specially so, since the work was carried out at relatively high stresses, which may approximate the situation ahead of the tip of a large propagating crack.

CONCLUSIONS

1. The strong effect of the microstructure upon the fracture ductility of 1/2 Cr Mo V steel is brought about by both the variation of strength level with heat treatment condition and the variation of the morphology of the intergranular crack path.

2. Maximum ductility but relatively low creep strength was obtained by transforming at temperatures high in the ferrite range. High creep strength and relatively good ductilities were obtained by transforming at temperatures high in the bainite range and subsequently tempering.

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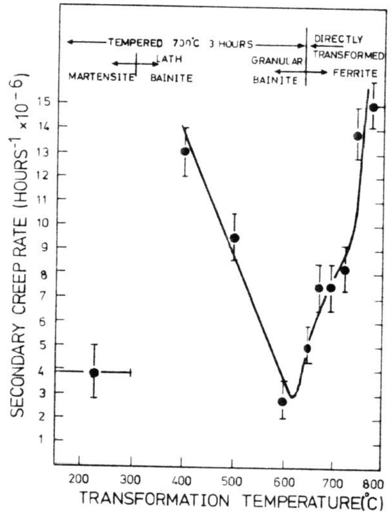
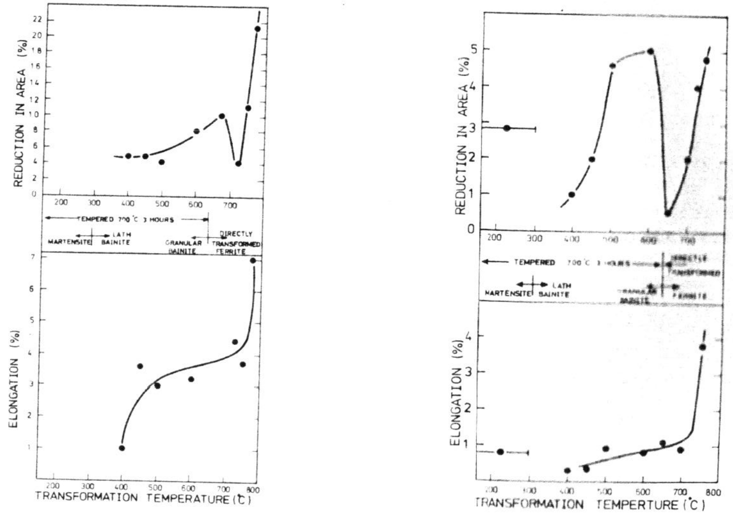
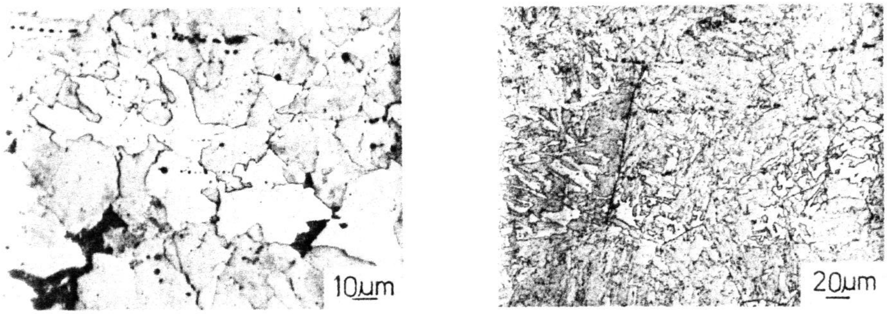


Figure 1 Secondary Creep Rates as a Function of Heat Treatment, 278 MPa



(a) (b)

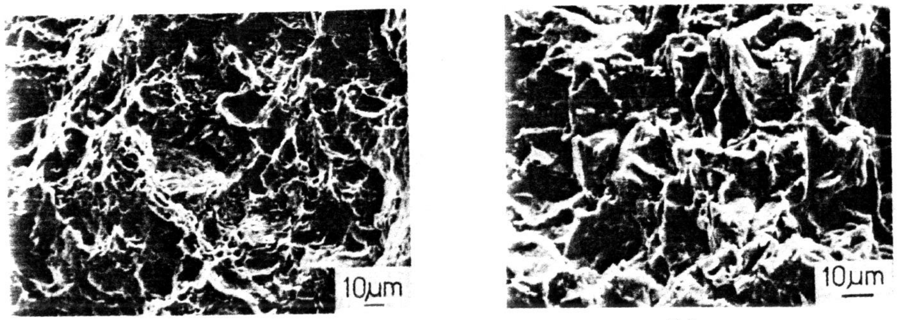
Figure 2 Creep Ductility as a Function of Heat Treatment, (a) 278 MPa, (b) 340 MPa



(a) (b)

Figure 3 Optical Micrographs of Intergranular Cavitation

(a) Ferrite Transformed at 973 K
(b) Tempered Bainite Transformed at 723 K

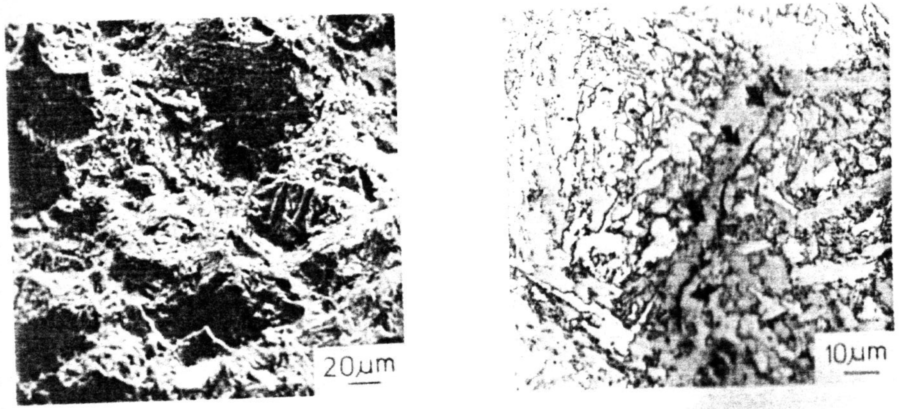


(a) (b)

Figure 4 Fracture Surfaces of Directly Transformed Ferrite Crept at 340 MPa. Specimens Isothermally Transformed at:

(a) 1023 K
(b) 973 K

Figure 5 Cavities on a Prior-Austenite Grain Boundary in the Fracture Surface of a Tempered Bainite Specimen



(a) (b)

Figure 6 Creep Fracture of Tempered Granular Bainite Initially Transfromed at 873 K

(a) Fracture Surface
(b) Optical Micrograph Showing Crack Path. Arrows Show Position of Prior-Austenite Grain Boundary

Table 1

Microstructure	Tempered Martensite	Tempered Bainite	Directly Transformed Ferrite
Transformation Temperature, K	Water Quenched	673, 723, 773, 873	923, 948, 973, 998, 1023, 1048
VPN	315	290, 290, 280, 325	360, 365, 350, 280, 260, 255