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THE INFLUENCE OF A FINE-SCALE PRECIPITATE DISPERSION ON CLEAVAGE FRACTURE IN LOW ALLOY STEELS

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INTRODUCTION

Almost all commercial low alloy steels produced for engineering purposes rely on a fine-scale precipitate dispersion for some of their strength. The influence of a fine-scale precipitate dispersion on strength is well documented [1], but in contrast the influence on the fracture process has received far less attention. There has been considerable interest in the role of coarse precipitates, particularly grain boundary carbides, which can crack during deformation and initiate cleavage failure [2,3], but the influence of the fine-scale precipitate dispersion is less well understood. This is partly due to the difficulty of relating sub-optical precipitate distributions to the observed fracture process.

An interesting approach to this metallographic problem has been adopted by Lindborg and Averbach [4], who fractured 5-10 μm thick foils of Fe-0.3C by scratching with a sharp point on a hard surface. Areas adjacent to the fracture surface were thin enough for transmission electron microscopy. Although admitting that their results were not necessarily representative of bulk specimens they documented the most frequently observed cleavage planes in ferrite and martensite, and some of these results have since been confirmed by other work [5]. Averbach [6] later stated that the cleavage plane in martensite was influenced by the habit plane of carbides precipitated during tempering. Tempering at 260°C gave ε-carbide on $\{100\}_\alpha$ planes and considerable cleavage was also noted on these planes. Tempering at 340°C and 480°C gave cementite on $\{110\}_\alpha$ planes which then appeared as an additional cleavage plane. More recently, Musiol and Brook [7] have suggested that the Fe_2Mo Laves phase precipitates as a platelet on $\{100\}_\alpha$ planes in Fe-20Co-5Mo and stabilises cleavage in this alloy, and they cite evidence from fracture surface observations.

The present work was aimed at developing a transmission electron microscopy technique to investigate the microstructural features of fracture, particularly in low alloy ferritic steels containing aligned carbide dispersions resulting from periodic precipitation at the austenite/ferrite interface during austenite decomposition (interphase precipitation) [8,9]. This characteristic non-random carbide dispersion has been observed in a number of commercial steels; for example, hot rolled and normalised, or controlled rolled low carbon steel microalloyed with Nb or V [10,11]. In this case the precipitation of Nb(C,N) or V(C,N) is known to give a valuable strength increment [10], but it has also been shown to reduce toughness [12]. It is uncertain whether this reduction results simply from the increase in yield stress [13], or whether it could be due to the precipitation of Nb(C,N) or V(C,N) platelets on $\{100\}_\alpha$. Interphase precipitation gives

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only one of the three possible variants of carbide habit. This has been found to be either the one closest to the plane of the austenite/ferrite interface [8,9] leading to parallel sheets of platelet precipitates on $\{100\}_{\alpha}$ [14] as illustrated in Figure 1a, or at an angle to the interface which is parallel to $\{110\}_{\alpha}$ [15] as in Figure 1b.

EXPERIMENTAL PROCEDURE

The steels used were high-purity melts of composition (wt.%) Fe-3.9Mo-0.16C, Fe-0.54V-0.14C, and Fe-1.01V-0.25C. Specimens were austenitised at 1150°C and then either isothermally transformed in the range 700-800°C or furnace cooled. Specimens were fractured either by tensile testing or notched impact at -196°C. A scanning electron microscope was used to examine the fracture surfaces both unetched and lightly etched in 3% nital. Transmission electron microscope specimens were prepared from some fracture surfaces by the following technique. Iron was electroplated onto the fracture surface from an acidic aqueous solution of $500gFeCl_2$ + $5gMnCl_2$ per litre at 93°C and a current density of $2000amp/m^2$, to give a deposit several mm thick. This was then annealed at 380°C and 0.25mm thick slices were cut from the specimen to include the fracture surface. These slices were mechanically ground to 0.05mm and then discs of 3mm diameter cut out such that the interface between plating and the original specimen was across the centre of the disc. The resulting discs were jet polished in 5% perchloric acid in 2-butoxyethanol at -20°C and 85V until a small perforation was obtained, this perforation being enlarged by ion beam thinning until it crossed the fracture surface. It was then possible to examine the fracture surface section by transmission electron microscopy.

RESULTS

Fe-3.9Mo-0.16C

Interphase precipitation in Fe-Mo-C alloys is much coarser than in Fe-V-C alloys, and at high transformation temperatures the precipitates can be resolved optically [16,17]. The M_6C precipitates formed at > 750°C have a globular morphology and no ferrite habit plane, making this a useful alloy for examining the sole effect of sheets of precipitates on the fracture process. Specimens were isothermally transformed at 800°C prior to fracture at -196°C Figure 2a shows a section through a nickel plated fracture surface and it is clear that the cleavage crack is not parallel to the precipitate sheets: this was always observed. Figure 2b is of a similar specimen broken by impact and the scanning electron micrograph shows traces of some precipitate sheets on the etched fracture surface. This is consistent with the crack not propagating parallel to the precipitate sheets.

Fe-0.54V-0.14C

Furnace cooling from the austenitising temperature gave an aligned dispersion of vanadium carbide which could be resolved in the scanning electron microscope. Figure 3a is an etched fracture surface from a notched impact fracture at -196°C. The traces of the precipitate sheets can be resolved on two different planes in the same grain, showing clearly that the cleavage crack path is not parallel to the precipitate sheets.

Fe-1.01V-0.25C

Figure 3b shows the same result as for Fe-0.54V-0.14C when a furnace cooled specimen was broken by notched impact. Thin foils were made from a specimen which had been isothermally transformed at 800°C and broken in tension at -196°C. Transformation at 800°C gives a finer precipitate dispersion than furnace cooling. Figure 4a shows the bright field image of the plated fracture surface. The straight cleavage facet can be clearly seen, together with rows of precipitates making an angle with the fracture plane. Figure 4b is a high resolution dark field image of the same area using a (111) carbide reflection and resolves the discrete nature of the precipitation within the sheets. The apparently continuous contrast of the precipitate sheets in the bright field image suggested dislocations tangled with the sheets [18] from the deformation induced by the cleavage crack. Transferring the traces of the cleavage facet and the precipitate sheets to the diffraction pattern, Figure 4c,d, shows these to be consistent with $\{100\}_{\alpha}$ cleavage and $\{110\}_{\alpha}$ precipitation plane. This is the cleavage plane expected in b.c.c. ferrite fractured at -196°C and the precipitate dispersion is consistent with Figure 1b. Further foils were made from a specimen isothermally transformed at 725°C and broken in tension at -196°C. Transformation at 725°C results in a finer precipitate dispersion than that from transformation at 800°C. Figure 5a shows a high resolution dark field image using a carbide reflection, and Figure 5b is a schematic representation of the same area. The crystallography of the crack path in Figure 5 is unclear because it proved too difficult to obtain accurate electron diffraction data from the specimen.

DISCUSSION

Examination of the fracture process in Fe-Mo-C alloys by conventional techniques shows that cleavage cracks are apparently not induced to follow the non-random interphase precipitate dispersion simply because the carbides are in sheets, i.e. the presence of coarse sheets of precipitates has no effect on the ferrite cleavage plane at -196°C. The same result is obtained for coarse dispersions in Fe-V-C alloys examined both by scanning and by transmission electron microscopy.

Possible evidence of cleavage fracture both parallel and perpendicular to the sheets was found in a specimen transformed at 725°C. In the absence of clear crystallographic evidence it is not possible to distinguish whether this corresponds to the situation in Figure 1a, with cleavage on $\{100\}_{\alpha}$, or to that in Figure 1b, with cleavage on $\{110\}_{\alpha}$. Further detailed work will be required to elucidate this situation.

The major conclusion emerging from this study is that the fracture process in low alloy steels will not be seriously modified by the alignment of precipitates in sheets by the phase transformation, and possibly not even in the extreme case of platelet precipitates lying in the sheets which are themselves coincident with the ferrite cleavage plane. However, the influence of non-random precipitate dispersions on plastic deformation in ferrite at the tip of cleavage microcracks, probably emanating from fractured grain boundary carbides, must also be examined. It is not known whether this process would be seriously modified by the state of the dispersion, that is, random dispersions resulting from ageing, or non-random dispersions resulting from phase transofrmations. A previous electron microscope examination of deformation in these alloys shows that dislocations interact with the coarser precipitate sheets to produce continuous lamellae of precipitate/dislocation tangles [18]. This process can be observed in the deformed zone adjacent to the cleavage crack in Figure 4a. It is possible that the anisotropic nature of this deformation might make the crack blunting process different to that which would occur in a random precipitate dispersion.

CONCLUSIONS

1. Thin foils adjacent to the fracture path of bulk specimens which are suitable for electron microscopy have been produced by a combination of iron plating and electrolytic polishing/ion beam thinning.
2. The relationship between the cleavage crack path and the fine-scale precipitate dispersion has been examined in three low alloy steels, with particular reference to the aligned precipitate dispersions obtained by interphase precipitation during austenite decomposition.
3. No firm evidence could be found to suggest that the aligned sheets of precipitates influenced the cleavage plane, which was shown to be consistent with $\{100\}_\alpha$ at -196°C for the vanadium steel isothermally transformed at 800°C.
4. Evidence was found of the non-random precipitate dispersion leading to anisotropic deformation in the plastic zone adjacent to a propagating cleavage crack.

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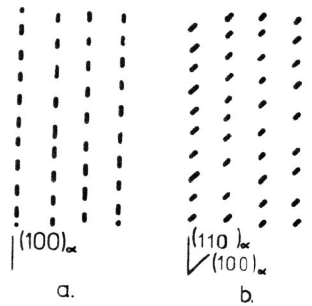
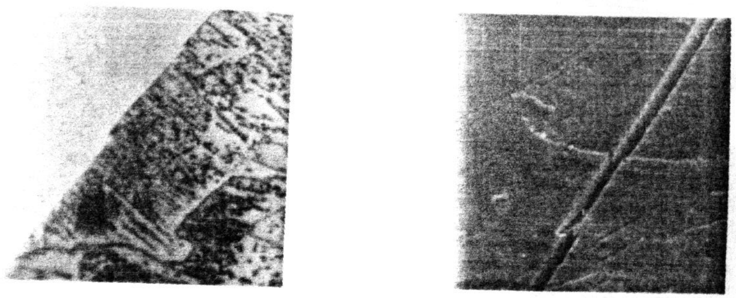
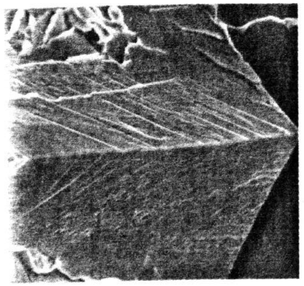


Figure 1 Schematic diagram of vanadium carbide platelets in sheets on (a) $\{100\}_\alpha$ and (b) $\{110\}_\alpha$

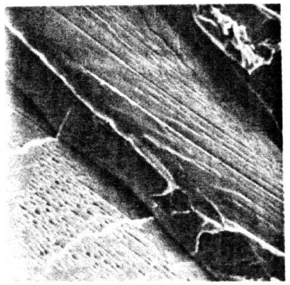


(a) (b)

Figure 2 Fe-3.9Mo-0.16C transformed at 800°C and fractured at -196°C: (a) nickel plated section (X1600), and (b) scanning electron micrograph of etched fracture surface (X1400).



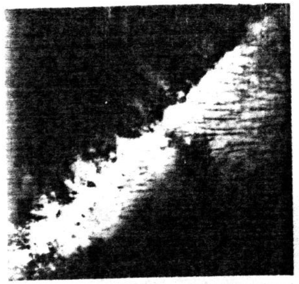
(a)



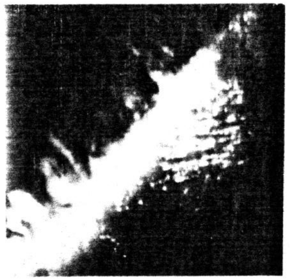
(b)

Figure 3 Scanning electron micrographs of etched fracture surfaces of (a) furnace cooled Fe-0.54V-0.14C (X1200), and (b) furnace cooled Fe-1.01V-0.25C (X1200).

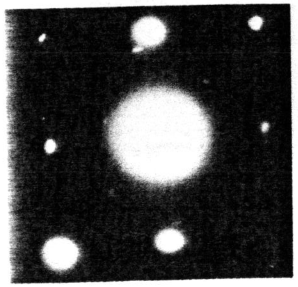
(a)



(b)



(c)



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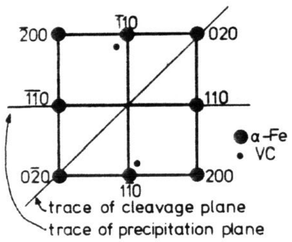
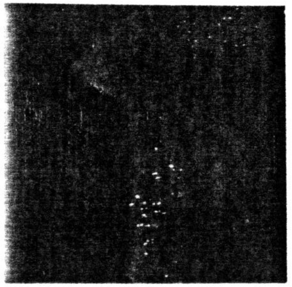
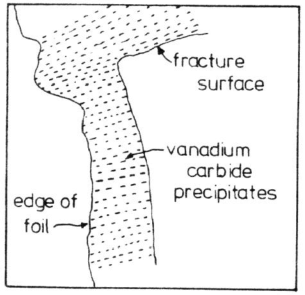


Figure 4 Transmission electron micrographs of iron plated fracture surface of Fe-1.01V-0.25C transformed at 800°C: (a) bright field (X17000), (b) precipitate dark field, (c) diffraction pattern and (d) schematic diagram of indexed diffraction pattern showing the traces of the cleavage and precipitation planes.



(a)



(b)

Figure 5 Transmission electron micrographs of iron plated fracture surface of Fe-1.01V-0125C transformed at 725°C: (a) precipitate dark field (X13000), and (b) schematic diagram of same area.