THE INTER-RELATION OF DEFORMATION-INDUCED MARTENSITE, STRETCH ZONE FORMATION AND FRACTURE TOUGHNESS IN Fe-Ni-C ALLOYS

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INTRODUCTION

The assessment of the fracture toughness of tough, low-strength, materials using the J-integral approach has attracted much attention over the last decade. The approach has gained some success and toughness testing procedures have been specified[1], although there is currently some debate over some of the details of the procedure. One such detail is the precise form of the 'blunting line' construction required to give the best representation of crack-tip blunting prior to the onset of slow, stable, crack growth. This paper discusses the toughness of austenitic and duplex austenite-martensite Fe-Ni-C alloys and its appraisal using the J-integral technique. The behaviour of these alloys is unusual in that there is considerable deviation from the behaviour expected from the 'bulk' properties of the material (such as the tensile properties) when deformation-induced martensite forms at the crack-tip in the austenitic material. This is due to the effect of this martensite on the micromodes of crack extension. When the martensite is induced the stretch zone, expected on the basis of the tensile properties of the material and the blunting line $J = 2\sigma_{flow}\,\Delta a$, does not develop. Instead, a 'brittle' fracture mode occurs directly from the tip of the fatigue precrack. This brittle mode may be intergranular or quasi-cleavage according to the precise composition of the alloy. The toughness is greatly reduced relative to cases where the austenite is fully stable with respect to strains at the crack-tip. This illustrates the importance of local metallurgical factors on toughness.

In this case, a small zone of deformation-induced martensite, ~80μm in extent, controls the fracture toughness. The implications of this to the current testing procedure will be discussed.

TECHNIQUE

The J-integral fracture toughness has initially been determined for an Fe 0.5C 24Ni alloy with a fully (metastable) austenitic microstructure and with a duplex austenite-80% martensite microstructure. The alloy is austenitic at room temperature while the duplex microstructure (consisting of internally-twinned plates of $\{259\}_\gamma$ habit martensite within the remaining austenite) may be developed by refrigeration. The effect on toughness of tempering these microstructures for 3 hours at 300°C was also investigated. In addition the toughness of (metastable) austenitic and duplex microstructures of an Fe0.5C 24Ni 0.25Mn alloy as well as a fully austenitic Fe 36Ni alloy, have also been determined. The latter alloy is stable with respect to the strains at the crack tip.

10mm thick single-edge notched specimens were used for the toughness determination. The specimens were precracked to $\frac{a}{w} \simeq 0.6$, due regard being paid to the precracking requirements laid down in the standards. Load and load-point displacement were measured during the toughness test in three-point bends and values of J were obtained using the expression given in the ASTM standard[1]. The multispecimen heat-tint technique was used to generate resistance curves of J versus crack extension (Δa). In addition, the DCPD method was employed to monitor the initiation of slow, stable, crack growth. Tensile properties and compressive yield strengths were measured on a 5000 Kg Instron machine. Optical metallography was carried out on standard metallographic sections and also on sections taken through the (nickel-plated) fracture surfaces. The sections were electro-polished to eliminate any surface martensite due to mechanical polishing. A Cambridge Stereoscan MKIIA was used for fractography.

RESULTS

The fully austenitic Fe 0.5C 24Ni alloy was found to be surprisingly

lacking in toughness for such a low-strength material (compressive yield strength, σ_γ = 288 MPa) - compare, for instance, the toughness with that of the stable Fe-36Ni alloy quoted later. The value of J at the onset of slow crack growth (J_i) was found to be 25 KNm^{-1}, well below that measured using the blunting-line construction recommended in the ASTM procedure (Fig. 1); the value at maximum load (J_{MAX}) was 197 KNm^{-1}, and the slope of the resistance curve, 167 KNm^{-1}/mm. Although the crack extended slowly and stably across the ligament during the toughness tests, the fracture mode was found to be predominantly intergranular, with a small component (less than 10%) of quasi-cleavage (Fig. 2). Slow crack growth by these "brittle" micromodes is very unusual, macroscopic slow crack growth generally being associated with dimpled rupture. It should be noted that even though the micromodes were brittle, the slope of the resistance curve was appreciable. A section taken normal to the fracture surface showed that deformation-induced martensite formed at the crack tip. Both internally-twinned plate martensite of $\{259\}_\gamma$ habit plane and chevron plates of $\{225\}_\gamma$ habit were induced[2]. This martensite was found to be highly localised to the crack tip, the depth of the martensite beneath the fracture surface being found to be 80μm (Fig. 3).

Fractographic analysis showed that the intergranular fracture occurred at the tip of the fatigue precrack, and the expected stretch zone did not form. In some areas along the precrack front a small region of dimpled rupture was followed by quasi-cleavage. The occurrence of these facets rather than intergranular fracture is promoted in cases where the fatigue crack has stopped part-way through a grain, and on subsequent monotonic loading the crack must therefore propagate through the remainder of the grain before it may follow the grain boundaries. Employing the blunting line construction is therefore inappropriate for this material since it does not describe the physical mechanisms occurring. Also, it would lead to non-conservative estimates of the initiation toughness of this material compared with the value J_i = 25 KNm^{-1} at which slow stable cracking started. It seems that the ASTM procedure needs to be modified in order to account for such eventualities.

In order to elucidate the effect of deformation-induced martensite on toughness, tests were conducted on the austenitic material at 300°C, which is above M_d for the plate martensite (i.e. above the temperature at which plate martensite can be induced by mechanical deformation). The toughness was greatly elevated, the value of J at the onset of growth was now $\sim 10^3$ KNm^{-1}; this was invalid on the plane strain 'validity' criterion. However, since the specimen thickness was the same for both tests at 300°C and the ambient, this serves only to emphasise that the toughness is greatly increased when no martensite is induced at the crack tip. Deformation modes typical of those associated with stretch zone formation were now observed by fractography.

As tested at ambient, the duplex austenite-80% martensite structure also provided extensive slow-crack-growth by a brittle intergranular mode. The toughness was considerably lower (J_i = 10 KNm^{-1}, J_{MAX} = 25 KNm^{-1}), and the slope of the resistance curve shallow, $\frac{dJ}{da}$ = 4 KNm^{-1}/mm (Fig. 4). Matrix strength is known to play an important role in promoting intergranular failure, so the duplex microstructure was tempered at 300°C to reduce the strength level (from σ_y = 1217 MPa down to σ_y = 968 MPa). Slow crack growth now occurred predominantly by dimpled rupture mode (Fig. 5) and the toughness, J_i, was increased to 29 KNm^{-1} (Fig. 6) but no evidence of stretch zones was found. Tempering the fully austenitic structure had no effect on toughness or micromechanisms.

The intergranular crack path on the room-temperature tests was found (by Auger Electron Spectroscopy) to be associated with the segregation of sulphur to the prior-austenite grain boundaries. Consequently, manganese was added to the base composition to eliminate this embrittlement. This Fe 0.5C 24Ni 0.25Mn alloy was, again, tested at ambient temperatures with both austenitic and duplex (austenite-80% martensite) microstructures. In the case of the 100% austenite structure the fracture mode was found to be fully quasi-cleavage (Fig. 7) and slow, stable, crack growth occurred during the toughness tests. No appreciable stretch zone formed at the crack tip, quasi-cleavage occurring directly from the tip of the precrack. The toughness was comparable to the austenitic Fe-Ni-C alloy (J_i = 24 KNm^{-1}, J_{MAX} = 270 KNm^{-1} for σ_y = 304 MPa). Deformation-induced martensite again occurred at the crack tip. The duplex microstructure also failed by quasi-cleavage, with a toughness somewhat greater than the corresponding Mn-free case (σ_y = 1300 MPa, J_i = 17 KNm^{-1}, J_{MAX} = 54 KNm^{-1}).

Finally, the toughness of the fully-austenitic, Fe36Ni alloy was

determined at ambient temperatures. This alloy was of comparable yield strength to the other austenites (σ_y = 336 MPa) but no phase transformation was induced by deformation. The fracture mode was dimpled rupture. In this case, a large stretch zone formed at the crack tip, $\approx$ 1mm in extent. The value of J at the onset of stable growth was found to be $\sim 10^3$ KNm^{-1}, which was invalid on the 'plane-strain' criterion.

CONCLUSIONS

This work has shown that brittle fracture modes - both intergranular or quasi-cleavage - can occur directly at the tip of the fatigue precrack in relatively low strength materials. In these cases, the blunting line construction is inappropriate, and it leads to non-conservative estimates of the toughness at initiation. The reason for this deviation from the expected behaviour is the influence on fracture of that deformation-induced martensite forms local to the crack tip.

Macroscopic slow crack growth has been observed to occur in these alloys by brittle micromodes. No comparable evidence of true slow crack growth by these brittle modes has been reported so far as we are aware. In some cases where these brittle modes occur the slope of the resistance curve, $\frac{dJ}{da}$, was appreciable, so that failure to detect initiation accurately would lead to overestimates of the initiation toughness. This emphasises the importance of detecting slow crack growth in toughness tests, whatever the micromode.

The deformation-induced martensite has a detrimental effect on the toughness of the austenitic Fe 0.5C 24Ni and Fe 0.5C 24Ni 0.25Mn alloys. The detrimental effect appears to arise from the effect of this martensite on the local crack tip conditions in the former case (where the crack path follows the prior-austenite grain boundaries), while in the latter case it arises from the intrinsic brittleness of the martensite (since the crack now goes through the martensite giving a quasi-cleavage mode). The fact that the formation of a zone of martensite, some 80μm in extent, at the tip of the moving crack can control the fracture behaviour of these alloys emphasises the importance of fine-scale metallurgical effects in this region.

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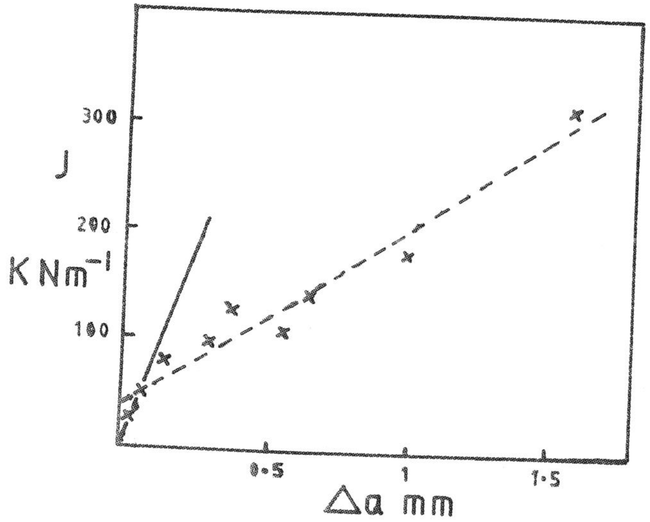

Fig. 1. Crack growth resistance curve, 0.5%C 24%Ni steel (fully austenitic)

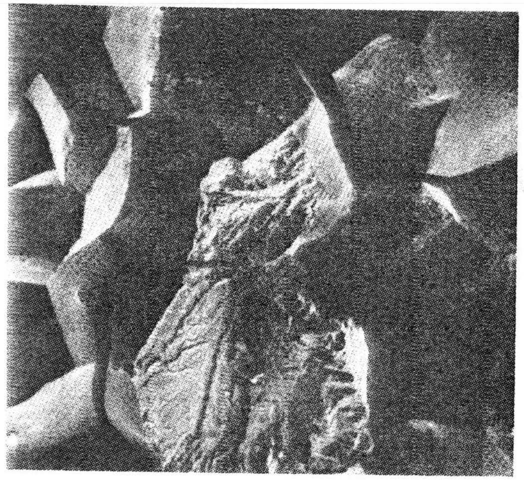

Fig. 2. Fracture appearance 0.5%C 24%Ni steel (fully austenitic) Mag. x 50

Fig. 3. Section normal to fracture plane showing sub-surface martensite 0.5%C 24%Ni steel. Nickel-plated fracture. x 100

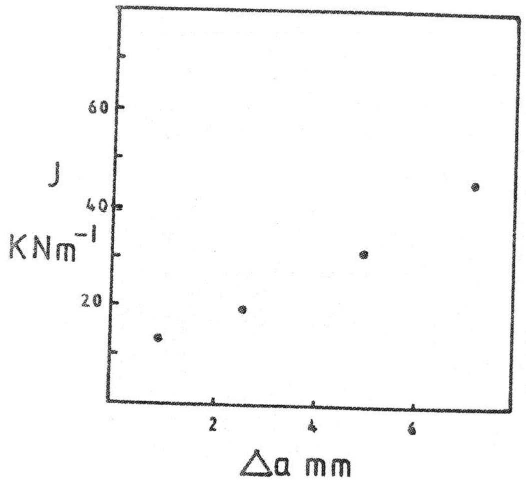

Fig. 4. Crack growth resistance curve 0.5%C 24%Ni steel (austenite-80% martensite)

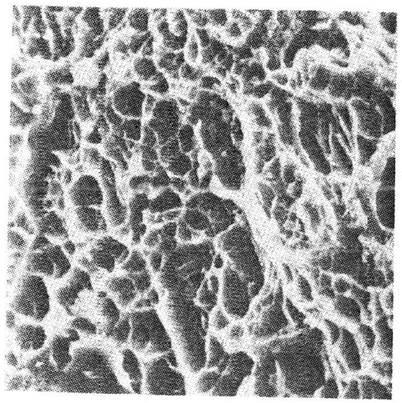

Fig. 5. Fracture appearance 0.5%C 24%Ni steel (austenite-80% martensite, tempered at 300^{o}C)

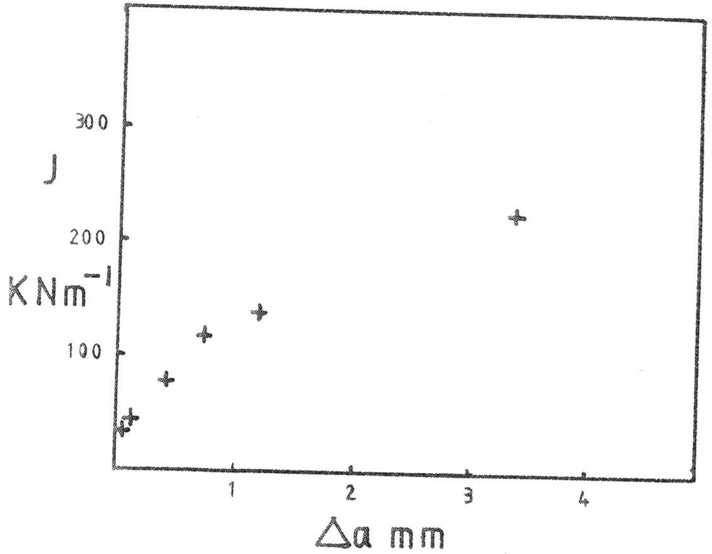

Fig. 6. Crack growth resistance curve 0.5%C 24%Ni steel (austenite-80% martensite, tempered at 300^{o}C)

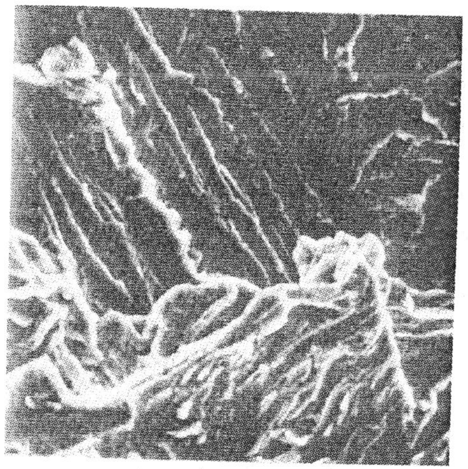

Fig. 7. Fracture appearance 0.5%C 24%Ni 0.2%Mn steel (austenitic) x 800