ENGINEERING

Cleavage Fracture of Polycrystalline Aggregates

BOTH Cottrell[1] and Petch[2] consider that the fracture process is divided into two parts: (1) the formation of a crack nucleus by dislocation coalescence; (2) the growth of this crack until it reaches a critical size when cleavage fracture spreads catastrophically. Using the well-established relationship between the lower yield stress and grain size[3] they show that fracture should just be able to occur at the yield point when:

$$\sigma_y k_y d^{1/2} = \beta\mu\gamma \tag{1}$$

where σ_y is the lower yield stress, k_y is a grain size parameter related to the stress required to unpin a dislocation, β is a mechanical restraint factor, μ is the shear modulus and γ is the effective surface energy of the fractured faces. The effective surface energy, γ, required to produce cleavage fracture in a correctly orientated grain includes a plastic work term and will be larger than the true surface energy. However, it will be smaller than the specific energy G_{1c} (called the plane strain fracture toughness by Irwin[4]) required to propagate the fracture through a large number of randomly orientated grains. Thus the size of the microcracks predicted by the Cottrell model[1] (using the effective surface energy γ) is an order of magnitude smaller than those observed by Hahn et al.[5]. Microcracks can only grow larger than one grain under the yield stress given by equation (1) if there are adjacent grains the orientations of which are favourable. Therefore only a small proportion of the microcracks formed will be able to grow large enough to enable the energy release to be sufficient for fracture propagation through a randomly orientated structure.

If, in a plain tension specimen, the probability that any one grain is favourably orientated with respect to the applied stress is p and the critical crack-length for initiation of a fracture through randomly orientated grains is nd, then the probability for the growth of such a crack at the yield stress is p^n. Thus the Cottrell model[1] will give the criterion for cleavage fracture at the yield stress in plain specimens provided that the number of grains in the specimens is greater than p^{-n}.

In a notched specimen, the stress is non-uniform and only a limited number of grains will be subjected to the yield stress given by equation (1). Near the root of the notch the stresses will be of the order:

$$\sigma = \frac{K}{\sqrt{2\pi r}} \tag{2}$$

where K is the stress intensity[6] and r the radial distance from the root of the notch. Only within a radius of:

$$r = \frac{K^2}{2\pi\sigma_y^2} \tag{3}$$

or

$$r = \left(\frac{Kk_y}{\beta\mu\gamma}\right)^2 \frac{d}{2\pi} \tag{4}$$

will the stress be greater than σ_y. The number of grains in this region is

$$N = \frac{1}{\pi^2}\left(\frac{Kk_y}{\beta\mu\gamma}\right)^4 \tag{5}$$

If the mean of the probability of favourable grain orientated over the highly stressed region is p and the critical crack length is again nd, cleavage fracture at the lower yield stress will occur only if:

$$N = \frac{1}{\pi^2}\left(\frac{Kk_y}{\beta\mu\gamma}\right)^4 > p^{-n} \tag{6}$$

There has not been a great deal of experimental work on the effect of grain size in notched specimens. The results that there are do indicate that cleavage fracture is dependent on grain size. However, there is some doubt[7] that the effect is entirely due to grain size per se and not

to changes in the internal structure brought about by the treatment that results in a finer structure. Thus it is not certain that the inequality (6) is necessarily satisfied and in certain cases the fracture strength need not be dependent on grain size. This may mean that the fracture strength of electro slag welds with their very large grain structure may not be as low as previously thought.

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METALLURGY

Effects of High Pressure on the Strength and Structure of Martensites

FOLLOWING the direct observation of fine-scale twinning (~ 100 Å spacing) in the martensitic structures formed by the rapid quenching of certain steels and other alloys[1], relationships between the incidence of twinning and the high mechanical strength of martensites have been postulated, explicitly in a qualitative strengthening model[2] and implicitly in a quantitative solution-hardening model[3]. In view of the suggested importance of the twinning contribution to strength, an attempt has been made to test its validity by experiments on martensitic transformations in the iron–carbon and iron–chromium systems at high pressure and in the iron–nickel system at atmospheric pressure. The experimental apparatus and techniques used have been described elsewhere[4].

In the Kelly–Nutting model[2], the rapid increase in martensite strength (beyond that attributed to solution hardening) and incidence of twinning with increasing carbon content are considered to arise from the effect of carbon in depressing M_s from above to below a critical temperature range for twinning (approximately 300°–220° C) which is essentially composition independent. It has been demonstrated previously for the iron–carbon system that the temperature at which martensite formation begins on cooling (M_s) is depressed substantially at high pressure[4]. Accordingly, if the twinning model were correct, the strength of martensite formed in all compositions which have their M_s depressed at a given pressure to below the critical temperature range should attain or approach that of fully twinned martensite formed at atmospheric pressure. Microhardness measurements on a series of iron–carbon martensites formed at atmospheric pressure and at 42 kilobars are compared in Fig. 1. At the high pressure, the lowest composition for which M_s is below the critical range is reduced from 0·75 down to 0·4 wt. per cent carbon[4]. However, the expected increase in hardness to high values characteristic of fully twinned martensites in this system (925 VPN) is not observed for the two alloys in this composition region. Instead, the hardness of the 0·4 wt. per cent carbon alloy increases to only 780 VPN and the higher carbon alloy to a similar value, that is, the hardness increase is smaller for the latter alloy. Furthermore, the 0·08 wt. per cent carbon alloy, even though it transforms above the critical range at 42 kilobars, exhibits an increase in hardness as large as that for the 0·4 wt. per cent alloy. The specimens of high-purity iron and the 0·75 and 1·2 wt. per cent carbon alloys show only minor changes; this result suggests that any plastic deformation arising from quasi-hydrostatic conditions in the high-pressure cell does not contribute strongly to the hardness increases in the intermediate alloys.