# MICROSCOPIC INHOMOGENEITY OF PLASTIC STRAIN AND CRACK PROPAGATION IN ALLOYS

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Strength of an alloy can be defined as resistance to plastic deformation *and* the propagation of cracks. While the former aspect is quantitatively understood on the basis of interactions of dislocations with various types of obstacles (which produce solid solution-, work-, precipitation hardening etc. [1]), are we far away from a corresponding understanding of crack propagation? Microstructural aspects have attracted much attention [2]. The emphasis of work in this field was, however, on the decohesion of interfaces such as brittle grain boundaries [3,4] and incoherent interfaces of dispersed particles [5,6,7,8]. It is the purpose of this paper to explore some other microscopic features that could be suitable for a correlation with macroscopic crack propagation. The discussion is restricted to the behaviour of rather ductile alloys in which separation is preceded by large amounts of plastic flow. Large portions of trans- or intercrystalline brittle fracture are excluded.

In fracture mechanics it is usually assumed that plastic strain ahead of crack tip is microscopically homogeneous and that no local volume change occurs together with plastic deformation. Using a Bilby Swinden model for the strain field at a crack tip [9] a quantitative calculation of fracture toughness from tensile test data has been attempted [10]. However, deformation different from the behaviour mentioned above may occur. This is shown schematically in Figure 1.

1) *Transcrystalline inhomogeneity* is favoured if few dislocation sources are activated and if it is difficult for dislocations to leave their slip planes. The latter is the case for low stacking fault energy solid solutions and even much more so for alloys that contain coherent particles that are cut by dislocations [11,12]. Microscopic inhomogeneity of transcrystalline plastic strain has also been reported for some alloys in which crack propagation is preceded by decohesion at interfaces of incoherent particles [13].

2) *Intercrystalline localization* of strain occurs not only in all grain boundaries that are able to slide at elevated temperatures [14] but also - non-thermally activated - if the grain boundary is surrounded by a very soft particle-free zone, while the interior of the grain is highly precipitation hardened [15,16].

3) *A localized strain-induced martensitic transformation* can produce an additional dissipative mechanism and a volume change as a unique feature. If positive, it will relax the stresses at the crack tip [17,18]. Observations on fatigue life of stainless steels indicate, that in metastable austenite cracks may be retarded by this phenomenon [19].

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Before a comparative discussion of these three micromechanical features of crack growth, the ratio of the plastic zone size $r_p$ and microstructural characteristics, such as grain size have to be considered (Figure 2). If the amplitude of stress intensity $\Delta K$ increases in a fatigue experiment so does the size of the plastic zone. If the plastic zone size attains the dimension of particle spacing, grain size, specimen thickness, qualitative changes in the deformation behaviour ahead of the crack tip can be expected. Most important is the ratio of the plastic zone size $r_p$ to the grain size d: $r_p < d$ to $r_p > d$. In the following experimental results on crack growth will be shown, obtained with alloys the microstructure of which had been designed to test the role of these three micromechanical parameters (Table 1).

The effect of transcrystalline inhomogeneity was tested with alloy 1 (Table 1) which could be heat treated to show homogeneity or inhomogeneity of strain at about the same tensile test properties. The result of fatigue crack growth measurement may appear surprising because the inhomogeneously deforming alloy shows a much slower crack velocity as compared to the homogeneously deforming overaged one (Figure 5). These differences are most pronounced at small $\Delta K$-values. At high values of $\Delta K$ and at $K_c$ the alloys become more and more insensitive to inhomogeneity of strain. Based on a multiple slip model [12,20] it can be explained that at low stress intensity and for a high tendency for plastic inhomogeneity the strain pro cycle becomes concentrated on one slip plane only. This permits partial reversibility of strain and thus a reduced crack progress (Figure 3b):

$$\frac{da}{dN} = (n_o - n_R)\ b \qquad (1)$$

$n_o$ is the number of dislocations that slide during a cycle, $n_R$ the number that is reversed. $n_R$ depends directly on the tendency to deform inhomogeneously and on $r_p$.

Grain size dependence of fracture stress is often derived from micromechanical models that assume dislocations concentrated in pile-ups, interacting with hard grain boundaries. It is likely that intercrystalline fracture is induced frequently by the other extreme, namely very soft grain boundary zones as shown in Figures 1 and 4a (alloy 2, table 1).

If the plastic zone size is larger than the grain size: $r_p > d$, the yield stress in the grain boundary zone $\sigma_{y1}$ is much smaller than that inside the grain $\sigma_y$, the soft zones at the grain boundaries are planar with a thickness of $d_1$ - then it can be observed that the strain is concentrated in these zones, the amount of which is proportional to grain size [16]. If decohesion takes place in the soft zone at a critical local strain $\varepsilon_1 = x/d_1$ the fracture toughness of such an alloy should increase with decreasing grain size [15].

$$K_c \propto \sqrt{\frac{\varepsilon_1\ d_1}{d}} \qquad (2)$$

Experiments indicate that such a grain size dependence exists for precipitation hardening aluminium alloys in certain aging conditions. The local fracture strain $\varepsilon_1$ depends on the local particle distribution (grain boundary precipitates) [5,6,7,8]. For low stress intensities with





$r_p < d$, no grain size dependence was found. High amplitude fatigue at $r_p > d$ shows a similar grain size dependence as fracture toughness [21].

Martensitic transformation will have to produce a phase with a larger specific volume to retard crack growth by local stress relaxation. The $M_s$-temperature must lie so much below room-temperature that only the stress concentrated at the crack tip is able to induce transformation. All these conditions are fulfilled by certain austenitic alloys of iron (Table 1, alloy 3). The size of the transformed zone $r_t$ increases with the shear stress $\tau$ and decreases with the difference between test temperature T and martensitic temperature $M_s$:

$$r_t = f\left[\tau,\ (T - M_s)\right] \qquad (3)$$

At a constant test temperature and a stress increasing by an increased $\Delta K$ in a fatigue experiment, it can be expected that $r_t$ passes through 3 ranges: I. $r_t = 0$; II. $0 < r_t < B$; III. $r_t > B$.

In range I no transformation affects crack growth. In range III the size of the transformed zones has reached the specimen dimension, and the crack proceeds into the already transformed material. Only in range II a relaxation can be expected because the transformed zone at the crack tip is surrounded by untransformed material (Figure 5). Fatigue experiments show that if a crack passes from range I into II, the crack growth rate can decrease with increasing stress intensity (Figure 5c), as a direct consequence of the localized phase transformation. At high stress intensities the crack growth rate approaches that of an alloy which has already been completely transformed by cooling.

A summary of characteristic features of the three types of heterogeneous deformation is listed in Tables 2 and 3. The experiments have indicated that crack propagation is influenced by microscopic inhomogeneity and volume change inside the plastic zone. The range of conditions for which effects are to be expected can be specified (Tables 2 and 3). A quantitative treatment must be based on a comparison of the energy dissipated for microscopically homogeneous and the inhomogeneous deformation. This was shown for the case of localized deformation in particle-free zones [15]. The fracture toughness decreases as the plastic strain that precedes cracking is limited to a portion of the material's volume ahead of the crack tip. Hahn and Rosenfield's equation [10] can be regarded as a special case of equation (2) which is valid for microscopically homogeneous strain ahead of a crack.

A quantitative treatment for the retarding effect caused by transcrystalline inhomogeneity for low amplitude fatigue crack growth is more difficult inspite of the qualitative understanding. For homogeneous deformation each half-cycle contributes to crack growth. Increasing heterogeneity induces an increasing portion of reversible strain (Equation (1)), by which crack growth is retarded. There are, however, additional effects which make the situation more complex. This is firstly the partial oxidation of the slip plane which has been exposed under tension, and secondly the fact that grain boundaries act as obstacles to crack growth for the case of inhomogeneous transcrystalline strain [12] (Table 2).

It is the purpose of this paper to show that the assumption of microscopically homogeneous strain or strain which is not associated with a volume change is not correct for many important alloys. It is expected that the recognition of these effects will help to find the optimum microstructure for crack resistant alloys.

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Table 1

| alloy No. | composition in wt % | heat treatment | microstructural features |
|---|---|---|---|
| 1a | Fe + 40 Ni + 6 Al | underaged:<br>10 min 1300 °C ↓ $H_2O$<br>80 h 500 °C ↓ $H_2O$ | coherent particles, inhomogeneous strain,<br>$M_s$ < -200 °C |
| 1b | | overaged:<br>10 min 1300 °C ↓ $H_2O$<br>7,5 h 720 °C ↓ $H_2O$ | incoherent particles, homogeneous strain |
| 2 | Al + 4.66 Zn + 1.20 Mg + 0.25 Mn + 0.24 Fe | homogenized, rolled, recrystallized, and aged:<br>1 h 200 °C | prec. hardened, particles-free zones, $d_1 \approx$ nm with grain boundary precipitates |
| 3 | Fe + 29 Ni + 7 Al | 0,5 h 1300 °C<br>water quenched to + 20 °C | homog. austenite<br>$M_s$ = -28 °C<br>$\frac{\Delta V}{V} \approx 4$ % |

Table 2 Effects of Transcrystalline Inhomogeneity

| | homogeneous | inhomogeneous |
|---|---|---|
| $r_p > d$ | no effect | no effect |
| $r_p < d$ | no effect | crack retardation |
| (ductile) grain boundaries | no effect | crack retardation |

Table 3 Effects of Different Types of Inhomogeneity

| | | transcrystalline (Fig. 3) | intercrystalline (Fig. 4) | martensitic (Fig. 5) |
|---|---|---|---|---|
| 1 | $r_p < d$ | retardation | no effect | -- |
| 2 | $r_p > d$ | no effect | grain size dependence | -- |
| 3 | $0 < r_t < B$ | -- | -- | retardation |

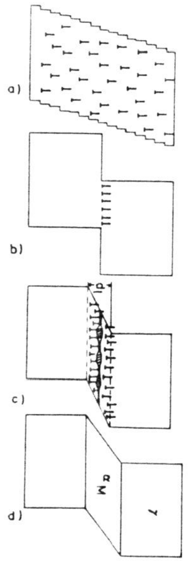



Figure 1 Inhomogeneity of Strain, Schematic:

a) Homogeneous
b) Localized Transcrystalline
c) Localized in Particle-Free Zone at Grain Boundary
d) Martensitic Phase Transformation

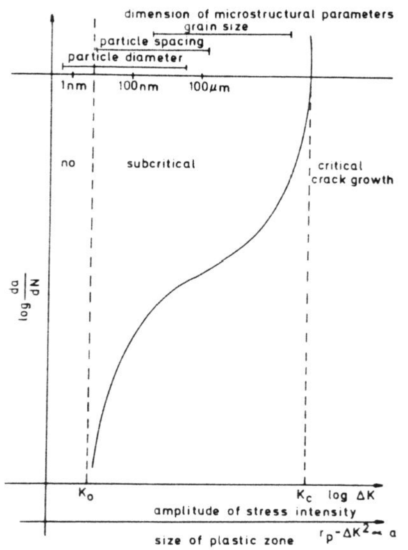
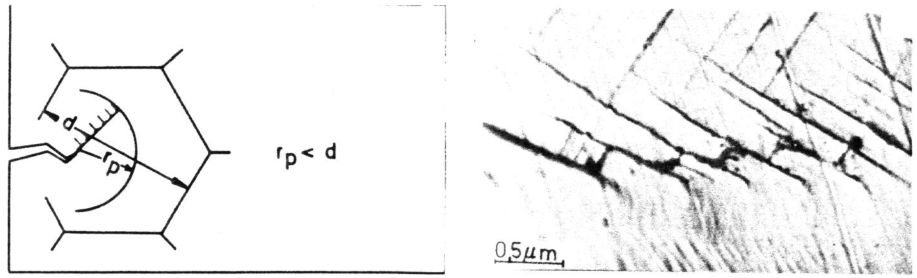


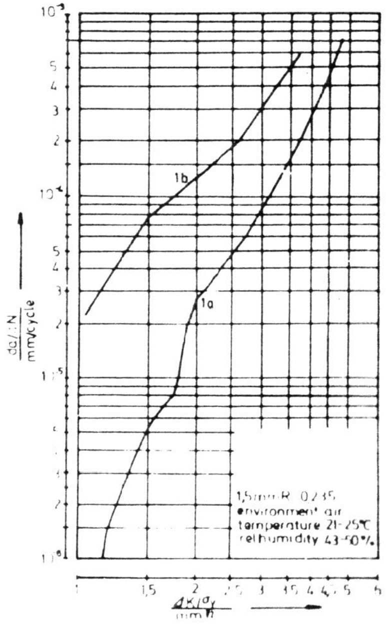

Figure 2 Fatigue Crack Growth Curve Indicating the Varying Ratios of Plastic Zone Size and Microstructural Dimensions with ΔK







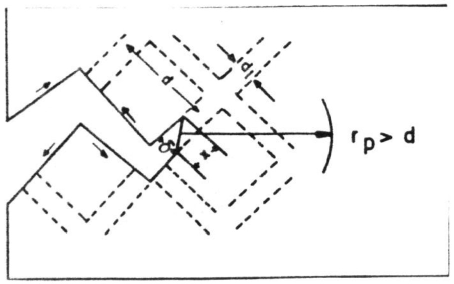

(a) (b)



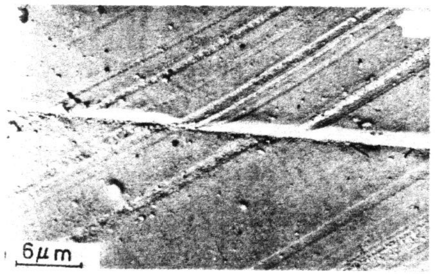

(c)

Figure 3 Effect of Transcrystalline Inhomogeneity on Crack Growth

a) $r_p < d$, Deformation Concentrated on One Slip Plane
b) Fatigue Crack Propagating Along {111} Planes in Alloy 1
c) Fatigue Crack Growth in Alloy 1a and 1b




(a)

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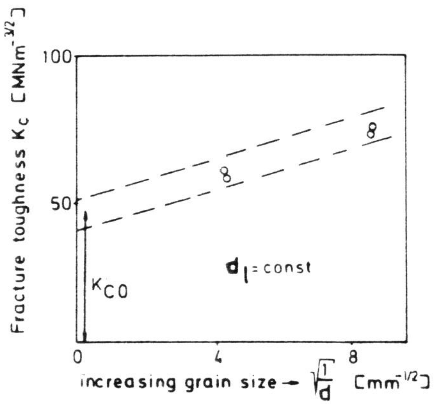


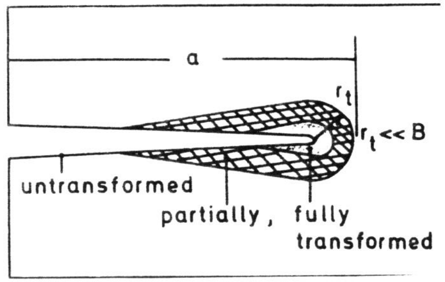

(c)

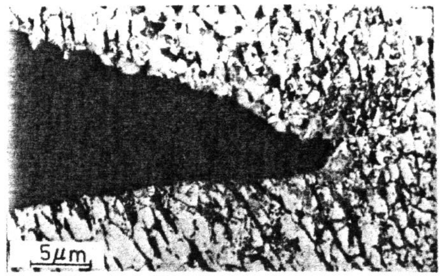
Figure 4 Effect of Localized Intercrystalline Deformation on Fracture Toughness

a) Model, Showing Planar Soft Zones in Which Deformation and Decohesion Occur
b) TEM Replica Showing Highly Localized Sliding at Grain Boundary (Scratched Surface)
c) Grain Size Dependence of $K_c$ for Alloy 2 in Age-Hardened Condition





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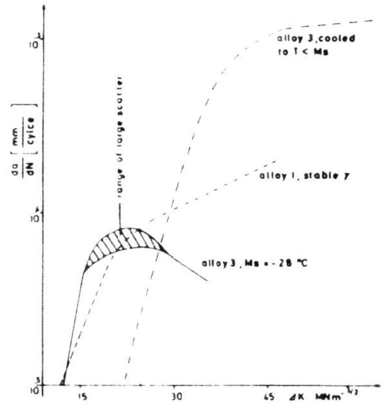


(c)

Figure 5 Effect of Localized Martensitic Deformation on Crack Propagation

a) Model, Showing the Extend of the Transformed Zone
b) Light Micrograph, Fatigue Crack Tip in Alloy 3 with Transformed Zone
c) Fatigue Crack Growth of a Transforming Alloy Compared with Stable Austenite Alloy 1 and the Transformed Alloy 3