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Alloy design for fracture resistance

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Abstract. Several methods for improving the strength of metallic materials are available and correlations between strength and various microstructural features have been established. The purpose of this paper is to review parallel developments favouring improved fracture resistance. Resistance to fracture in monotonic loading, cyclic loading and when fracture is environment-aided have been considered in steels, aluminium alloys and anisotropic materials. Finally, the question of optimising alloy behaviour is discussed.

Keywords. Fracture resistance; microstructural design; cleavage fracture; ductile fracture; fracture toughness; fatigue fracture; fatigue crack growth rates; low cycle fatigue; stress corrosion cracking; optimisation.

1. Introduction

Materials engineering came into being when it was realised that it is possible to control the properties of solid materials by controlling their internal microstructures. Of considerable technological significance is the work which brought out the role played by a variety of microstructural features such as grain boundaries and second-phase particles in arresting the motion of dislocations and thereby contributing to substantial strengthening of solids. This work dominated the concern of the materials scientist during the two decades following the discovery of dislocations and the advent of the transmission electron microscope. During the seventies, attention was turned to a parallel problem namely improvement in resistance of materials to fracture. In the same way that stopping *dislocations* was the concern in strengthening materials against deformation processes, the materials engineer's interest is currently focussed on means to arrest the spreading of *cracks* and other such flaws in order to render the materials more fracture-resistant. In this endeavour, as before, microstructural control is an available and important aid. It is the purpose of this paper to indicate how this is so for metals and alloys.

It has been seen that when a metal is strengthened against one particular mode of deformation, sometimes there is consequential impairment of its resistance to another type of deformation. A classical example is that of refinement of grain size, which enhances resistance against dislocation glide. but causes the metal to be more prone to grain-boundary sliding and diffusion creep. It has become necessary now also to examine what fracture processes arise due to particular means of alloy strengthening. Further, a variety of fracture processes occur beeause of loading and environmental conditions. It is necessary therefore to examine whether any deleterious effects fracture modes occur when microstructures are modified to develop resistance against one particular type of failure. This aspect will also be dealt with in this paper.

2. Fracture resistance in monotonic loading

It is well established that of metallic materials can be strengthened by the following means (Bhat & Arunachalam 1981): (a) solute addition, (b) refinement of grain size, (c) incorporation of second phase particles, (d) dislocations, (e) phase transformation and (f) texture control.

We shall consider the effects of these strengthening processes on the fracture tendencies of metals and alloys with the help of appropriate examples. In order to do this, it would have been useful to deduce expressions correlating parameters reflecting fracture resistance and other material mechanical properties whose relationships to microstructural features are clearly established. Unfortunately, no such clear-cut expressions based on rigorous treatment of the fracture process are available. Even so we shall present below a few correlations which, though incomplete, do contain clues useful for the present discussion.

The ductile-to-brittle transition temperature has been used as a measure of the fracture toughness of the material especially prior to the adoption of the mechanics approach and the development of the K_{Ic} concept. Using a combination of a dislocation mechanism for crack nucleation, the Griffith criterion and the Hall-Petch analysis for grain size dependence of strength, Smallman (1963) has deduced the following ‘ ductile-to-brittle transition equation ’:

$$(\sigma_i d^{1/2} + k_y)\, k_y = \psi\, \gamma\, \mu, \tag{1}$$

where σ_i is the friction stress obtainable from the Hall-Petch intercept, d the polycrystal grain size, k_y the Hall-Petch slope, ψ, a constant dependent on the stress state, μ the shear modulus and γ the fracture surface energy. Features that raise the left hand side of the above equation enhance the impact transition temperature (*i.e.*, brittleness) and *vice versa*.

An important relation for fracture stress (σ_f) as has been given by Cottrell (1957) is applicable when cleavage fracture occurs and is:

$$\sigma_f \approx \frac{4\,\mu\,\gamma_w}{k_y}\, d^{-1/2}, \tag{2}$$

where γ_w is the plastic work done around a crack as it moves through the crystal.

Fracture toughness, now clearly understood as the resistance of a material to crack propagation in structures containing defects of varying size and geometry, is represented in linear elastic fracture mechanics by the parameter K_{Ic}. K_{Ic} is the critical stress intensity factor at which fast fracture occurs and is determined under plane-strain conditions and mode I (crack opening mode) of crack propagation. Hahn & Rosenfield (1968) have related K_{Ic} to tensile properties, which is another useful correlation given below:

$$K_{Ic} = n\,(2E\,\sigma_y\,\epsilon_f\,/\,3)^{1/2}, \tag{3}$$

where n is the monotonic strain-hardening exponent, E the Young's modulus and ϵ_f the true fracture strain in uniaxial tension.

Another useful correlation is due to Krafft (1964), especially relevant to the ductile fracture situation typified in scanning electron micrographs by the appearance of dimples, namely

$$K_{Ic} \approx En\,(2\pi\,d_T)^{1/2}, \tag{4}$$

where d_T is the 'process zone size' which has been shown to correspond to the diameter of an equiaxed microvoid, which in turn depends on the particle (or inclusion) spacing.

With the help of the above correlations it becomes possible to see generally the influence of many of the microstructural parameters on fracture resistance. These correlations have a bearing on the following discussion on the development of fracture resistance in alloys and is presented in a sequence parallel to the strengthening effects listed above. We shall base many of the observations on steels and aluminium alloys and refer to anisotropic materials such as titanium and magnesium alloys only when the influence of crystallographic texture is considered.

2.1 *Effect of solutes*

The effect of solutes on the fracture toughness of steels has not been as systematically studied as their contribution to flow stress in ferritepearlite steels and austenitic steels (Pickering & Gladman 1963; Dyson & Holmes 1970). A general observation, however, is that substitutional solutes which show appreciable strengthening increase the impact transition temperature. This can be understood in terms of the solute effect in raising σ_i in equation (1). The effect of interstitial solutes, carbon and nitrogen, is even more detrimental to fracture toughness and the impact transition temperature. These interstitials significantly raise σ_i as well as k_y in equation (1) and one of the cardinal principles of the design of tough steels is to have the lowest possible interstitial content. The example of carbon-free ultra-high strength maraging steels which are characterised by superior fracture toughness substantiates this view. Among the various solute additions to iron alloys there is at least one element, namely nickel, whose addition is not detrimental to toughness. The reason for this is not clear but may lie in the effect of nickel on stacking fault energy (SFE) and thus on the propensity to deformation twinning. It is also noteworthy that manganese is a favoured addition when high fracture toughness is desired (*e.g.* rail steels and steels for low temperature applications) possibly because of the beneficial effect of manganese in lowering the transformation temperature and causing refinement of the ferrite grain size and pearlite colony size.

This discussion will not be complete unless we mention the interacting effects of solute elements between themselves and with other defects. Frequently substitutional solutes can form insoluble stable compounds such as TiC and TiN by combining with the interstitial elements. Some degree of precipitation may occur but it is more interesting to note that such second phases often pin down grain boundaries and aid grain refinement (*e.g.* aluminium as a grain refiner in steel) when a marked improvement in toughness occurs. Interactions of solutes with dislocations (line defects) are known to cause strain ageing with considerable strengthening but with attendant loss of ductility. Segregation of metalloid atoms such as antimony, phosphorous and tin to grain boundaries (planar defects) can also result in severe embrittlement. In an important contribution Jokl *et al* (1980) have recently explained this drastic embrittlement on the basis of a new concept that the plastic work (γ_p) is a function of the ideal work of fracture (γ) and a small decrease in γ due to grain boundary segregation results in a large reduction in γ_p. Although grain boundary embrittlement may be solved by the addition of other elements *e.g.* carbon in molybdenum reduces embrittling oxygen coverage of grain boundaries (Arun Kumar 1981), an easy remedy to temper as well as to strain-age embrittlement at present seems to be the avoidance of the concerned temperature range of exposure.

2.2 *Effect of grain size*

Equation (2) given above was the first to demonstrate the inverse dependence of fracture stress on grain size ($\sigma_f \propto d^{-1/2}$) for cleavage fracture. The importance of the celebrated data obtained by Low (1954) lies not only in the verification of equation (2) but also in showing the greater sensitivity of σ_f than σ_y to grain size resulting in intersection of σ_f and σ_y plots as a function of $d^{-1/2}$. The implications of this are as follows: for large grain sizes greater than the intersection grain size, fracture occurs when $\sigma = \sigma_y = \sigma_F$. However at smaller grain sizes, $\sigma_f > \sigma_y$ and *strain hardening* is required before the fracture stress level is reached, the extent of strain hardening required being greater as grain size decreases and the ratio (σ_f/σ_y) increases. On this basis as well as on the basis of equation (1), one can readily understand the depression of ductile-to-brittle transition temperature with decreasing grain size. It may, however, be noted that although in the transition range the grain size has a marked effect, below the transition temperature the grain size has no great effect on fracture toughness.

Schwalbe (1977) has deduced a formula for K_{Ic} in terms of grain size which is

$$K_{Ic} = 3\,\sigma_{cl}\,d^{1/2}\,(\sigma_{cl}/\sigma_y) \tag{5}$$

where σ_{cl} is cleavage fracture stress expressed (Rosenfield *et al* 1977) as $\sigma_{cl} = 343+103\,d^{-1/2}$. d is the ferrite grain size and this formula has been found to fit measured K_{Ic} values in several steels. It is interesting to note that σ_y is influenced by temperature, deformation rate, grain size, alloying and heat treatment whereas σ_{cl} depends mainly on ferrite grain size.

A more detailed analysis of grain size dependence of K_{Ic} is due to Armstrong (1977) who has deduced a Hall-Petch type of relation, which is so well-established in the yield stress dependence of grain size. Emphasizing that both for yielding and fracture the critical action of a stress concentration is involved, Armstrong (1977) has shown that

$$K_{Ic} = \left[C\left(\frac{\pi\, s\, a\, e}{a+s}\right)^{1/2} \right] [\sigma_f + k_y\, d^{-1/2}], \tag{6}$$

where s is a critical plastic zone size at the tip of a crack of size a. For

$$\frac{s}{a} > 1\cdot 0,\ \left(\frac{s\,a\,e}{a+s}\right)^{1/2} \approx s^{1/2}.$$

C, a numerical constant has been found to be $(8/\pi)^{1/2}$ for plane strain deformation. The Hall-Petch like plots of K_{Ic} have now been experimentally obtained for ferrite grain size as well as prior austenite grain size (Stonesifer & Armstrong 1977; Joshi *et al* 1977). The comparison shows that the ferrite grain size effect is the larger of the two influences pointing to the important inference that the strong or weak fracture toughness dependence on grain size matches the corresponding strong or weak effect for the yield stress-grain size correlation. The Hall-petch type plot for K_{Ic} of variously-treated aluminium alloys has also been demonstrated (Armstrong 1977). More data of this type will be invaluable in understanding the significance of the Hall-Petch type correlation for K_{Ic} *versus* $d^{-1/2}$.

There is a general significance deducible from the results obtained for fracture stress, the impact transition temperature and K_{Ic} with refinement of grain size which was succinctly expressed more than 250 years ago by Réaumur (quoted by Armstrong 1977) that the quality of a steel is established by the fineness of its microstructure. This general statement is vindicated by observations of increased fracture toughness with refinement in the scale of a variety of transformation products and serves as an unfailing guideline in the design of tough steels. An outstanding product that has resulted from an awareness of this observation is the high strength low alloy steel (Cohen 1975; Raghavan & Ramaswamy 1981).

2.3 *Effect of second-phase particles*

A microstructure-fracture toughness correlation that has received much attention is has to do with the presence of inclusions in steels. Birkle *et al* (1966) have shown that at constant strength level, K_{Ic} is enhanced by a factor of 1·5 with reduction of sulphur content from 0·049 to 0·008 wt % in an AISI 4340 ultrahigh strength low-alloy steel. A reciprocal type of relationship between toughness and volume fraction of inclusions shown by Baker (1974) for non-metallic inclusions in steel is generally valid for all systems containing a dispersion of second-phase particles. A similar result has been obtained for the fracture toughness of aluminium alloys (Mulheiin & Rosenthal 1971). A measurable parameter that incorporates many of the inclusion characteristics, namely volume fraction, shape and size that contribute to the effect of inclusions on toughness is the total projected length of inclusions per unit area. In a random array of elliptical inclusions with semi-axes p, q, r and a total volume fraction of f, the total inclusion projection (P) on a section normal to the r-axis and in the direction of p is given by:

$$P = 2\,f\,\nu/3\,\pi\,p, \tag{7}$$

where ν is the aspect ratio of the inclusions. It has been shown (Baker 1974) that for various deformed type I and type III MnS systems, a single plot showing monotonic decrease of short transverse toughness with increase in P is obtained despite large differences in volume fraction, aspect ratio and size.

The importance of inclusions and other second-phase particles in metallic materials arises from the fact that ductile fracture initiation occurs at these dispersoids. Equation (4) depicts the general observation that K_{Ic} increases with inclusion spacing. An interesting situation concerns inclusion deformation and fracture behaviour during *processing* of alloys. A typical instance is the work of Baker (1974) who found decreasing toughness with increasing hot deformation of steel. This is explained in terms of decreasing distance between MnS inclusions as a result of working. These inclusions are unbonded to the matrix and during deformation the interface parts and the particles elongate *decreasing* inter-inclusion distance. Consequently the overall toughness is lowered in much the same way as in the effect of increasing volume fraction of inclusions. An opposite trend of results has been obtained by Ashok (1976) in the case of $CuAl_2$ particles in an Al-6 wt % Cu alloy. These particles are strongly bonded to the matrix and during hot rolling, particle fracture rather than interface parting occurs. Considering also that grain elongation takes place, the result is an *increase* in inter-particle distance. The particle redistribution during deformation causes improvement in fracture toughness. A noteworthy feature is that hard, brittle particles should be distributed as homogeneously as possible since inhomogeneities cause small local particle spacings which are also deleterious and result in weak crack paths. There is also the effect of alignment of inclusions which can occur during processing, which is invariably responsible for reduction in toughness in the transverse and especially the short transverse direction. The problem of anisotropy is minimised by cross-rolling or addition of rare earth metals such as cerium (Luyckz *et al* 1970). Similarly calcium is used to globularise oxides and minimise anisotropy in toughness (Hilty & Popp 1969).

At constant volume fraction f of the second-phase particles, the mean particle distance λ and the particle size λ^* are interrelated:

$$\lambda = \lambda^* (2/3f)^{1/2} (1-f). \tag{8}$$

Even so there is evidence to suggest (Schwalbe 1977) that refinement in the size of particles (important for void nucleation) has a greater beneficial effect on K_{Ic} than increasing inter-particle distance (important for void growth). The size effect arises because the larger the particle, the greater the probability that it will contain a flaw, and also the greater the probability that interfacial decohesion may occur due to blockage of slip bands (McEvily 1978). Removal of large-sized inclusions using secondary remelting processes such as electroslag refining or vacuum arc remelting has been shown to result in doubling K_{Ic} at constant strength of AISI 4340 steel (Hebsur *et al* 1980a).

Besides the size, shape, orientation and distribution of second phase particles, their type is also of considerable importance. Sulphide inclusions are more deleterious than carbides while carbides are more harmful than intermetallic precipitates. The superior fracture toughness of maraging steels, compared with ultrahigh strength quenched and tempered low alloy steels, is attributable to the presence in the former steels of a finer distribution of fracture-resistant intermetallic precipitates in comparison with coarser and more brittle carbides in the latter.

In the high strength aluminium alloys, because of the fcc structure of the matrix material, the problem of ductile-to-brittle transition does not exist. There is a general loss of fracture toughness, however, when yield strength is increased by the precipitation reactions due to a decrease in resistance to ductile rupture. The attention therefore is mainly on particle dispersions. These in aluminium alloys can be divided into three categories: (a) coarse insoluble intermetallic particles formed during solidification due to the ineivtable presence of iron and silicon; (b) intermetallic particles formed during solid state precipitation due to added elements namely chromium, manganese or zirconium, and (c) precipitates developing during ageing such as GP zones, transition and equilibrium precipitates. The coarse intermetallics of category (a) do not contribute to strength but because they are brittle, provide preferential crack paths. A reduction in iron and silicon contents results in substantial improvement in fracture toughness (Staley 1978). The particles of category (b) do not also influence strength except indirectly through their retarding effect on recrystallisation during processing. Systematic studies of their effect on fracture toughness are lacking. Limited data available on Al-Zn-Mg-Cu alloys suggest that manganese containing intermetallics (*e.g.* $Al_{20}\ Mn_3\ Cu$) are more detrimental than those containing zirconium (*e.g.* $Zr\ Al_3$) or chromium ($Al_{12}\ Mg_2\ Cr$). The general effect of precipitates (category (c) particles) is to enhance strength with concurrent reduction in fracture toughness. When fracture is of the transgranular precipitate type (those that form during underageing or those that develop during overageing), the precipitate has no distinguishable effect on the combination of strength and toughness except during quenching or ageing treatments, which promote intergranular fracture lower toughness (Staley 1975; Kirman 1971).

2.4 *Effect of dislocations*

In steels, higher dislocation densities can arise due to cold work, quenching strains or strains resulting from low transformation temperatures. Finer initial grain size provides higher work-hardening rates with attendant increase in dislocation densities during straining. The effects of alloying elements on stacking fault energy are also important in the way dislocation densities and work hardening rates are. If the recovery and recrystallisation temperatures are exceeded during working, annealing out of dislocations occurs, the noteworthy aspect being the formation of subgrains in warm-worked alloys.

In general, an increase in dislocation density causes a decrease in tensile ductility and increase in impact transition temperature. It appears that uniformly distributed dislocations are less detrimental to toughness than dislocation arrays (Pickering 1978). Again unlocked dislocations are less harmful to toughness than are dislocations locked by precipitates or solute atmospheres (strain ageing is known to result in loss of ductility and toughness). Systematic work on the effect of subgrains on fracture toughness parameters is not available.

In duralumin-type age-hardenable alloys, the introduction of cold work as an intermediate treatment during quenching and subsequent ageing has been tried. In underaged and peakaged tempers and low levels of cold work, toughness increments and strength decrements have been observed. Higher levels of cold work in naturally-aged tempers have provided superior combinations of strength and toughness while the same in average tempers causes poorer combinations of strength and toughness (Staley 1978).

2.5 *Effect of transformation*

The use of high resolution electron microscopy is undoubtedly a major development in the characterisation of the various microstructural elements and thereby in our understanding of the effect of heat treatment on the fracture toughness behaviour of steels. The transformed products in steels exhibit a bewildering variety and it is only in recent years that a clear picture has emerged although quantitative correlations of fracture toughness parameters and microstructural features are not yet available.

Let us first consider ferrite-pearlite microstructures. It has been observed that in medium-high carbon steels, reduction in ferrite grain size and pearlite colony size improves fracture toughness. Increasing the volume fraction of pearlite by increasing the carbon content, markedly decreases the maximum uniform strain, total strain at fracture, increases the impact transition temperature and lowers the ductile Charpy shelf energy. The morphology of pearlite has complex effects. A decrease in the interlamellar spacing impairs toughness possibly because of its positive effect on strength. However, the simultaneously observed decrease in the pearlite-cementite plate thickness improves impact resistance because thinner carbide plates can bend rather than crack and initiate cleavage fracture (Pickering 1978). Because of these opposing effects, an optimum pearlite interlamellar spacing can be found to result in the highest impact toughness.

On the basis of fairly extensive studies of various ultrahigh strength steels, namely quenched and tempered steels of the AISI 4340 class, maraging steels and TRIP (transformation-induced plasticity) steels (Fe-0.3C-12Ni-9Cr-2Mn) the following conclusions emerge about microstructure-fracture toughness relations (Zackay 1975):

(a) free ferrite, whether present as separate grains or as platelets in upper bainite lowers toughness;
(b) lower bainite and tempered martensite free from lath boundary films of carbides provide high toughness;
(c) the substructure of martensite is important too; dislocated martensite being tougher than twinned martensite;
(d) the presence of retained austenite films around martensite laths substantially improves fracture toughness; and
(e) the production of strain-induced martensite of suitable chemical composition in TRIP steels at an appropriate rate relaxes triaxiality and raises in an overall sense the fracture energy resulting in the highest strength and toughness combinations in ferrous systems.

The effect of higher austenitising temperatures (1200°C instead of the conventional 870°C) employed in the case of AISI 4340 steel provides an extremely interesting situation. A systematic trend of increasing plane-strain fracture toughness (K_{Ic}) and decreasing Charpy V-notch energy is observed as the austenitising temperature is raised while the yield strength remains unaffected. It is important to realise that low austenitising temperatures result in small austenite grain size but a large fraction of brittle undissolved carbides. Higher austenitising temperatures lead to greater dissolution of carbides but cause an increase in austenite grain size. Using these observations, Ritchie & Horn (1978) have explained the apparent paradox by associating a reduction in critical fracture strain with the decrease in Charpy V-notch energy and an increase in characteristic distance for ductile fracture, due to dissolution of void initiating particles, with the increase in sharp-crack fracture toughness.

2.6 *Effect of crystallographic texture*

In anisotropic materials such as titanium alloys, zirconium alloys and magnesium alloys, the fracture behaviour is sensitively dependent on the crystallographic texture. The effect of texture on fracture toughness is complicated and depends upon whether ductility or strength dominates in contributing to the energy consumed during fracture. In commercial titanium sheets, Frederick & Hanna (1970) reported that the fracture toughness is higher in the longitudinal direction than in the transverse direction. This is attributed to the easy $\{10\bar{1}0\}$ $\langle 11\bar{2}0\rangle$ slip in the longitudinal specimens facilitated by the duplex type of texture (ideal and (0002) peaks at transvere direction) present in the sheets. In Ti-6Al-4V sheets (Parkinson 1972) the texture consists of basal poles parallel to the transverse direction. This gives rise to higher transverse strength and hence higher transverse fracture toughness under plane-stress conditions (as in thin sheets). However, in thicker sheets, as the plane-strain conditions are approached, there is a shift to higher longitudinal values. Here ductility dominates the fracture process. Recent work of Bowen (1978) on thick plates of highly textured Ti-6Al-4V showed that high values of fracture toughness are obtained in longitudinal specimens with the direction of crack propagation aligned in the short transverse direction. In general, the K_{Ic} values are found to be higher when the crystallographic deformation mode $\{10\bar{1}0\}$ $\langle 11\bar{2}0\rangle$ or $\{11\bar{2}2\}$ $\langle 11\bar{2}3\rangle$ slip was parallel to the planes of maximum shear stress for plane-strain conditions. The Charpy impact values also showed a similar trend.

In polycrystalline magnesium, the low temperature fracture stress is highly texture-dependent and considerable improvement in fracture stress can be achieved by having strong basal pole intensities parallel to the sheet normal (Wilson 1966). These results indicate that the fracture process in magnesium materials is texture-dependent and further work on these lines is needed on commercial magnesium alloys.

3. Fracture resistance in cyclic loading

In classical terminology, the fatigue ratio (fatigue limit or fatigue strength at high endurances/ultimate tensile strength) has been used as a rough guide to the fatigue strength of materials. Laird (1976) has plotted averages of scatter bands for fatigue ratio of various alloy systems as a function of ultimate tensile strength normalised with respect to Young's modulus. This plot forcefully brings out the existence of a great deal of similarity in the fatigue behaviour of several engineering alloys. It turns out that fcc copper and aluminium alloys are somewhat inferior to steels and titanium alloys. The reason for this has been traced to the occurrence of strain ageing in the latter two systems. The generally lower cyclic strength, as compared with monotonic strength is due to, as has been clearly established now, the formation of persistent slip bands in pure metals and single phase alloys and work-softened bands in polyphase alloys. Microstructural design for fatigue resistance must essentially aim at combating the tendency towards such instabilities. A general observation also is that in moderately strong alloys, decreasing SFE through solute addition and increasing cyclic hardening at the crack tip have been found to maximise plastic work involved in opening the crack tip at each cycle and thus to high fatigue resistance (Laird 1976). In anisotropic materials, the fatigue life and fatigue crack growth rates are widely different in longitudinal and transverse directions. In Ti-6Al-4V alloy tested in the longitudinal direction, the smooth bar fatigue limit is higher (Larson & Zarkades 1976) and the fatigue crack growth rates are smaller (Bowen 1976). The differences are again related to the texture-dependent plastic deformation processes that cause fatigue crack initiation and propagation.

3.1 *Fatigue crack propagation studies*

Fatigue crack propagation (FCP) studies have been carried out extensively in recent times. These studies have assumed great practical significance because they determine the growth potential of subcritical (often preexisting in practice) flaws to critical dimensions to failure. It is usual to plot, on a log-log basis, the variation of FCP rate da/dN as a function of the alternating stress intensity ΔK. A sigmoidal plot is commonly obtained which is divisible into three regimes: Regime I at values of $da/dN \lesssim 5\times10^{-6}$mm/cycle in which the prime interest has been in the threshold value ΔK_{th} below which the crack stops growing; Regime II up to da/dN values of about 5×10^{-4} mm/cycle in which linearity is observed and the expression $da/dN = C(\Delta k)^m$, where C and m are material constants, is applicable (if the initial and final crack lengths at which the crack becomes unstable are known then the safe number of cycles can be estimated by solving for dN and integrating). Regime III at still greater da/dN values corresponds to accelerated growth rates up to final fracture. The microstructural effects can be discussed with reference to these three regimes.

FCP behaviour in region I is strongly sensitive to microstructure. At such low crack growth rates ($<10^{-6}$ mm/cycle *i.e.*, about an interatomic spacing per cycle) it has proved difficult to relate microstructural effects directly to changes in fracture modes because the scale of plasticity approaches the order of the microstructural size. However, efforts in this direction are vigorously continuing (Ritchie 1979; Scarlin 1979; Taira *et al* 1979; Gerberich & Moody 1979). We shall limit ourselves to briefly cataloguing the observed effects.

In low strength ferritic-pearlitic steels, ΔK_{th} values decrease markedly with increasing yield strength. In ultrahigh strength martensitic steels, increasing cyclic yield strength led to significant increase in near-threshold propagation rates and reduction in ΔK_{th}. Thus cylic softening can be regarded as very beneficial to improving region I behaviour.

While refining grain size can be helpful in raising the fatigue limit (especially in low stacking fault energy materials), improved resistance to near-threshold crack propagation has been observed with coarser grain sizes in low strength steels (Masounave & Baflon 1976; Gerberich & Moody 1979; Yokobori 1979) and titanium alloys (Robinson & Beevers 1973). However contrasting behaviour has been observed in the case of high strength steels (Ritchie 1979).

Results of the effect of particles in regime I FCP are also available. In the case of medium strength pearlitic rail steel, decreasing the volume fraction of inclusions led to marked increases in the fatigue limit with however no systematic effect on ΔK_{th}. In strong low-alloy steels, Hebsur *et al* (1980 a,b) have observed substantial lowering of FCP in region I as a consequence of electroslag refining. In aluminium alloys (Knott & Pickard 1977) underaged structures have better near-threshold resistance than peak and overaged structures whereas in tempered martensitic steels, spheroidised structures offer by far the best resistance (Ritchie 1977a).

A particularly striking effect of microstructure has been observed (Suzuki & McEvily 1979) in AISI 1018 mild steel. By suitable heat-treatment procedures, duplex microstructures where the martensitic phase is continuous and encapsulates the ferritic phase were found to increase strength ($\sigma_y = 450$ MPa) and threshold intensity ($\Delta K_{th} = 20$ MN/m$^{-3/2}$) in contrast to the microstructures in which the ferritic phase surrounds the martensite (σ_y) $= 290$ MPa and $\Delta K_{th} = 10$ MN/m$^{3/2}$). This appears to be a promising way of improving the FCP resistance of low carbon steels.

Comparisons at constant strength have been made in an ultrahigh strength steel (Ritchie 1979) between isothermally transformed structures containing an interlath network of retained austenite within a lower bainite-tempered martensite matrix and quenched and tempered fully martensitic structures containing no austenite. The beneficial effect of austenite is thus brought out once again and this may be due to with the interacting environmental effects which are important in region I (see also the following section). In the same steel, impurity-induced embrittlement (temper embrittlement) gave rise to vastly accelerated growth rates and a reduction in ΔK_{th} by almost 30% (Ritchie 1977b).

In regime II, where fatigue fracture often occurs by a transgranular ductile striation mechanism, there are several investigations to suggest that FCP is unaffected by microstructure (Ritchie 1977a), at least in steels. This is perhaps a singular instance where in regard to mechanical behaviour, microstructure plays no part. Although this view has prevailed, there are attempts to investigate this aspect with greater care. A recent contribution by Yokobori (1979) on the basis of dimensional analysis and comparison of mathematical equations suggests that, for the case without mean stress, the following equation is appropriate:

$$da/dN = B(\Delta K/\sqrt{S}\,\sigma_c)^{n^*}, \tag{9}$$

where σ_c and S are constants with dimensions of stress and length respectively. Yokobori (1979) has found that B and $\sqrt{s}\,\sigma_c$ are nearly independent of monotonic yield strength but n^* increases with increasing ferrite grain diameter in a low carbon steel. The plots for different grain size intersect at $\ln \Delta K = \ln \sqrt{s}\,\sigma_c$. The effect of microstructural features, other than grain size on n^*, has not yet been investigated but Yokobori's result is pioneering insofar as microstructual influence in region II FCP in steels is concerned.

In aluminium alloys, experimental evidence is gathering in favour of microstructural influence on FCP in regime II (Hahn & Simon 1972; Raju 1980). Raju (1980) has emphasized on the basis of analysis of *da/dN versus* ΔK as well as *versus* plastic zone size and a strength correla-parameter, that in order to extract the microstructural effect, comparisons have to be made at constant yield strength. A clear and marked influence of microstructure in regime II emerges in the case of alpha-beta titanium alloys. Data have been recently gathered and analysed in terms of interacting environmental effects and alternate microscopic fracture modes (Gerberich & Moody 1979).

In stage III, the fatigue crack growth rate is rapid since the maximum stress intensity approached K_{Ic} and the failure is microstructure-controlled. The fracture features in this stage resemble the static modes such as cleavage, intergranular and fibrous fracture. All the microstructural parameters that affect K_{Ic} will also influence the stage III FCP. For example, the removal of non-metallic inclusions in alloy steels improves the K_{Ic} and also reduces the FCP rates in stage III (Hebsur *et al* 1980 a,b). Similarly in aluminium alloys, the resistance to FCP in stage III is nearly doubled by decreasing the volume fraction of intermetallic particles (Staley 1978). Additional resistance to FCP may be obtained by having a larger dispersoid spacing.

3.2 *Low cycle fatigue*

Cyclic stress-strain response has attracted large-scale response because of its relevance to designing alloys for fatigue resistance. Cyclic plastic strain-controlled low cycle fatigue (LCF) has been systematically studied for this purpose (Rama Rao 1978; Gandhi *et al* 1979) as also because for components containing a notch, plasticity near the notch root experiences a plastic strain controlled condition. In principle any component subjected to a start-stop operation constitutes an instance of LCF. A model for FCP discussed in the foregoing section which incorporates LCF properties has been suggested by Antolovich & Saxena (1975). A distinction between LCF and high cycle fatigue (HCF) is made in terms of transition fatigue life (N_{tr}) below which the elastic-plastic strain situation obtains while above N_{tr}, purely elastic strain controlled fatigue occurs. N_{tr} is structure-sensitive decreasing, with increasing strength in steels (Laird 1976).

On the basis of dislocation substructures developed during LCF, a distinction is made between metals and alloys in terms of planar slip and wavy slip materials. When SFE is low, narrow straight surface slip bands and planar dislocation structures occur. In contrast, high SFE materials develop wavy slip bands and dislocation cell structures. An important observation is that the cyclic stress-strain curve is independent of prior treatment for wavy slip metals (Laird 1976).

The influence of microstructural variations on LCF may be illustrated by the work of Malakondaiah & Rama Rao (1977) on an age-hardenable high-strength aluminium alloy which was subjected to LCF following several thermal and thermomechanical treatments after solutionizing. It was found that the overaged alloy showed the best fatigue resistance and the underaged alloy the least with the other treatments following in between the two. Data indicated a good correlation between the observed variation in fatigue life and the fatigue hardening exponent n' arising from the relationship between the stable stress range $\Delta\sigma$ and the plastic strain range $\Delta\epsilon_p$ namely,

$$\Delta\sigma = k'\Delta\epsilon_p^{n'}$$

where k' is another constant. The interpretation of the observed correlation between n' and fatigue life is the following. For a strong material such as the aluminium alloy RR58, inherently capable of resisting stress concentrations, an enhancement in n' raises material resistance to strain concentrations by bringing about an enhanced degree of strain homogenisation. Thus in n' we have a satisfactory index to characterise the material microstructure in the context of LCF. Tomkins' (1968) life-predictive equations corroborate this idea. The overaged alloy contains coarse, uniformly distributed impenetrable particles which resist disordering by the to-and-fro passage of dislocations and represents the best microstructure for LCF resistance.

Laird (1976) has also pointed out that in precipitation-hardened alloys localised softening can be prevented by choosing either an overaged microstructure of the type described above or large precipitate plates interspersed with smaller more close spaced precipitates.

4. Resistance to environment-aided fracture

Metals and alloys undergo spontaneous fracture when subjected to the conjoint action of a corrosive environment and tensile stress. This is generally referred to as stress corrosion cracking (SCC). Environment-aided fracture can also occur due to hdyrogen embrittlement and this type of fracture is not discussed here. The susceptibility to SCC is represented in terms of time-to-failure under constant load in the presence of an environment, or more recently in terms of a threshold stress intensity for stress corrosion crack propagation, K_{Iscc}, below which SCC does not occur. The susceptibility to SCC depends on a large number of factors which may be electrochemical, metallurgical or mechanical in nature. When materials are strengthened by any of the several microstructural variations described above, the SCC susceptibility is also affected. The considerations involved in the alloy design for SCC resistance are discussed below.

The strengthening achieved by alloying leads to an increased susceptibility to SCC. This could be due to a preferential segregation of alloying elements in the microstructure providing anodic areas for corrosion crack initiation or due to a lowering of SFE which promotes the formation of planar arrays of dislocations. For example, the addition of zinc or aluminium to copper decreases the SFE and also promotes the transgranular stress corrosion crack propagation (Swann & Embury 1965). On the other hand, the addition of nickel to 18 % chromium steel increases SFE and also improves the resistance to SCC. Further, the additions of minor alloying elements like nitrogen and phosphorous to stainless steel decrease its resistance to SCC since they tend to promote the formation of planar dislocation arrays. Thus the SCC susceptibility of solid solution-strengthened materials is related to the dislocation substructure through SFE provided segregations are eliminated.

The SCC susceptibility is also influenced by grain size, the resistance being greater in fine-grained materials (Parkins 1964). The materials strengthened by grain refinement will therefore be resistant to SCC and the time-to-SCC failure is related to the grain size through a Hall-Petch type relation (Sudhaker Nayak 1980).

Dispersed-particle strengthening does not influence the SCC behaviour of materials but some of the precipitation-hardened materials are prone to SCC. The most important example of this type is the Al-Zn-Mg alloy (Kelley & Nicholson 1963). If these alloys are heat-treated in the conventional way, preferential precipitation of $MgZn_2$ occurs along grain boundaries which promotes intergranular SCC. However, a recent study (Poulose *et al* 1974) has shown that the stress corrosion crack velocity is reduced by overaging the alloy to give a large volume fraction of the precipitate at the grain boundary. This is at the expense of some loss in toughness. Another example of intergranular SCC caused by grain boundary precipitation is the sensitised stainless steel. In this case also, resistance to SCC may be obtained by solution treatment or stabilization with titanium or columbium. Thus SCC considerations are important in precipitation-hardened materials.

Strengthening by cold work makes the material prone to SCC since the high dislocation density enhances the rate of corrosion reactions, and the residual stresses aid the stress corrosion crack propagation. From the SCC viewpoint, therefore, this strengthening method is not favourable.

The susceptibility of materials strengthened by transformation depends on the microstructure developed, and high resistance to SCC can be obtained by suitable heat treatments. For example, in the 300-M steel, it is shown (Ritchie *et al* 1978) that high SCC resistance may be obtained by isothermal transformation at 250°C when compared with the conventional oil quench from 870°C. This is interpreted in terms of the beneficial role of retained austenite in increasing the resistance to SCC.

In anisotropic materials, texture significantly affects stress corrosion failure. This arises from the fact that the nucleation of stress corrosion cracks as well as their propagation are partly crystallographic in nature. Recent work on titanium and Ti-6Al-4V sheet materials (Sudhaker Nayak *et al* 1980 a,b) has shown that stress corrosion crack initiation can be avoided by having a texture that provides a close-packed crystallographic plane like $\{10\bar{1}1\}$ or $\{11\bar{2}2\}$ to be parallel to the crack initiating surface. This texture is achieved in cold-rolled titanium and in annealed longitudinal specimens of titanium to some extent. The texture dependence of stress corrosion crack propagation was examined by Fager & Spurr (1968) and Boyer & Spurr (1978). The susceptibility to SCC is highest if the texture is such that the maximum stress is applied to the basal plane which is close to the cracking plane. This occurs in the transverse specimens of titanium alloys with $\{0002\}$ poles at the transverse direction (TD). The resistance to SCC can thus be improved if the texture is such that the basal planes are not stressed. This condition is met in the longitudinal specimens of Ti-6Al-4V with $\{0002\}$ peak at TD and in this case, an unusual crack branching tendency is observed (Larson & Zarkades 1974). As the textures and deformation modes are essentially similar in titanium and zirconium materials, the above considerations hold good for zirconium materials also. A recent study (Adams *et al* 1978) supports this conclusion.

5. Optimisation of mechanical behaviour

The ideal alloy for structural applications at ambient temperatures is one which possesses the best combination of strength, fracture toughness, fatigue resistance and resistance to environmental embrittling phenomena such as stress corrosion. It is instructive to examine whether for a given material it is possible to arrive at a composition and microstructure which results in such optimum mechanical behaviour.

We have seen that the various strengthening mechanisms do not influence fracture resistance even in monotonic loading in an identical fashion. The discussion in § 2 has indicated that frequently fracture toughness is impaired when strength improvement is promoted. The outstanding exception is the case of grain size refinement which favours increased strength as well as fracture toughness and simultaneously lowers impact transition temperature. Recent work on FCP indicates that grain size refinement in a low carbon steel lowers crack propagation rates in region II. However, there are observations pointing to a deleterious effect of finer grain size in FCP in region I and at threshold. Insofar as SCC is concerned, finer grain size once again is a desirable feature. It is not uncommon to encounter conflicting effects when fracture resistance in differing situations is considered. For instance strain ageing which results in locked dislocations impairs fracture resistance in monotonic loading but is regarded as beneficial in cyclic loading. Similarly, alloying to reduce SFE improves fatigue resistance in general by mitigating microstructural instability but is very harmful from the viewpoint of SCC resistance.

In view of the wealth of information available on AISI 4340 type ultrahigh strength quenched and tempered low alloy steels, we are in a position to raise the question in this case whether there exists a microstructural condition which displays optimum response to deformation and the variety of fracture conditions we have dealt with. The composition and the established heat treatment provide ultrahigh levels of strength ($>$ 1250 MPa) in this steel. An innovative procedure to modify the microstructure further is to employ higher austenitising temperatures. Such a treatment causes, besides better dissolution of carbides, retention of austenite in the form of enveloping films around martensite laths. This results in substantial improvement of fracture toughness without detriment to strength. Further, the presence of retained-austenite films has been seen to enhance SCC resistance and ΔK_{th} in FCP which is sensitive to environmental effects. Alas! the higher austenitising temperature and the resulting microstructure markedly lower Charpy V-notch energy, which is an important engineering consideration. Further, the overall fatigue resistance is also adversely affected because regions of soft retained austenite induce strain localisation during cyclic deformation. It appears therefore that on the basis of the present state of the knowledge of alloy behaviour the best one can do is to choose a microstructure appropriate to the specific application but it is simply impossible to aim at optimisation of mechanical behaviour from the viewpoints of strength and fracture resistance under varying loading and environmental conditions. This conclusion is reinforced when we realise that even for steel, for which sub-stantial data are available, we have not brought into consideration the requirements of easy processing and acceptable mechanical behaviour at elevated temperatures (Bhatia 1980). Although it seems daunting to set such optimisation as a goal, undoubtedly the alloy designer must feel stimulated and inspired by the existence of such a challenge.

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