Optimising the fatigue resistance of materials

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Summary

"Fatigue and fracture are complex phenomena which are much affected by design detailing, surface finish and environment. Nonetheless, there is some scope for increasing fatigue resistance by material 'engineering', through correct selection, processing and heat treatment, and this paper presents some information on how this may be achieved in component and structural applications and why, in mechanistic terms, it works."

1. Introduction

Materials selection is now accepted to be an integral part of the design process and, because of the wide range of materials available to the designer/engineer, formal guidelines for materials selection procedures have been drawn up [1] and commercial software packages are available [2]. It is probably true, however, that in many industries more attention is given to selecting materials to resist wear and corrosion than to resist fracture and fatigue. Obviously, this reflects the task that the component or structure has to perform and its dominant mode of failure. Equally, there are a number of industries in which fatigue and fracture are important and proper attention is given towards maximising a design's fatigue resistance, both through materials selection and correct detailing.

Fatigue and fracture are, however, complex phenomena and a materials resistance to service cracking can be quite drastically changed by relatively small differences in heat treatment, surface condition and operating environment. It therefore seems economically viable to pay more general attention to fatigue resistance during design, as it may often be feasible to increase fatigue performance while maintaining other required properties, e.g. wear or corrosion resistance, at the same level.

A problem in this regard is that, as mentioned above, optimisation of the fatigue resistance of engineering products cannot be done solely through materials selection. The nature of the fatigue process requires that consideration be given to design detailing (i.e. to stress concentrations) and material processing/fabrication. This can be illustrated nicely with an example drawn from the automobile industry [3]. In the design of a connecting rod for a medium sized petrol engine, stress analysis had indicated that the small end shoulder was the crack initiation point and had a stress concentration factor, K_t, of 1,3. The connecting rod was made of EN15, a forging steel, and lasted for some 800 000 km. A small design modification reduced K_t to 1,2 and increased the life to 1 600 000 km. It was then possible to change to a cheaper EN8 steel, with a lower fatigue limit, and still achieve a life of better than 1 000 000 km. Clearly, over the production run of the engine, significant cost savings would be achieved. This example highlights a major influence on fatigue performance: that of local surface stress concentrations, which may be either designed in or inadvertently produced during machining, fabrication or service.

A further complicating factor in the optimisation process is that some knowledge of fatigue machanisms is required before a designer can fully appreciate the interplay among manufacture, detailing and alloy composition which affects fatigue life. Equally, the scope for increasing fatigue resistance through judicious material selection is somewhat limited, at least where crack growth dominates the fatigue life of the product.

Nonetheless, fatigue resistance can be increased by material 'engineering' (through correct selection, processing and heat treatment) and this paper presents some information on how this may be achieved in both component and structural applications and why, in mechanistic terms, it works.

2. Fatigue: a two-stage process

To understand why a particular strategy increases fatigue resistance requires a little insight into the nature of fatigue. In metals fatigue, is a two-stage process of crack initiation and growth and, because the micromechanisms of growth are different in these two stages, factors that increase the resistance of metals to crack initiation may, in fact, decrease their resistance to crack growth by causing higher growth rates for a given stress. The problem is simplified in ceramic materials, as intrinsic flaws tend to eliminate the initiation stage. Against that must be set the difficulty of processing ceramics in specific ways to produce a more fatigue or fracture resistant material and the fact that true cyclic fatigue effects are not commonly found in ceramics. Polymers, and composites also show two-stage behaviour, as a 'damage zone' needs to be established before a crack moves through it. In general, predicting the fatigue behaviour of nonmetals is rather diffi-

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cult, as it is much more affected by material and processing variables than in metals. Certain toughening mechanisms used in ceramics and composites can be applied to metals in specific circumstances, and hence will be briefly considered. The main thrust of this paper, however, will be towards engineering metals with particular emphasis on steels and aluminium alloys.

2.1 Crack Initiation

Fatigue crack growth in metals requires local plasticity (in fact, fatigue in any material can only occur in the presence of nonlinear constitutive behaviour which leads to permanent residual strain – this is why fatigue in ceramic materials generally only occurs under compressive loading). Where there are surface discontinuities whose stress concentration increases the nominal stress to a value higher than the yield strength of the metal, it is easy to see why local plasticity and hence fatigue crack growth occurs. On smooth specimens with well defined stress concentrations this process is not immediately apparent. In essence, local plasticity arises from the polycrystalline nature of metals because the shear yield stress within a single grain, favourably orientated at 45° to the load axis, is around $\frac{1}{3}$ of the yield strength measured in a standard tensile test. The higher observed yield strength for the bulk material is related to the difficulty of propagating plastic deformation across grain boundaries into less favourably orientated grains.

This plasticity within a grain is concentrated within discrete slip bands, which either soften more or harden less than the surrounding matrix during load reversals. Hence deformation is concentrated within these bands and high strain amplitudes result. As plastic deformation is generally not fully reversible on a microscopic level, the slip band develops a rough surface profile as shown below. This produces very high local stress concentrations and microcracks are initiated along the edge of the slip band. Initial growth follows the slip band direction as a crack always grows along the direction of maximum plastic strain. Crack growth here is referred to as Stage I and is usually at an angle of 45° to the load axis, either on the surface or within the depth of the material [5].

Once a crack has developed it has its own crack tip plastic zone, whose size is a function of crack length and the square of the applied stress. As the crack grows, this plastic zone increases in size and eventually swamps the local slip band plasticity. At this point the crack turns to grow at 90° to the applied load (or direction of maximum principal stress) because that gives the largest plastic zone and hence growth rate, and is energetically favourable. The crack is now in Stage II and the growth mechanism remains the same throughout the life of the component. This changeover from Stage I to Stage II growth may also be assisted by appropriate orientation of slip systems in grains surrounding that in which the crack initiated, but never persists beyond a couple of grain diameters below the surface.

As the strength of the alloy is increased, this mechanism is modified, in that the grain size decreases and nonmetallic inclusions or second-phase particles become important as sites of local stress concentration. The extent of Stage I growth generally decreases in proportion with the grain size and plasticity is initiated at the inclusions. The details of crack initiation still remain similar to the sequence outlined for low strength alloys.

Considering this mechanism, the reason why the fatigue limit in metals is a variable, but low, percentage of the yield strength becomes clear. Although the single crystal shear yield stress is $\frac{1}{3}$ of the macroscopic polycrystalline yield stress, the fatigue limit is influenced by the strain localisation at slip bands, the amount of localised plastic strain required for crack initiation [5] and the difficulty in propagating slip across a grain boundary. These factors interact to give fatigue limit values for metals typically within the range 0,35-0,6 of the UTS.

2.2 Crack Propagation

Once a crack enters Stage II, it is amenable to treatment by continuum mechanics. Characterisation in terms of the stress intensity factor, K, gives a material constant curve, independent of component geometry, when ΔK is plotted against growth rate, da/dN, on log axes. A typical curve is shown below in Fig. 2.

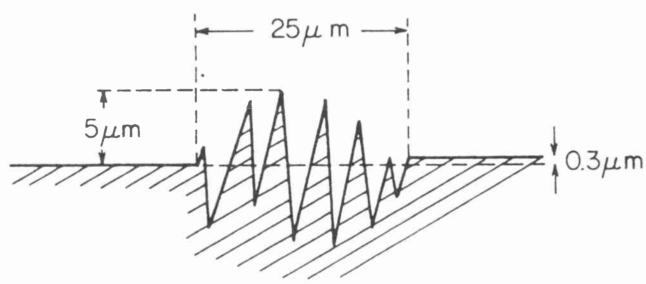


Figure 1 – Approximate surface profile developed at a slip band in copper after 30 000 cycles at a plastic strain range of 0,0025 [4].

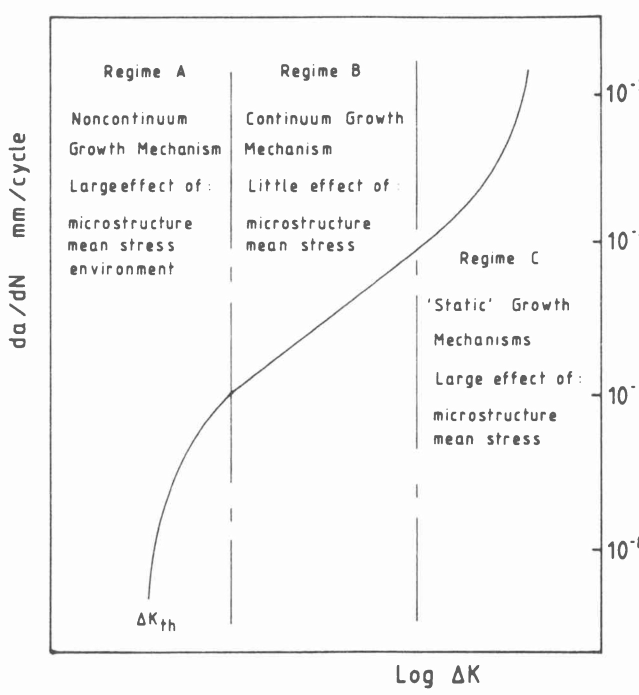


Figure 2 – Fatigue crack growth rate curve showing the factors which influence growth in the three regimes.

The linear regime can be described using a simple equation known as the Paris Law:

$$da/dN = C\Delta K^m$$

where C and m are material constants. Clearly, this equation is easy to integrate and hence a life estimate can be obtained from known defect sizes and applied stress levels. The equation can be extended to include regimes A and C through straightforward modifications, so that integration can be performed from any initial crack size up to the critical size giving fast fracture.

Often, particularly in welded structures, the combination of initial defect size and applied stress is sufficient to give stress intensity values which lie within the linear regime. In terms of fatigue resistance, this regime is troublesome because, as indicated in Fig. 2, growth rates are little affected by microstructural changes and, by extension, changes in alloy composition. This fact is recognised in BS 5400 Part 10 [6] wherein welded joints are classified into nine classes, each of which has a single stress-life curve (obtained by integration of the Paris equation) irrespective of steel alloy actually used. An extension to aluminium alloys has been drafted using the same principles. It is possible, in certain steels, to develop dual-phase microstructures which are very fatigue resistant. An example of this, together with an explanation of why it works so well, will be given later.

Clearly, the threshold for crack growth, ΔK_{th}, could be used to find a 'fatigue limit' stress for cracked components. Generally, however, the threshold stress intensity range is rather low and corresponds to useful levels of stress only when cracks are small (typically < 1 mm). The threshold value and near-threshold growth rates are much affected by alloy composition and heat treatment, however, and there is scope to achieve high threshold levels.

3. Increasing resistance to crack initiation

In essence, as mentioned earlier, any factor which results in a reduced strain localisation in a metal improves resistance to crack initiation. Slip bands, which are the first stage in crack initiation, form either within grains, as outlined above, or can form when dislocations pile-up at an obstacle in the metal. When the stress in the pile-up reaches a critical value, a so-called 'dislocation avalanche' occurs which can lead to a slip band crack in the grain, decohesion along a grain boundary or particle-matrix interface, or cracking of a second phase particle.

Considering this, it becomes clear why increasing the yield strength of an alloy increases its fatigue strength; not only is the resistance to slip band formation within a grain increased, but the smaller grain size which is concomitant with higher strength, at least in steels, reduces the peak stress within a dislocation pile-up and hence reduces the likelihood of an avalanche process (the peak stress is a reflection of the number of dislocations within the pile-up, which in turn reflects the slip distance which is related to grain size). This mechanism of increasing fatigue strength only works if the metal has a dispersion of small inclusions and a good surface finish. As yield strength increases, the notch sensitivity of alloys generally

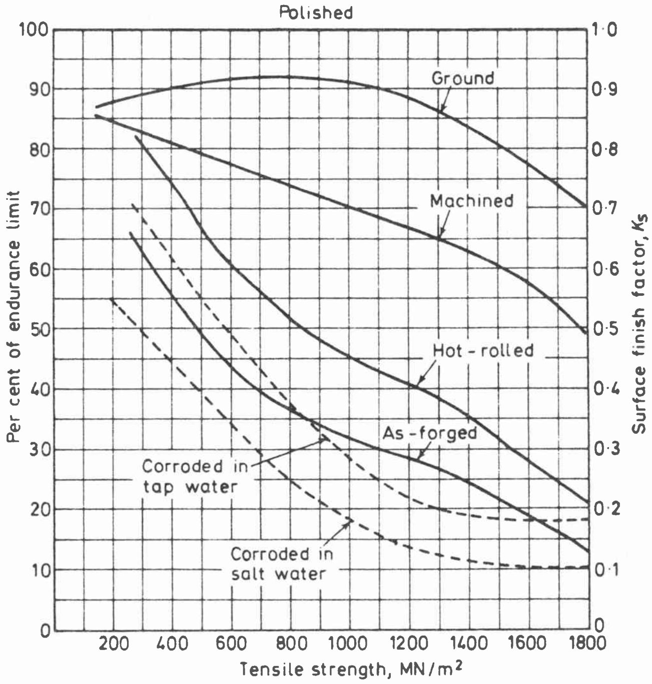


Figure 3 – Effect of surface roughness on fatigue limit of steels [7].

increases, and if large inclusions are present they will readily initiate cracks.

The effect of surface roughness on fatigue limit is shown in Fig. 3 for steels, plotted as a percentage of the endurance limit for a polished specimen against tensile strength [7]. It is apparent that an environment which promotes pitting will negate any beneficial effects on increased tensile strength unless the steel has adequate corrosion resistance.

In the case of steels and aluminium alloys, Al_2O_3 inclusions are crack initiators and an improvement in fatigue strength would come from decreasing the size of such inclusions. Other inclusion types also play a role in crack initiation in aluminium alloys; for example, it has been found that the probability of Al_2CuMg or Al_7Cu_2Fe particles initiating slip band cracks in 2024-T4 decreased rapidly as the size decreased below 7 μm [4]. Small dispersoids may be beneficial to fatigue resistance by limiting the extent of slip bands. $Cr_2Mg_3Al_{18}$ particles approximately 0,5 μm in size improve the fatigue limit in 7075 or 2024 aluminium alloys by $\frac{1}{3}$ [4].

Generally, in precipitation-hardened alloys, thermomechanical processing which gives jagged grain boundaries improves the fatigue resistance [4].

Cast steels with a martensitic structure generally exhibit a better fatigue strength at long lives than obtained from a pearlitic structure, although this may be reversed at short lives ($< 10^3$ cycles) where life is growth dominated. This is again related to the higher strength of the martensite. Considering cast irons, fatigue resistance appears to be more affected by the size of graphite flakes than the hardness of the matrix [8], although the best performance derives from fine graphite and pearlitic matrices. Where the graphite is in spheroidal form an even better fatigue resistance can be obtained, as shown in Fig. 4 [8]. In this

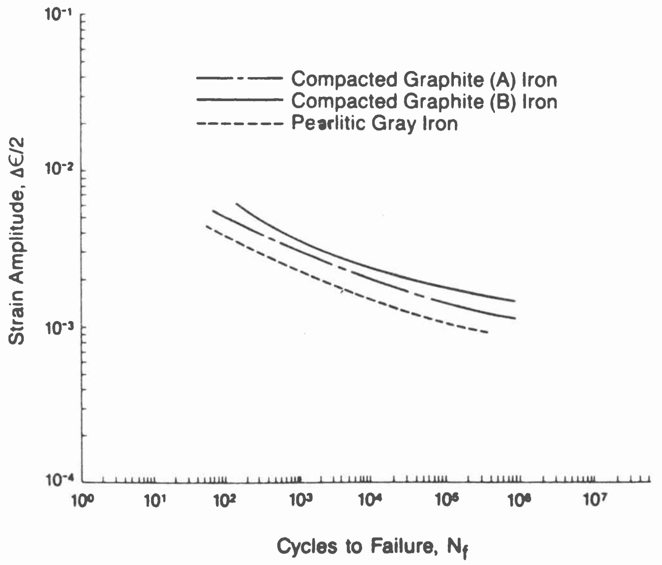


Figure 4 – Strain-life curves for three cast irons [8].

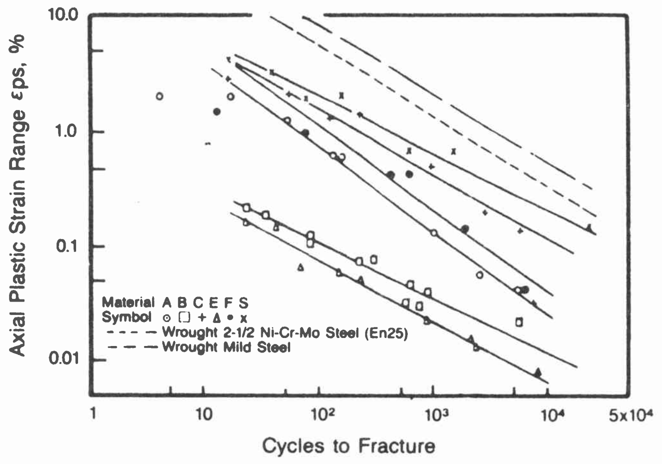


Figure 5 – Strain-life curves for smooth cast iron specimens [8].

Table 1: Chemical analysis and heat treatment of cast irons whose fatigue behaviour is shown in Fig. 5.

Code Letter	C, %	Si, %	P, %	S, %	Mn, %	Ni, %	Cr, %	Mo, %	Al, %	Heat Treatment	Metallographical Report
A	2,6 to 2,21	1,73 to 1,92	0,026 to 0,040	0,005 to 0,011	0,25 to 0,48	34,4 to 35,4	–	–	–	None	Predominantly nodular graphite in an austenitic matrix; small amounts of carbide at the grain boundaries
B	3,62	1,55	0,018	0,066	0,49	–	–	0,90	–	None	Medium/coarse flake-graphite in a matrix containing about equal amounts of ferrite and pearlite
C	0,265	0,38	0,026	0,012	0,64	0,31	1,19	0,29	0,080	Anneal 925 °C, furnace cool, normalize 880 °C, AC; temper 700 °C, AC	Moderately coarse feathery bainite with segregation banding at the grain boundaries
E	3,29	1,10	0,12	0,088	0,69	1,18	–	0,39	–	6 h at 500 °C	Medium flake-graphite in a completely pearlitic matrix containing very small amounts of phosphide eutectic
F	3,5	2,64	–	0,019	0,36	1,26	–	–	–	4 h at 950 °C, AC 2 h at 625 °C, AC	Well formed nodular graphite in a pearlitic matrix with small amounts of ferrite associated with the graphite
S	0,215	0,37	0,024	0,013	0,70	0,16	0,14	0,06	0,052	Anneal 925 °C; furnace cool	Fine lamellar pearlite in a ferritic matrix

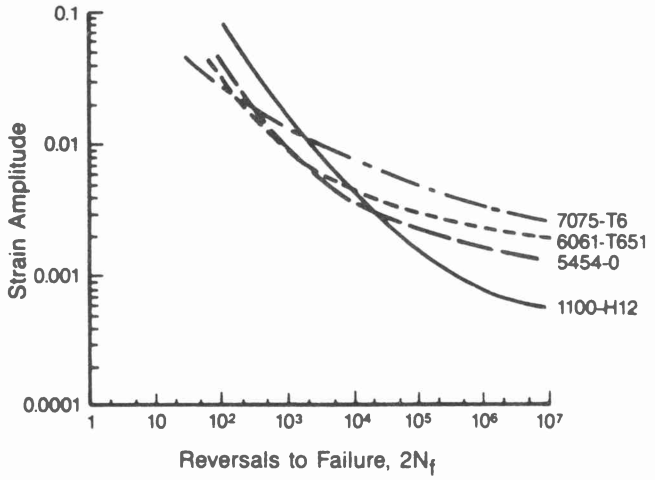


Figure 6 – Fatigue behaviour of several aluminium alloys [8].

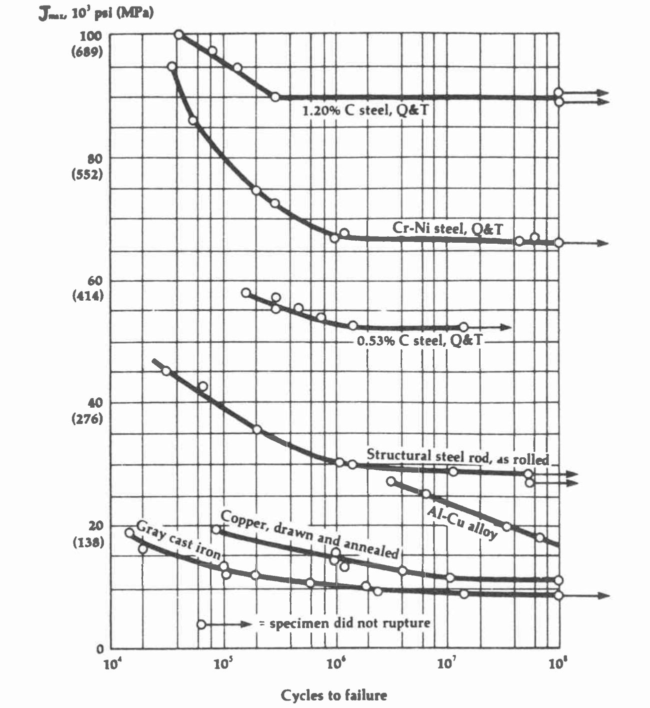


Figure 7 – Comparison of the fatigue behaviour of several ferrous and nonferrous alloys [9].

figure spheroidal graphite is called compacted graphite and the B iron has a higher spheroidal graphite content than the A iron. The fatigue behaviour of a variety of cast irons (which are detailed in Table 1) is compared in Fig. 5 [8].

As a generalisation, alloy additions that increase the cyclic strength in aluminium alloys will increase the fatigue strength. The overall improvement, however, is much affected by precipitate type and distribution [8], as discussed above. Naturally aged alloys (T3 condition) containing Guinier-Preston zones exhibit pronounced hardening, while artificially aged alloys (T6) that contain additional transition phases show a modest hardening. Treatments involving cold work have little effect on cyclic properties. The fatigue performance of four aluminium alloys is compared in Fig. 6, while that of several different metallic alloys is compared in Fig. 7 [9].

As a final point, it is probably true that more significant advantage can be obtained from attention to design detailing and finishing (e.g. grinding smooth and allowing sufficient radii at changes in section), and by surface treatments which induce compressive residual stresses (e.g. shot or hammer peening) than through complex alloying additions or heat treatments. While mentioning surface treatments, it is worth noting that coatings (e.g. Ni or Cr plating) often reduce fatigue strength because of intrinsic cracking or pick-up of nonmetallic material during the process. In certain applications, however, the additional benefit which can be obtained from careful processing becomes cost effective, and the techniques outlined above are used to advantage.

4. Increasing resistance to crack propagation

Whenever a jointed structure (welded, bolted or rivetted) is under consideration the fatigue life is propagation dominated and crack initiation is not a significant concern. Fatigue design and life prediction then utilises fracture mechanics based growth rate curves, as seen in Fig. 2. As noted in a previous section, for these purposes the growth rate in regime B is relatively little affected by alloy composition, heat treatment or mean stress (stress ratio) and a single design curve, derived for a specific joint type, can be used for design purposes. Nonetheless, in any particular application fatigue resistance can be increased through the use of tough, ductile steels. Generally speaking, high strength and high toughness do not occur in conjunction in metallic alloys, although certain high toughness, high strength alloys have been developed. Thus the higher the strength (and hence the better the intrinsic resistance to crack initiation), the lower the toughness and the faster fatigue crack growth rates become for a given stress level.

Where increase in yield strength is derived from grain refinement in a clean steel, however, toughness also increases and controlled rolled grain-refined, precipitation-hardening, high-strength low-alloy (HSLA) steels have been developed which optimise the combination of cost, strength and toughness. These steels are low carbon steels which contain small amounts of Al, V, N, Ti and Nb (typically 0,03-0,10% of each element). Grain refinement is achieved through grain boundary pinning by fine dispersions of nitride and carbide particles such as AlN, Ti (CN) and NbC, which also increase strength. Microstructures can be either ferrite – pearlite in normalised steel or acicular (lath) ferrite, bainite or martensite in quenched and tempered steels.

In pearlite structures, toughness increases with fineness of the pearlite and a decrease in pearlite content. Toughness is generally highest with fine acicular ferrite (which is also the optimum microstructure for welded joints). Fig. 8 indicates the effect of carbon content, or pearlite volume fraction, on Charpy impact toughness in normalised steels [10].

Damage-tolerant aluminium alloys such as 7475, 7050 and 2124 have been achieved primarily through control of the alloy microstructure. Table 2 gives the three types of second-phase particles that are known to influence fracture and fatigue behaviour.

Fracture of brittle constituent particles leads to prefer-

Table 2: Second-phase particles which influence fatigue behaviour in aluminium alloys.

	Size		
Type	μm	mil	Typical Examples
Constituent particles	2-50	0,08-2	Cu_2FeAl_7, $CuAl_2$, $FeAl_6$
Dispersoid particles	0,01-0,5	0,0004-0,02	$ZrAl_3$, $CrMg_2Al_{12}$
Strengthening precipitates	0,001-0,5	0,00004-0,02	Guinier-Preston zones

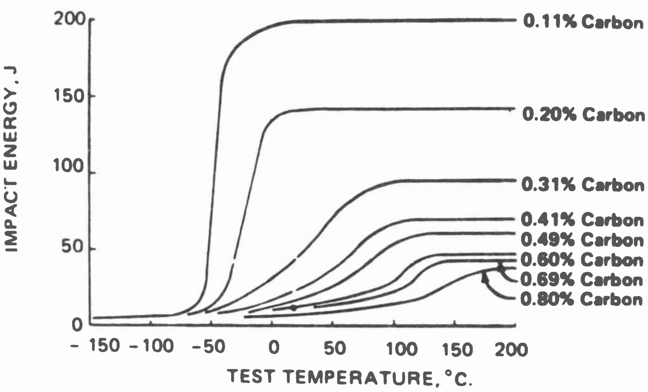


Figure 8 – Effect of carbon content on impact toughness in normalised steels [10].

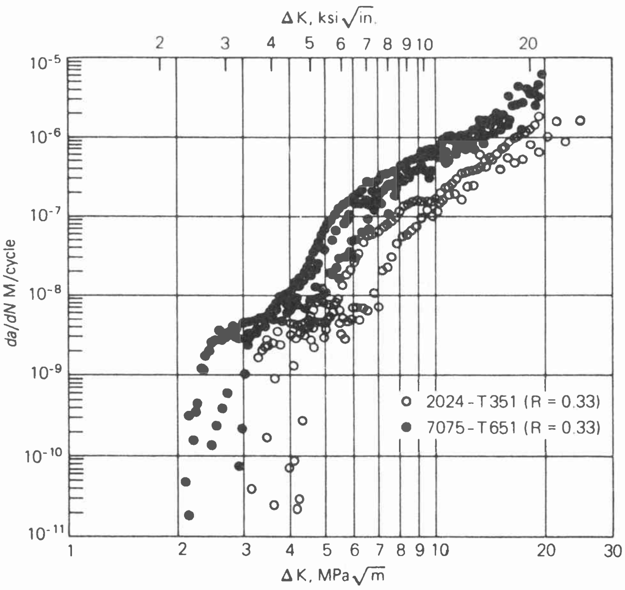


Figure 10 – Fatigue crack growth behaviour of 2024-T3 versus 7075-T6 plate [11].

ential paths for crack advance. Hence the toughness of high strength alloys can be improved by reducing iron and silicon levels. The improved toughness of 7475, 7050 and 2124 relative to 7075 or 2024 derives, in the main, from use of higher purity base metal. Dispersoid forming elements should be held to the minimum required for control of grain structure, mechanical properties or resistance to stress corrosion cracking. Cr is particularly deleterious to crack propagation toughness in 7XXX alloy sheet, as seen in Fig. 9 [11]. The primary effect of hardening precipitates on fracture toughness is through the influence on yield strength; higher yield strength equates, generally, to lower toughness. However compositional changes, particularly in Mg level, can affect toughness in 7XXX alloys through subtle influences on the overall precipitate structure. Typical improvements in crack growth rate behaviour that are achieved through these means are

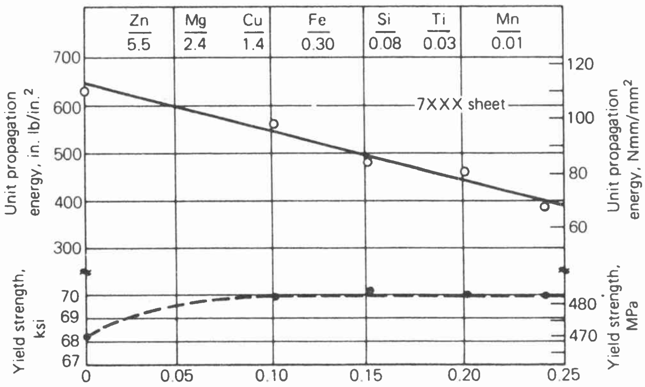


Figure 9 – Effect of Cr content on crack propagation energy in 7XXX sheet [11].

shown in Fig. 10. Within the 7XXX alloy system, increasing Cu content improves fatigue performance in high humidity. This is attributed to the increased resistance to corrosion.

It is important to note that whenever the fatigue life of a product is crack growth dominated, regular nondestructive inspection has to be specified at prescribed intervals as part of the overall design process.

4.1 Threshold Design

As stated earlier, the threshold value of stress intensity range, ΔK_{th}, can be viewed as a fatigue limit for cracked components. For a typical high toughness pressure vessel steel with $\Delta K_{th} \approx 5$ MPa$\sqrt{m}$, the variation in threshold stress with crack length is given in Fig. 11. As welded structures commonly contain cracks up to 0,5-2,0 mm in length, it is clear that the threshold stress (which gives infinite life) is rather low. Hence the acceptance of the possibility of crack growth in such structures, which is coupled with in-service inspection. However, it is possible to increase the threshold value by a number of strategies, most of which involve reducing strength and give a relatively small increase in threshold. One possibility does appear to offer a combination of high strength and low growth rates – duplex microstructures, which usually contain either islands or 'fibres' of martensite within a tough matrix or network of ferrite or austenite [12, 13]. These will be briefly discussed shortly.

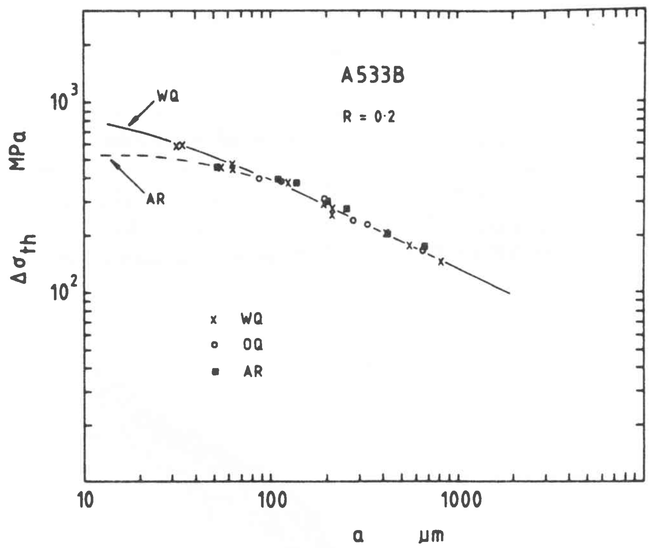


Figure 11 – Variation in threshold stress amplitude with crack length at R = 0,2 for A533B steel. The yield strength of the normalised condition (AR) was 439 MPa and that of the quenched and tempered *structure (WQ) was 799 MPa.*

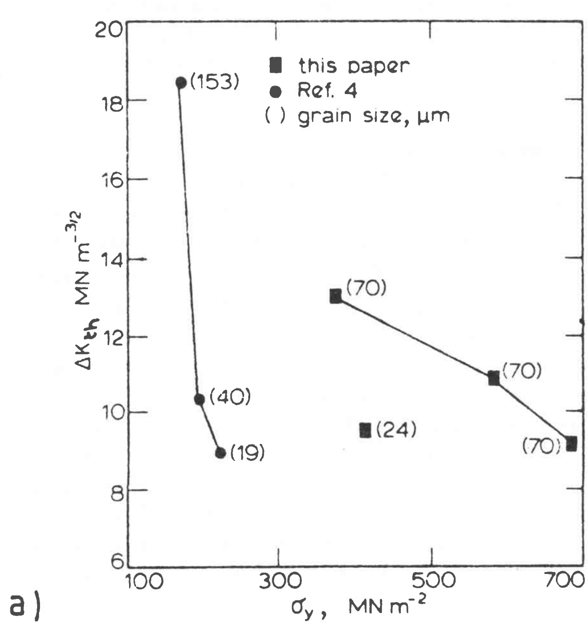
Changing either the grain size or the yield strength in steels changes the threshold value. Usually the individual effect of these two factors is not clear, as they generally occur in conjunction. But, as seen in Fig. 12a for an Fe-Ti-C low alloy steel [14], there can be a considerable effect of either of these parameters acting on their own. Rather high threshold values are found in this steel and appear to be the result of crack branching and crack deviation from the plane perpendicular to the maximum principal stress (which is a function of grain size). The applied load ratio, R = $\sigma_{min}/\sigma_{max}$ in the fatigue cycle, also has a significant effect on the threshold value as seen in Fig. 12b [14]. This arises from increased crack closure, or crack wake wedging, which occurs near the minimum cyclic strees at low stress ratios.

4.2 Duplex Steels

Duplex, fatigue resistance microstructures of the type mentioned above have been reported in carbon steels and alloy steels [12, 13]. Mechanical properties of several duplex steels are given in Table 3 [12] and the fatigue crack growth rate curves for normalised and duplex 1018 steel are given in Fig. 13. In this particular case, there is a dramatic increase in threshold and reduction in growth rates with the duplex microstructure, even though the strength is some 70% higher in this material. Although not as impressive, significant increases in threshold value were observed for the other duplex microstructures. These differences were explained in terms of the more tortuous crack path, and hence higher closure, found in the duplex microstructures [12].

A rather interesting example of the development of a high strength, high toughness, fatigue resistant maraging steel is found in fencing blades [13]. The objective here is to develop a material which is fail-safe, i.e. a blade will not fracture in two, but will suffer gross deformation once a fatigue crack of sufficient size exists in it. Starting with a carbon steel, a duplex microstructure is developed in the following way. 10 mm square bars are electroplated with nickel and tightly packed into a 200 mm diameter sealed canister. This is extruded to produce 50 mm bar at 1000 °C. Subsequent hot working produces a 10 mm square bar, which is 100% dense. At the same time, the nickle coating diffuses into the steel to produce an inter-connecting layer of austenite. Strength of the steel can be varied by altering the proportion of the austenite phase by varying the thickness of the initial nickle coating. Hardness of the tempered martensite phase can be changed through heat treatment.

The toughness of this material is compared with that of



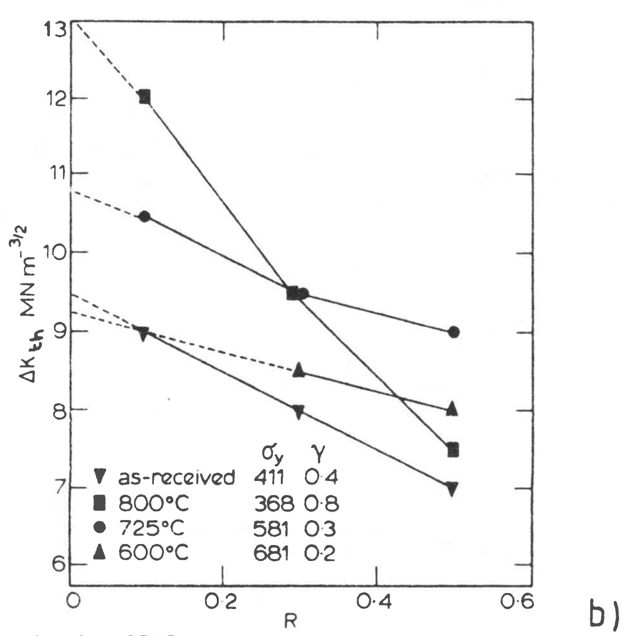


Figure 12 – Threshold value as a function of [14]:
(a) grain size and yield strength at R = 0;
(b) load ratio and microstructure.

Table 3: Mechanical properties of duplex steels.

Material	0,2% Offset Yield Strength (MPa)	Ultimate Tensile Strength (MPa)	Elongation 25,4 mm Gauge Length	ΔK_{th} (MPa$\sqrt{m}$)
AISI 1018 (Duplex)	427	842	10	13,8
2,25 Cr-1 Mo (Duplex)	414	798	17	13,9
AISI 1045 (Duplex)	777	1387	<2	6,1
10B35 (1) (Duplex)	753	1145	5	7,3
10B35 (2) (Duplex)	690	1104	5	8,3
AISI 1018 (Normalized)	255	441	41	8,3
10B35 (Normalized)	397	651	33	7,3

the conventional carbon steel in Fig. 14. This exceptional toughness arises from crack tip blunting at the interface between the austenite matrix and the martensite 'fibres'. This composite microstructure gives a very slow fatigue crack growth rate, again due to crack blunting along the length of the fibres. Such innovative processing techniques appear to offer significant benefits for property optimisation in metals.

5. Conclusions

The intrinsic fatigue resistance of metals can be significantly enhanced through careful alloying additions and controlled thermo-mechanical processing, through the techniques outlined above. The role of careful attention to design detailing and fabrication (including surface improvement treatments) can never, however, be underestimated. A majority of fatigue problems arise from poor detailing or careless fabrication, rather than incorrect materials selection.

Where detailing and fabrication are optimised and fatigue resistance is of prime importance to the function of a product, it is hoped that this brief discussion will provide some insight into ways of achieving that goal.

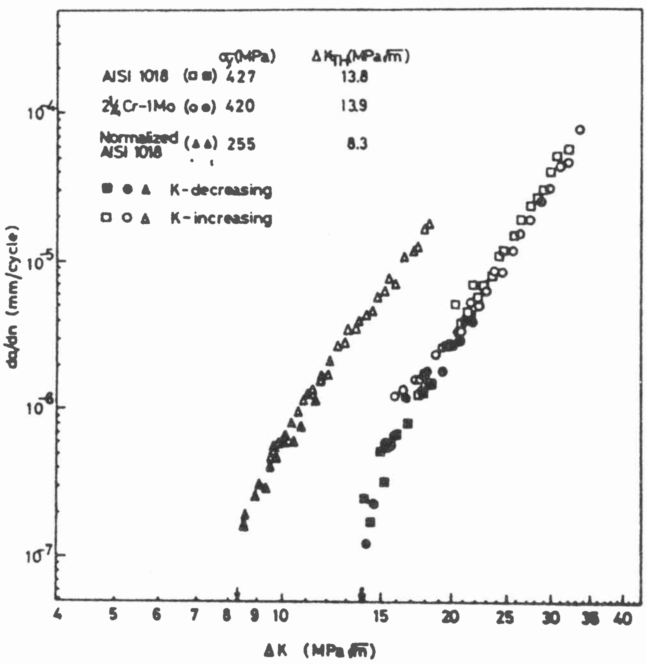


Figure 13 – Crack growth rate curves for normalised and duplex AISI 1018 steel [12].

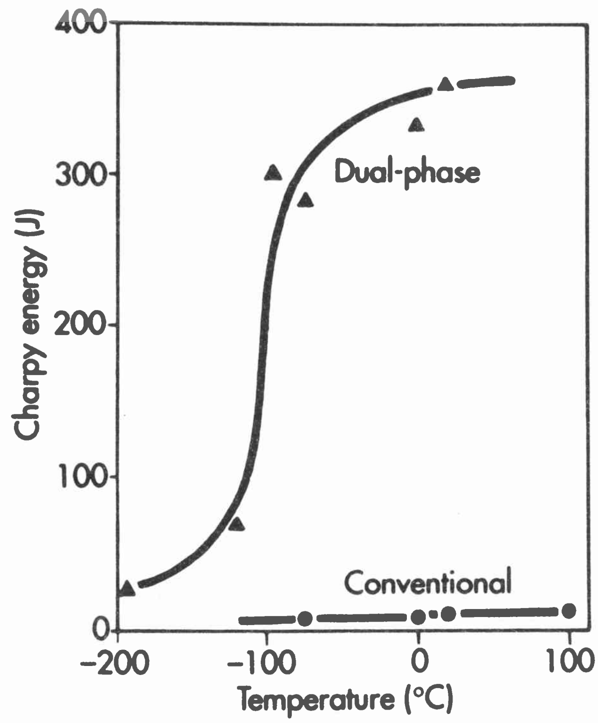


Figure 14 – Comparison of Charpy impact energy for dual-phase steel and conventional steel [13].

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