DIFFERENCES BETWEEN THE GROWTH OF SMALL AND LARGE FATIGUE CRACKS. THE RELATION TO THRESHOLD K-VALUES

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DELTFT - THE NETHERLANDS

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ABSTRACT

Differences between the behaviour of small cracks and large cracks are analysed. Aspects included cover microscopical and fractographic observations. Differences are related to restraint on cyclic slip, number of active slip systems, crack growth mechanism and the irregularity of the crack front geometry, both on a micro level and a macro level.

Another difference is associated with residual plastic deformation in the wake of the crack. Large cracks will be subject to more effective crack closure than small edge cracks. Illustrative crack growth data are presented. Comments are made on the meaning of nominal K-values. Threshold K-values for large cracks are not necessarily valid for small cracks.

Keywords: fatigue crack, microcrack, macrocrack, threshold, crack front.
1. INTRODUCTION

The potential usefulness of the stress intensity factor for the correlation of fatigue crack growth rates appears to be widely accepted by now. The first basic idea behind the application of $K$ is that it characterizes the distributions of deformations, strains and stresses in the crack tip field in a sufficient way. The second basic idea is a similitude argument. Similar distributions of different cracks, as indicated by the same $K$-value, should produce the same crack extension $\Delta a$. A cyclic load is characterized by two $K$-values, and the similitude implies:

$$\Delta a = \frac{da}{dn} = f(K_{\text{max}}, K_{\text{min}}) = f(\Delta K, R)$$  \hspace{1cm} (1)

The crack closure phenomenon was discovered by Elber [1] and he introduced $\Delta K_{\text{eff}} = K_{\text{max}} - K_{\text{op}}$, with $K_{\text{op}}$ being a function of the stress ratio $R$. The above equation is then replaced by:

$$\frac{da}{dn} = f(\Delta K_{\text{eff}})$$  \hspace{1cm} (2)

Because this relation was apparently capable to account for stress ratio effects of several materials, it reassures the idea that the $K$-factor gives an adequate indication of the stress intensity around the tip of a fatigue crack, provided the crack tip plastic zone is small.

Problems around the similitude approach became evident when very small cracks were considered. The $K$-concept may break down for several reasons, such as: (1) material inhomogeneity, with characteristic dimensions comparable to the crack size, (2) plastic zone sizes which are no longer small if compared to the crack size, and (3) crack growth mechanisms, which can be different for small cracks and large cracks (e.g. stage I instead of stage II).

It is surprising that the crack growth mechanism was not questioned
that much for large cracks. It certainly was known, at least to microscopists, that a crack front is far from being a straight line on a micro level. Perhaps it was tacitly assumed that similar K-levels will produce similar crack growth mechanisms, and as a result the similitude model appears to be still correct.

Crack initiation and early crack growth are technically relevant. 
(i) For finite life problems the major part of the life is generally covered by the growth of micro cracks and small macro cracks (say up to 5 mm).
(ii) For infinite life requirements fatigue limits are a most significant material property. The fatigue limit may be either a threshold for crack initiation, or a threshold for crack growth from small material defects or micro fatigue cracks.

The major part of our present knowledge on fatigue crack growth is mainly coming from large cracks. It is important to know whether this information is applicable to small cracks as well. The problems addressed in this paper are related to the comparison of the behaviour of large cracks and small cracks. For this purpose fracture mechanics aspects are summarized first, followed by material considerations pertinent to different conditions for small and large cracks. Illustrative crack growth observations of different materials are presented. The paper is concluded with a number of conclusions.
2. ASPECTS OF FRACTURE MECHANICS

Fracture mechanics by now is a well developed branch of continuum mechanics. Aspects and assumptions related to fatigue cracks are summarized in table 1. With respect to the geometry of the crack it is already a significant complication if, instead of a through crack with a straight crack front (2d problem), a part through crack has to be considered (3d problem). Unfortunately small cracks usually are part through cracks.

In view of the stress singularity at the crack tip, occurring in linear elastic fracture mechanics, it always was recognized that a plastic zone will occur. The size and the shape depend on the deformation conditions prevailing at the crack tip, and on the plastic behaviour of the material. Fracture mechanics has produced various estimates on plastic zone sizes, usually giving only one dimension, thus ignoring the shape of the plastic zone. More elaborate analysis can produce better information which then requires a modelling of the plastic strain hardening behaviour of the material.

The difference between the monotonic plastic zone and the reversed plastic zone is recognized, especially so by the work of Rice [2], which made it clear that the reversed plastic zone is much smaller than the monotonic zone. The introduction of residual plastic deformation in the wake of the crack, accounting for crack closure, has been a subject of calculation. However, such calculations were restricted to simple cases (through crack, simple deformation conditions).

Even for a through crack in a plate specimen the reality is not as simple as desirable for continuum mechanics calculations. The crack front is slightly curved and shear lips are formed at the plate surface (variation of opening mode along the crack front). Moreover there is a transition from plane stress conditions at the surface to predominantly plane strain at mid thickness which implies a variation
of plastic zone size along the crack front, and consequently a variation of crack closure over the material thickness [3]. Beyond any doubt it has been shown that available fracture mechanics procedures can give useful first order approximations of stress intensities, plastic zone sizes, etc. However, if such results are applied to specific crack growth data, it has to be ascertained whether the real crack growth phenomenon in the material still is in acceptable agreement with the assumptions made for the calculations. This requires both fractographic observations to be made on one hand and a good understanding of the relevance of calculated results on the other hand.
3. MICROSCOPICAL AND FRACTOGRAPHICAL OBSERVATIONS

Differences between the growth of micro cracks and macro cracks can be related to the crack growth mechanism and to material aspects, in the latter case more specifically to material inhomogeneity. A survey of aspects involved is given in tables 2 and 3, which will be partly discussed below.

Polycrystalline material. Initially a micro crack may grow in just one or a few grains, whereas a macro crack has its crack front running through many grains. It is possible that a micro crack still has a fairly regular crack front. For a macro crack it is difficult to believe that the crack front on a microscale is still very regular. Actually there is sufficient evidence that the crack front level is rather irregular. This is illustrated by pictures in Figure 1. The striations in Figure 1a show that the macro crack is progressing on plateau's at different heights. Figure 1b shows sharp bends of the striations. Both TEM pictures were obtained with the usual replica technique. However, the SEM pictures in Figures 1c and 1d were made from plastic casting of cracks. Bowles developed a vacuum infiltration technique to produce a casting of a crack [4]. The crack front itself can then be studied. Note in Figure 1c that it shows a rounded crack tip, a wavy crack front and striations from both sides of the fatigue crack in one picture. The crack front obtained in vacuum, Fig. 1d, is completely different. A systematic tendency to crack branching was observed.

The length of the crack front in view of the various irregularities shown in Figure 1 is consistently longer than the nominal length assumed in continuum mechanics calculations. Fortunately, it need not invalidate the applicability of Eqs. (1) and (2), provided the same irregularity always occurs for the same $\Delta K$ and $R$ (or $\Delta K_{eff}$) values. For macrocracks this may be expected, but comparing a macro crack and a micro crack the similarity seems to be questionable.
Aspects related to slip. Cyclic slip causing fatigue at low stress amplitudes can be considered to be microplastic deformation. Nevertheless the plastic strains involved on the micro level are high. They therefore will meet with restraint of the elastic surrounding. The restraint is quite different for a small slip band crack at the free surface as compared to slip at a crack tip inside the material. At the free surface dislocations can run out of the material. In addition to a lower restraint it also implies a less effective strain hardening. In most f.c.c. and b.c.c. materials a number of equivalent crystallographic slip planes is available. It should be expected that a low restraint on cyclic slip will predominantly promote the slip system with the highest resolved shear stress (i.e. the highest Schmid factor). Initial slip band cracking (Forsyth Stage I) observed in several materials confirms this argument.

After penetration into the material, away from surface grains and with a crack front running through many grains, it becomes extremely difficult to maintain slip band cracking in one single direction in each grain. That is more or less incompatible with a coherent crack front. In view of increasing restraint on cyclic plasticity and crack front coherency more slip systems will be activated and as a result the crack will grow in an apparently non-crystallographic direction.

Some years ago Louwaard [5] found an anomalous crack growth behaviour in coarse grained 7075-T6 sheet material, which was not found for fine grained material. At low ΔK-values (tests in vacuum at -100°C) crack growth, also inside the material exhibited a constant preference for slip band cracking, which disappeared at higher ΔK-values, compare Figures 2a and 2b. During the slip band cracking stage crack branching was frequently observed (Fig. 2a), which is necessary indeed to maintain a more or less coherent crack front. As a result a rather irregular crack front topography must have occurred, which produced an erratic crack growth and relatively low crack rates. The transition to the more usual growth behaviour (Fig. 2b) implied a more regular crack front, and the growth rate jumped to a much higher value (about 4x).
Crack growth mechanisms. The growth mechanisms are outside the discussion here, also because these mechanisms can be different from material to material. However, one observation should be mentioned here. The environment can have a considerable effect on the growth mechanism. As a result, different fracture topographies can occur, and for aluminum alloys this has been found indeed [6]. Bowles' observations in figures 1c and 1d indicate significant branching in vacuum, which did not occur in normal air. As a result, the actual K factor was much lower than in air and that essentially contributes to the environmental effect. However, also on the macro level effects of environment on the fracture topography can be observed. Figure 3 shows differences between fatigue fractures obtained in vacuum, air and salt water. The macro waviness of the fractures is quite different. The less aggressive environment produces the more undulated fracture surface and thus the longer crack front, which implies a lower apparent K-value. Again this is an essential part of the explanation of the environmental effect. Shaw and Le May [8] reported a similar observation for AISI 4140 steel. The question to be raised now is whether the crack surface topography of small cracks will be similarly affected by the environment. There is hardly any evidence available on this issue. However, from the slip restraint argument discussed before it appears questionable whether the geometries of micro cracks and macro cracks can respond to the environmental impact in a similar way.

Plasticity in the wake of the crack. Significant plasticity in the wake of the crack, causing crack closure, has amply been shown for large crack, even at large distances behind the crack tip. For small cracks the size of the wake is small anyhow. It would be surprising that the effect of crack closure would be similar for small and large cracks. This issue is considered later.

Elastic anisotropy and grain boundaries. It is generally accepted that elastic anisotropy and grain boundaries are not of great interest to macro cracks, because the material behaviour is then controlled by the
average bulk properties. Obviously this need not be so for microcracks, especially if they are still contained within a single grain. Cyclic slip should then be affected by the crystal orientation and the grain boundaries. Recent microscopical work of Kung and Fine [9] and Morris [10] has shown that grain boundaries have a significant effect on microcracks which are still in the initiation phase. This is noteworthy because aluminum alloys have a low elastic anisotropy and many slip systems, while cross slip is relatively easy. For other materials larger grain boundary effects should be possible. More complex situations can occur in material with two phases. Unfortunately detailed microscopical studies on fatigue crack nucleation and early microcrack growth are rather tedious and the literature is not very abundant in view of the large variety of material structures.
Inclusions and early crack growth. Scattered through the literature there is evidence that micro cracks in some materials start from inclusions. Early work of De Lange [11] was done on unnotched rotating bending specimens of an Al-alloy. With a special replica technique micro cracks were already found in the range of 1 to 10 µm. Up to a crack length of 100 µm the crack rate was significantly higher than expected from the growth data for cracks in the range of 100 µm to 5 mm (see Appendix A of [12]). A similar trend is evident from Pearson's data [13] reproduced here in Figure 4. The intersection of the two curves in Figure 4 occurs at a crack length of about 130 µm (0.005 in). In an NLR study [12] the growth of small cracks in Al-alloy notched and unnotched specimens could be followed from 0.1 mm (= 100 µm) onwards. The crack rates of small semi-circular corner cracks and larger through cracks indicated the same K-controlled growth behaviour [14].

This was confirmed in NASA work [15], where it was also found that scatter occurred primarily in the crack initiation phase, whereas it was much less in the crack growth phase. The observation on scatter is easily accepted if microstructural variations are significant for crack nucleation and early micro crack growth. It is noteworthy that De Lange's observations of some specimens suggest a subsurface crack nucleation from inclusions. Recent fractography of specimens tested under flight-simulation loading [16] gave similar indications, subsurface intermetallic particles of some microns being the origin of the crack. A breakthrough to the surface will then suggest an initial high crack rate.

Inclusions and macro crack growth. It is usually thought that macro crack growth is depending on material bulk properties, with inclusions being of minor importance. Striation patterns have confirmed that inclusions had a small local effect only [17]. However, Forsyth and Bowen [18] and Bucci et al [19] reported for Al-alloys that inclusions retarded macro crack growth by initiating secondary cracks. This produced an irregular crack front and consequently a slower growth.
4. SOME ILLUSTRATIVE CRACK GROWTH DATA

The macroscopic irregularity of crack fronts has been shown to be also significant for the results of static fracture toughness tests on CT specimens. This is shown by the $K_{IC}$-values in table 4. Precracking of the Ti-alloy specimens at a low fatigue load level caused "structural sensitive cracking" on preferred crystal planes [20]. As a consequence an irregular crack front was obtained. However, precracking at a higher fatigue load produced a flat structure-insensitive crack plane. For the irregular crack front the $K_{IC}$-value was about 50 percent higher than for the more straight crack front. De Jong [21] found a smaller, but similarly systematic trend for a 7075 alloy. On the non-planar fatigue crack 3 or 4 ridges in the growing direction connected crack growth on slightly staggered levels. The planar cracks did not have such ridges.

In a recent test series [22] fatigue cracks with an irregular fracture topography were provoked by employing an edge notch crack starter with a blocked profile. As shown in Figure 5 the crack rate was systematically lower than for a macroscopically flat crack. This difference disappeared after some crack growth when the initial irregularities had leveled out. This is another indication that larger irregular crack fronts are apparently associated with a higher crack growth resistance.

In an investigation on crack growth in mild steel plates Truyens [23] found some apparently inconsistent trends. The scatter band of $\frac{da}{dn}$-$\Delta K$ data was small, see Figure 6. However, there were deviations at the lower side of the band. Fractographic examination indicated that initially low crack rates were due to multiple crack initiations at the starter notch. This causes an irregular crack front which explains the lower growth rates. In another test series scatter was low again, but now the initial growth was faster than expected from the scatter band, see Figure 7. Crack closure measurements indicated that $S_{op}$ was low in the beginning, thus causing a higher $\Delta K_{eff}$ which leveled off after some crack growth, see the inset in Figure 7. Some crack growth has to occur
before there is sufficient residual plastic deformation in the wake of the crack to increase the crack opening stress level $S_{op}$.

Similar results as shown for a saw cut notch in Figure 7 have been reported by several authors for fatigue crack growth starting from open holes, e.g. [24-26]. Also for stationary flight-simulation load histories the same trend was observed, both for cracks starting from a saw cut as well as from an open hole [27]. In Figure 8 another example of test results is presented, which shows that plastic deformation in the wake of the crack has to be built up, before it can become effective in reducing $\Delta K_{eff}$. In the flight-simulation load histories of 3 test series the only difference was in the rarely occurring high loads. If the stress level of these loads was increased the crack growth rate was lower. The results for $S_{max} = 187$ MPa show that the growth rate reduction is not obtained immediately. First some crack growth has to occur after which an effective "plastic wake field" is built up.

The difference between the wake field of a small edge crack at a hole and a large central crack in a panel is illustrated in Figure 9. For an edge crack at a hole a $K$ solution for a finite width panel was obtained by Newman [28]. Assuming the plastic zone size to be proportional to $K^2$: 

$$r_p = C \left( \frac{K}{\sigma_{yield}} \right)^2$$

Figure 9 shows that $r_p$ for the small edge crack increases rapidly. For a central crack in a finite width panel this occurs much more gradually. If the small edge crack and the large central crack have the same $K$-value, and thus the same $r_p$-value, the same crack rate will be predicted. However, the plastic wake field is fully different and it should be expected that crack opening can be obtained more easily for the small edge crack as compared to the large central crack. Hence $\Delta K_{eff}$ will be higher for the small crack. Actually the difference is a
history effect related to dK/da. It is difficult to say whether this is also the explanation for Pearson's results in Figure 4, because other differences between small and large cracks may have contributed as well.

A final investigation to be referenced here was carried out by Helle [29] on a modern Cu alloy used for ship propellers (Cu 9 Al 5 Ni 5 Fe). He studied crack initiation and the growth of both small cracks and large cracks. Most tests were carried out in salt water to $10^7$ and $10^8$ cycles. Micro crack growth at the surface started as slip band cracking (shear mode) provided $S_{max}$ was not too high. The material had a low stacking fault energy. After some growth a transition to some quasi cleavage mechanism (tensile mode) occurred. The transition was controlled by $K_{max}$, which in Figure 10 implies a linear $\Delta K$-R relation. For large cracks Helle determined $\Delta K_{thr}$ by careful load shedding procedures. As shown by Figure 10 $\Delta K_{thr}$ of the large cracks is significantly higher than the K-values at which small tensile mode cracks are still growing. Apparently there is a dilemma and Helle offers a number of arguments to explain it. The two more important ones to be mentioned here are:

1. Early slip band crack growth is associated with cyclic plasticity mainly on one slip system. As a result residual plastic deformation perpendicular to the crack path will hardly be built up. Crack closure of these small cracks will thus be less effective. For larger cracks with slip on two (or more) differently orientated slip planes a more usual plastic wake field will be present.

2. If a large crack is growing at a very low crack rate the inhomogeneity of the stress intensity along the crack front becomes important. Due to the high elastic anisotropy of the Cu alloy the K value for some grains can drop to a level considerably lower than the nominal one. As a result a transition of the crack growth mechanism from the tensile mode to slip band cracking occurred in some grains, as confirmed by fractographic observations. Local slip band cracking causes a more irregular crack front, which then induces a further drop of the stress intensity, more slip band cracking and further increase of the crack
front irregularity. Crack arrest is then possible at nominally high
K-values, higher than K-values at which small cracks can still grow.
Helle's results are a dramatic illustration of a different behaviour
for small and large cracks, which could only be understood by combining
fractographic observations and an analysis of the mechanisms involved.
5. DISCUSSION

Literature information on the difference between the growth of small cracks and large cracks easily gives the impression that small cracks behave somewhat exceptional, if not anomalous. Although this may be true, it was shown in the previous sections that the geometry of large cracks can differ significantly from the shapes assumed in calculations based on continuum mechanics. Fatigue fracture surfaces of large cracks are macroscopically not flat, while microscopically there are even more irregularities. The nominal K-value based on the apparent crack length can be adopted for fatigue crack growth predictions, provided similar fractographic features can be expected. This may apply to fatigue cracks in many materials. Unfortunately in the literature quite often fractographic observations are not reported.

The application of data of large fatigue cracks to small cracks is more questionable. It was discussed before that initial edge cracks sometimes grow faster than large cracks at the same nominal K-value. In view of this trend El Haddad et al [25] introduced a corrected crack length. Instead of

\[ \Delta K = C \Delta S \sqrt{\pi a} \]  

(3)

they adopted:

\[ \Delta K = C \Delta S \sqrt{\pi (a + a_o)} \]  

(4)

where \( a_o \) was supposed to be a material constant. For a steel a value \( a_o = 0.24 \text{ mm} \) was quoted [30] and as result of this small value \( a_o \) can be ignored for large cracks. However, for small cracks \( a_o \) leads to higher K-values than the nominal ones of equation (3). With equation (4) similar \( da/dn-\Delta K \) plots were obtained for small and large cracks. It should be mentioned that a plasticity correction was also required. The empirical \( a_o \)-correction was motivated by more severe conditions for
small surface cracks, due to lower restraint on slip and less effective strain hardening [31]. Actually these arguments imply that the similarity between small cracks and large cracks is lost, in spite of the good empirical correlation found.

The applicability of K factors to very small cracks has also drawn attention in view of threshold values in relation to the fatigue limit. Kitagawa and Takahashi [32] found a constant \( \Delta K_{th} \) for various crack length values (steel, \( S_u = 800 \text{ MPa} \)), except for cracks smaller than \( 2a = 0.5 \text{ mm} \). A similar trend was found by El Haddad et al [30]. It confirms that very small edge cracks grow faster than predictions by the nominal K-value.

Smith [33] briefly discussed the minimum crack length below which the meaning of K would become doubtful in view of material inhomogeneities as compared to plastic zone sizes. He mentioned minimum crack length values of \( a = 0.26 \text{ mm} \) for mild steel and \( a = 0.05 \text{ mm} \) for a Ni Mo steel. Lankford [34] in a similar study mentions values of \( 1 \text{ mm} \) for mild steel and \( 0.01 \text{ mm} \) for a high strength steel (AISI 4340). For aluminium alloys it appears that for small cracks with \( a > 0.1 \text{ mm} \) the crack rate correlates quite well with the data of larger cracks [13-15]. For lower \( a \)-values this it not at all certain. It now appears that K underestimates the stress severity at the crack tip if the crack length is smaller than a few tenth of a millimeter, or a small number of grain diameters. Probably this minimum crack length for the applicability of K will be smaller for high yield strength materials.

The phenomenon of non-propagating cracks was analysed by Hammouda et al [35] and by El Haddad et al [36]. In both publications an initial decreasing K-value for a small growing crack was derived by employing arguments based on notch root plasticity. The arguments were different in the two publications, but they have in common a decreasing strain amplitude at an increasing distance away from the notch root surface. The crack tip should have the benefit of it, and a K reduction is deduced
from this effect. In the opinion of this author crack growth will stop only if cyclic crack tip opening has (practically) gone down to zero (removing the stress singularity at the crack tip). This can occur as a result of increasing crack closure and increasing crack front irregularities. As long as there is cyclic crack tip opening there will be cyclic crack tip plasticity in view of the stress singularity. As a consequence there will be some crack growth. This appears to be the crux of the threshold problem. Threshold conditions require that cyclic crack tip opening is practically zero. This need not imply that the nominal $\Delta K_{th}$ is zero, but rather that $\Delta K_{\text{eff}}$ at the crack tip should be negligible. Increasing crack closure can be a result of plastic deformation built up in the wake of the crack. It is interesting that Ritchie referred to "crack tip blocking" by corrosion products to explain environmental effects on $\Delta K_{th}$ of a high strength steel [37]. This is another way of crack closure leading to a reduced $\Delta K_{\text{eff}}$ and threshold conditions. The threshold problem will not be discussed here any further. There are many variables involved, which were recently discussed by Blom [38]. An extensive survey of recent investigations will become available as the Proceedings of the International Symposium on Fatigue Thresholds, held in Stockholm, June 1981.
6. CONCLUSIONS

(1) Most crack growth data in the literature were obtained for large cracks. However, small cracks are technically relevant, because the growth period of small cracks covers a large part of the fatigue life, while non-propagating cracks are important for infinite life. It is therefore highly opportune to see whether the same fundamentals apply to small cracks and large cracks.

(2) A large crack has an irregular crack front on a microscale. Also on a macroscale it can be irregular, even for a through crack with a apparently straight crack front. As a result the length of the crack front can be much larger than assumed in fracture mechanics. Crack branching and other irregularities can further complicate the picture.

(3) The applicability of linear elastic fracture mechanics to large cracks seems to be fortuitous in view of the previous conclusion. However, it may well be explained because the same irregularities will occur to all large cracks in the same material.

(4) For small cracks the irregularities of the crack front as found for large cracks need not be present, at least not to the same amount. As a consequence, the application of $\frac{da}{dn}-\Delta K$ data of large cracks to small cracks can only be justified if microscopic studies have shown that similar growth mechanisms do occur on a microscale.

(5) From a plasticity point of view microcracks are in a different position as compared to large cracks. The restraint on cyclic plasticity and cyclic strain hardening are less effective at the material surface as compared to the interior of the material. This can even lead to a different cracking mechanism.

(6) There is a second plasticity reason, why small cracks differ from large cracks. A growing crack produces residual plasticity in the wake of the crack. The plastic wake fields of a small crack and a large crack will be different because of different growth histories. As a result crack closure for a small edge crack will be less effective than for a large crack.

(7) As a consequence of the previous conclusion threshold $\Delta K$-values for
small cracks and large cracks will rarely be the same.

(8) Threshold conditions for a fatigue crack can only be obtained if the cyclic crack tip opening has practically gone down to zero. In view of crack closure, irregular crack fronts and other complications $\Delta K_{th}$ can still be positive.

(9) There are indications that the stress intensity factors can be used for cracks as small as a few tenth of a millimeter, depending on the type of material. Probably the usefulness of $K$ for small cracks will be better for high yield strength materials.
REFERENCES


[29] H. Helle: Investigation on ship propeller fatigue. This comprehensive study will be published in 1981.


Table 1: Survey of aspects of continuum mechanics.
<table>
<thead>
<tr>
<th>Aspects of mechanisms</th>
<th>MICRO CRACKS</th>
<th>MACRO CRACKS</th>
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<tr>
<td>crack front length/(grain diam.)</td>
<td>small</td>
<td>large</td>
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<tr>
<td>restraint on slip</td>
<td>low</td>
<td>high</td>
</tr>
<tr>
<td>number of active slip systems</td>
<td>1 can be sufficient</td>
<td>≥2 (usually)</td>
</tr>
<tr>
<td>slip band cracking</td>
<td>possible</td>
<td>difficult</td>
</tr>
<tr>
<td>crack growth by</td>
<td>- shear decohesion</td>
<td>different mechanisms can apply</td>
</tr>
<tr>
<td></td>
<td>- tensile decohesion</td>
<td></td>
</tr>
<tr>
<td></td>
<td>- environmentally assisted decohesion</td>
<td></td>
</tr>
<tr>
<td>striations</td>
<td>no(?)</td>
<td>yes</td>
</tr>
<tr>
<td>plasticity in wake of crack</td>
<td>limited</td>
<td>evident</td>
</tr>
<tr>
<td>fracture surface topography</td>
<td>?</td>
<td>not flat on micro level</td>
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Table 2: Aspects of the crack growth mechanisms.
<table>
<thead>
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<th>Aspects of material inhomogeneity</th>
<th>MICRO CRACKS</th>
<th>MACRO CRACKS</th>
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<td>elastic anisotropy</td>
<td>can be significant</td>
<td>averaged effect</td>
</tr>
<tr>
<td>inclusions</td>
<td>can start micro cracks</td>
<td>small effect (sometimes retarding)</td>
</tr>
<tr>
<td>grain boundaries</td>
<td>can be significant</td>
<td>average effect</td>
</tr>
<tr>
<td>material surface layer examples</td>
<td>nitriding</td>
<td>large effect (local properties)</td>
</tr>
<tr>
<td></td>
<td>decarburizing</td>
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<td></td>
<td>shot peening</td>
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<td>rough surface</td>
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Table 3: Aspects of material inhomogeneity.

<table>
<thead>
<tr>
<th>References</th>
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<th>Planar crack</th>
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<tbody>
<tr>
<td></td>
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<td>$K_{max}$ (MPa $\sqrt{m}$)</td>
<td>front</td>
<td>$K_{IC}$ (MPa $\sqrt{m}$)</td>
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<tr>
<td>Stubbington/Gunn [20]</td>
<td>Ti-6Al-4V (LT)</td>
<td>low (10.8)</td>
<td>no</td>
<td>72.9 (1.48)</td>
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<td></td>
<td></td>
<td>high (19.1)</td>
<td>yes</td>
<td>49.3</td>
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<td>De Jong [21]</td>
<td>7075-T6 (TL)</td>
<td>14-18</td>
<td>no</td>
<td>30.6 (1.13)</td>
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<td>14-18</td>
<td>yes</td>
<td>27.1</td>
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Table 4: Effect of an irregular crack front on $K_{IC}$. 
Fig. 1a: Striations on different plateau's. Random load fatigue in air.

Fig. 1b: Sharply bent striations. Fatigue in air.

Fig. 1c: Curved crack front with rounded tip. Fatigue in air [4].

Fig. 1d: Crack branching along crack front. Fatigue in vacuum [4].

Fig. 1: Microcharacteristics of fatigue cracks in an Al-alloy (2024-T3). Figures 1a and 1b: replicas in TEM. Figures 1c and 1d: plastic castings in SEM.
Fig. 2: Different crack growth mechanisms at low and high $\Delta K$ [5].
Fig. 3: Effect of environment on the macroscopic topography of fatigue cracks in an Al-alloy (7075-T6, thickness 6 mm) [7].

Fig. 4: Crack growth results of Pearson [13] for an Al-alloy (L65 - 2014-T3).
Fig. 5: Slower crack growth due to an irregular crack front [22].
Fig. 6: Low initial crack growth rates due to erratic crack initiation. Results of Truyens [23].

Fig. 7: High initial crack growth rates due to less effective crack closure. Results of Truyens [23].
Fig. 8: Slower crack growth due to more crack closure after some crack growth [27].

Fig. 9: Different plasticity in the wake of a small edge crack and a large central crack.
Fig. 10: $\Delta K_{\text{threshold}}$ of large crack much higher than $\Delta K$ for the growth of small cracks. Results of Helle [29].