Chapter Two

Damage and cracking morphology

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Abstract

This chapter summarizes the experimental and analytical studies of damage and crack morphology. Close attention is paid to the role of material microstructure and loading conditions on damage evolution in engineering materials. Fracture mechanisms and geometrical specifics of damage initiation, accumulation and crack propagation are analyzed primarily in terms of their relation to and influence on the material microstructure. We have considered mechanisms of crack-trajectory formation in brittle materials. Fatigue, corrosion and fretting phenomena have been analyzed from the angle of damage morphology. Experimental analysis of failure and crack nucleation of single-crystal superalloys is described in the last section. In this chapter we briefly characterize technique used for revealing the damage morphology in deformed materials.

1 Introduction

Failure is a very localized process evolving in time and leading to global structure instability when some limiting state is met. The goal of fracture mechanics is not limited to developing of crack-growth algorithms but also should include (i) prediction of the relationship between material microstructure and a part susceptibility to failure; (ii) description of the failure mechanisms and damage history from fractography; and (iii) determination of design requirements accounting for different modes of failure. An understanding of damage mechanisms is fundamental to design. This chapter describes how the materials properties and loading schemes affect the damage initiation and propagation. Analysis of fracture surfaces of failed components yields information such as fracture-origin location, stress characteristics, and service environment. There are several distinct patterns on fracture surfaces, such as radial striations; regularly spaced ‘rib’ markings, each of which is specific for the
particular distinguished damage mechanism. The fracture-surface markings allow determination of the approximate crack-propagation rate, direction, and the origin. Fracture morphology studies structural mechanisms and geometrical specifics of damage initiation, accumulation and crack propagation and their relation to and influence on the material microstructure. In this chapter we briefly characterize techniques used for revealing the morphology and microstructure in deformed materials. We analyze the major microscopic features of the fracture surfaces and their relationship with the fracture-initiation and propagation processes.

Among micromechanical-deformation mechanisms, crystallographic slip, twinning, diffusive phase transformations, and grain-boundary sliding are the most closely related to material damage. They determine the evolution of the microstructure in materials, and in turn, cause the failure process, which might be schematically illustrated as the following time sequence: damage initiation, void growth, shear localization or cleavage and fracture. Brittle fracture has been studied since Griffith (1921), however, dependence of the crack-path geometry and crack growth-stability on loading schemes still cannot be easily predicted. Therefore, formulation of the crack-growth criteria and their connection with material properties and morphology are the primary objectives of engineering mechanics. In contrast to brittle fracture, ductile fracture is rate sensitive and characterized by micro-void nucleation, growth, and coalescence. A considerable amount of inelastic deformation may occur prior to fracture. This mechanisms leads to fracture criteria that strongly depend on the loading history and material microstructure evolution (McClintock and Argon [1]). Fracture-mechanics and, more generally, continuum-mechanics approaches to fracture are based on the assumption that material is homogeneous. However, material deformation and microstructure are different in different volume elements, especially close to the crack or notch tip (Sih [2]). Without a careful morphological analysis one would not be able to identify the origin, growth rate, and transition regimes of elastic-plastic damage for different loading parameters. Most structural materials and metals are polycrystalline. Because of the misfit in neighboring grain crystallographic structures, the grain boundaries have low (with respect to grain itself) surface energy, which makes them natural sites of failure nucleation, starting, for example, from grain boundaries sliding and subsequent shear localization. Grain boundaries can be viewed as arrays of crystal defects (for example, dislocations), Sutton and Balluffi [3]. It makes these regions more “porous” than the crystals themselves and significantly increases the rate of impurity diffusion to the boundary area. These atoms segregate to the boundaries with damaging effect on the fracture toughness (Ashby and Jones [4]). Thus, slight changes in material crystallographic texture or amount of impurities could cause the drastic change of the failure mechanism from transgranular to intergranular fracture. The importance of the distinguishing of damage nucleation mechanisms cannot be overestimated especially in fatigue and related phenomena (such as fretting fatigue and corrosion fatigue) where up to 90% of the part life is controlled by the crack initiation. Engineering practice now
requires development of a unified degradation model, which analyzes and predicts the failure modes at all physical length scales from microstructural consideration to the macroscale part level. Fracture morphology is only one available experimental approach that allows for defining and classifying micromechanics of damage and in turn leading us to the formulation of advanced macromechanical constitutive laws.

2 Summary of the deformation behavior under monotonic loading

Energy dissipation in a loaded structure could take place either by plastic deformation or by microcracking. Material plasticity decreases structure susceptibility to the cracking, but if the density of plastic defects becomes too high, secondary cracking might be initiated. The most typical mechanism of plastic deformation in crystalline materials is slip along definite crystallographic planes, or in other words, generation and propagation of dislocations inside crystals. Under applied stress, dislocations run into each other, generating new dislocations and also forming a dislocation pile up next to an obstacle. Depending on multiple slip geometry, these intersected dislocations could be immobile or sessile and play the role of such an obstacle. This means that slip-band intersection may result in crack appearance. Another important mechanism of plastic deformation is crystallographic twinning. The high stacking fault energy materials (e.g. Al, Cu) deform predominantly by crystallographic slip. In contrast, for materials with low stacking fault energies

![Figure 1. Dislocations in a nickel superalloy.](image)
(brass, bronze, h.c.p. materials, e.g. Mg, Zn) in addition to slip, deformation twinning plays an important role in maintaining generalized plastic flow. The overall plastic deformation of a crystal is always inhomogeneous at length scales associated with slip and twinning. Twinning is a more complicated deformation mode than slip because, in addition to the sudden large twinning shear ($\gamma=0.707$ for f.c.c. structures), it produces a volume fraction of the grain with a very different lattice orientation from that of the rest of the grain; in a sense, it produces new grains.

Strong lattice reorientation is the major cause of materials localized anisotropic response and plays an important role in a crack nucleation. If two or more twin systems are activated in the grain, they might intersect creating a highly deformed zone. Practically, no materials can sustain two sequential lattice rotations on (for f.c.c. materials) $41^\circ$ each. The twin intersection is called Rose’s hole, and it is a crack-nucleation site.

Deformation mechanisms in polycrystals, especially in polycrystalline materials with low symmetry, for example magnesium, are much more complex than for single crystals. The most important non-crystallographic mechanism of inelastic deformation is grain-boundary sliding (Hauser et al. [5]) as shown in the micrograph of Fig. 3. Shear displacement across grain boundaries is observable at relatively low strains. One can see from Fig. 3 that the grain boundary gives rise to multiple slip and twinning in the adjacent grain-boundary area. This additional inhomogeneity may result in damage nucleation.
Summary of deformation failure under monotonic loading

All structural elements are subjected to different influences and loadings that eventually may lead to fracture. Failure phenomena are very complex and vary from brittle fracture, creep fracture, fatigue, and ductile fracture. Brittle fracture manifests itself as crack propagation after little or no plastic deformation (McClintock and Argon [1], Cherepanov [6], Hertzberg [7]). Brittle fracture may proceed along grain boundaries, called intergranular fracture or through the grain itself, called transgranular or cleavage.

There are three primary factors that control brittle fracture, namely, material toughness, initial crack size, and stress level. A macroscopic fracture-mechanics analysis is based on calculation of stress-intensity factors (SIF) for elastic or small-scale-plasticity deformation [1, 2, 5]. Originally, the SIF concept was introduced as the asymptotic solution of an elastic straight crack problem. Small-scale yielding is the second basic concept of the linear theory and holds when the plastic zone at the crack tip is much smaller than the crack length [7]. It allows use of the elastic singular solution and its approximation (SIF) as a good approximation to the actual stress/strain fields.
A plastic zone develops at the crack tip, which is approximately equal to

$$r_p = \alpha \left( \frac{K_I}{\sigma_y} \right)^2 ; \quad \alpha = \begin{cases} \frac{1}{3\pi} & \text{plane strain} \\ \frac{1}{\pi} & \text{plane stress (diffuse)} \\ \frac{\pi}{8} & \text{plane stress (Dugdale band)} \end{cases}$$ \hspace{1cm} (1)

A correction for the plastic zone can be done through the calculation of the crack effective length. Macroscale fracture mechanics (both linear elastic fracture mechanics (LEFM) and non-linear fracture mechanics (EPFM) J. Rice [8]) assume the fracture toughness $K_{IC}$ or critical energy-release rate $J_C = \frac{K_{IC}^2}{2\mu}$ are material constants. For cracks longer than several microns ($\sim 10^{-5}$ m), for example, Balakin [9], Nemat-Nasser and Hori [10], this is always true for most engineering materials. The local fracture criterion is met when $K_I$ reaches $K_{IC}$ and a crack starts propagating. For the global strength of the material, it is important to know if the crack propagation is stable or unstable. If $\frac{\partial K_I}{\partial a} > 0$, the crack will propagate in an unstable fashion because the SIF increases with the crack length. Otherwise, crack growth is stable. As an example, the stress-intensity factor for a central crack with the length of $2a$ is $K_I = \sigma \sqrt{\pi a}$, where $\sigma$ is gross-area normal tensile stress, and its growth will be unstable. The crack separated by two point forces will propagate slowly.

Figure 4: Typical cleavage steps on a fracture surface.
because the SIF is \( K_I = \frac{P}{\sqrt{\pi a}} \), and for the crack growth an increase of load is required.

Ductile fracture occurs after significant plastic deformations. A single crystal or each grain in a polycrystalline undergoes multiple slip along different crystallographic slip planes, which is especially significant for fcc metals (Cu, Ni, Al, etc.). There is no crack nucleation due to sufficient plastic-deformation mechanisms and the material undergoes plastic deformation until the start of shear localization or gradual reduction of the cross-sectional area – necking. However, if there are strong obstacles in the ductile materials, such as particles of second phase, interfaces, etc., then dislocations in the slip line are stopped and piled up, generating stress concentration running into a shear crack. This is an explanation why high-symmetry crystallographic structures crack along slip systems due to shear.

Figure 5: Plastic zone formed by extensive slip lines around a crack tip in the single f.c.c. crystal. (Photo courtesy of D. P. DeLuca).

Plastic deformation can accommodate prescribed strain without failure of the structure. Therefore, the general trend in material science is to combine high
strength with some ductility. For most materials, local yielding occurs at the crack tip. Subsequently, much attention should be paid to the mechanisms of crack-tip plastic deformation instead of the analysis of an elastic singular stress field. Due to local plastic deformations, the crack tip becomes blunted and the crack propagates in a more stable fashion without catastrophic failure. From the macroscopic point of view, the plastic zone is a nominally homogeneous region. However, close to the crack, the materials morphology plays an important role in the formation of an inhomogeneous plastic region. Extensive crystallographic slip around the crack tip at high temperature is shown on the micrograph of single Ni-base superalloy crystal in Fig. 5.

Viscoplastic fracture starts with nucleation and growth of voids, situated mostly inside grains. This means that the size and the growth rate of these pores are the primary characteristics of the fracture. Traces of these pores are clearly observable on any fractography of ductile fracture. At elevated temperatures, the process becomes rate dependent and can take place under relatively low stresses. A typical micrograph of a damaged turbine blade is shown in Fig. 6. Much attention has been paid to the study of void nucleation, growth, and coalescence as a micromechanism of ductile fracture, and to the formulation of constitutive models for micro-porous ductile solids over a wide range of temperatures. Gurson’s model [11] is the most complete rate-independent model suitable for low homologous temperatures. For elevated temperatures, Duva [12] and Zavaliangos and Anand [13] outline the structure of rate- and temperature-dependent constitutive models, which, to some extent, account for void mutual interaction. It has been shown that void volume depends on porosity, applied stresses and strain rate. Particular phenomenological expressions are defined by the form of material constitutive relations, for example in [13] it was derived that

$$\dot{V} \sim \dot{\epsilon}_0 \left( f \frac{\sigma - \sigma_f}{(1 - f^m)^n} \right)^{\frac{m}{n}}$$

where $f$ is the porosity and $m$ is a material fitting parameter. Porosity or void-volume fraction is the macroscopic state parameter defining material damage and its growth rate is highly nonlinear, and in some cases might be exponential (Belak and Minich [14]). Finally, the cross section of the specimen decreases and failure occurs. Thus, the criteria of ductile fracture should be based on critical plastic strain and/or a critical value of porosity.

In order to investigate mechanisms of deformation and failure at different

![Figure 6: Voids due to high-temperature creep of the turbine blade airfoil. Magnification in 50x.](image-url)
temperatures, we have conducted mechanical tests on h.c.p. polycrystalline magnesium alloy AZ31B at room temperature and at 350°C. At this elevated temperature (the melting point for pure magnesium is 650°C) twinning is suppressed and non-basal slip systems are activated. In single crystals in addition to basal, prismatic $<a>$, and pyramidal $<c+a>$ systems, slip may occur on the six second-order pyramidal $<c+a>$ systems, and the degree of plasticity increases markedly, becoming comparable to that of cubic metals. At 300°C the possible elongation before failure is approximately nine times that at room temperature (Raynor [15], Yoo [16]). A very good example of this may be seen from the analysis of fracture surfaces after tension tests. At high temperature the specimen was fractured leading to an ideal (and very sharp) cup and cone appearance of Fig. 7 (left). For comparison in Fig. 7 (right) is shown the specimen tested at room temperature. All specimens that were deformed in our tests (simple tension, compression and plane strain compression) at room temperature exhibited brittle fracture. After compression tests, the angle between the compressive axis and the normal to the crack plane is close to 35°.

The tension specimen exhibits relatively little necking before fracture and fractured by the apparent shear. Fig. 8 shows the SEM micrographs of the fracture surfaces after tension at 350°C and at room temperature. These micrographs illustrate that at room temperature the fractures were relatively brittle and magnesium is cracked by twin-induced cleavage fracture. With suppression of twinning and activation of $<c+a>$ slip systems, the intrinsic
resistance to cleavage increases as well as ductility and the fracture surface shows typical plastic dimples.

Another type of cracking occurs during high-temperature creep under low stress. Small cavities are formed on the grain boundaries and especially at triple points (Meyers and Chawla [17]). The most extreme mechanism of intergranular creep fracture is diffusion failure, taking place under very high temperatures and low stress levels, when steady-state dislocation diffusion is noticeable. This mechanism is not pronounced up to the temperature of almost two thirds of the melting point. As an example, for the testing temperature of 400°C, the self-diffusion coefficient for magnesium does not exceed the value of $3.3 \cdot 10^{-11} \text{cm}^2\text{s}^{-1}$, which is at least two orders of magnitude less than the minimum threshold value we should take into account in plastic-deformation analysis.

As has been discussed, the different fracture mechanisms correspond to different ranges of applied stresses and temperatures. These ranges are defined usually by fractography or fracture vs. time analysis. Extending the idea of phase diagrams, the classification of fracture mechanisms can be presented in a simple graphical form called Frost and Ashby fracture mechanism maps. One puts normalized homologous temperature ($\frac{T}{T_m}$) along the abscissa and normalized stress ($\frac{\sigma}{E}$) along the ordinate. Regions of different fracture types are set on the map. An excessive list of fracture deformation maps is given in (Ashby and Jones [18]).

Figure 8: Fractographs of the magnesium alloy AZ31B after simple tension at (a) 350°C and (b) room temperature.
4 Crack geometry in brittle materials

All experimental results known to the author show that a macroscopic crack propagates in the direction perpendicular to the uniaxial tensile loading. In most practical situations the loading conditions are more complex than uniaxial. Under pure shear, the crack will propagate in its plane only in the case of strong material anisotropy. Such conditions are realized in single crystals with pronounced crystallographic glide planes where shear localization transfers to the shear or Mode II crack. In contrast, for isotropic materials, normal separation or Mode I controls the crack trajectory, i.e. a shear crack will turn off its original plane to the direction where $K_I$ at the tip of a small crack wing is maximum. Experimental studies have revealed that collinear cracks under normal uniaxial tension do not coalesce along the straight path, but avoid each other, (Melin [19]). It was observed that in brittle materials, a crack propagates due to the random microcracking in the process zone around the crack tip. Thus, on the micro-scale, a crack is not a straight line but jagged. Shear stress generated by the first crack tip causes non-zero $K_{II}$ at the tip of the second inclined micro-crack and turns it.

Shear of the initially inclined flaw is also the basic mechanism of the crack growth under compression. Fracture under compression is a very important problem in geomechanics and in structural parts with strong residual stresses where compression is the predominant loading mode. The crack propagates slowly aligning itself along the line of maximum principal compressive stress. There are two well-distinguished situations that can be evaluated by asymptotic methods: initial crack branching and long crack with small initial inclined flaw (Erdogan and Sih [20]). A schematic of crack growth under compression is shown in Fig. 10.
Initial branching is described by a singular integral equation (Hori and Nemat-Nasser [21]), which can be reduced to finding the asymptotes of SIF for small, in comparison to the initial crack, branch sizes (Arshon and Staroselsky [22]). We have shown there that the SIF has the following general expression

\[
\frac{e^{-\theta}}{2\pi \cdot \sqrt{\pi c}} (K_I + i K_{II}) = a_0(\theta, \gamma) + a_1(\theta, \gamma) \sqrt{\frac{c}{c}} + O\left(\frac{1}{\sqrt{c}}\right). \tag{2}
\]

The principal terms of this asymptotic expansion may be taken as approximate formulae that are useful for qualitative analysis especially when numerical methods become unreliable. The real part of the complex coefficient \(a_1 \cdot e^{i\theta}\) is negative with any choice of parameters, which means that \(\frac{\partial K_I(\ell, \theta)}{\partial \ell} < 0\). Therefore, the growth of crack branches is stable at the initial stage of propagation. The limiting values of the SIF at the end of the small crack

\[
k_i = K_i \bigg/ \sqrt{\pi c (\sigma_1 - \sigma_2) \sin \gamma (\cos \gamma - \mu \sin \gamma) - \mu \sigma_2 + \tau_c}\bigg) \tag{3}
\]

do not depend on initial crack inclination \(\gamma\). Here \(i = I\) and \(II\); \(c\) is a half-length of the initial crack, \(\mu\) is the friction coefficient, \(\tau_c\) is the dry friction coefficient.

Figure 10: A schematic of crack growth under compression: (a) 3D representation; (b) plane cross section; and (c) large crack with a central splitting force.
threshold, $\sigma_1$ is the largest and $\sigma_2$ is the second (side) compressive principal stresses. As is evident from the calculations shown in Fig. 11, $K_I$ decreases with increase in friction. $K_I(\Theta)$ reaches a maximum and $K_{II}(\Theta)$ vanishes at the same value of $\Theta = \Theta(\mu)$. Hence, $K_I^2 + K_{II}^2$ also reaches a maximum at the same value of $\Theta$, i.e. the use of any of the mentioned criteria ($K_I \to \max$; $K_{II} = 0$; and $K_I^2 + K_{II}^2 \to \max$) of brittle crack growth gives the same values of the branching angle. Our calculations have shown that branching angle is practically independent of the initial crack inclination and linearly decreases with friction as follows: $\Theta^*(\mu) = 68^\circ - 5^\circ \mu$. The value of the angle is close to experimental data. Parametric curves of $x = \pm c \pm l \cos(\theta(l)); \ y = \pm l \sin(\theta(l))$; $(0 \leq l \leq l_0 < 0.1 - c)$ give an idea of the form of curvilinear branches. As follows from our numerical analysis of the asymptotic two-term solution, the character of the crack curvature under uniaxial compression ($\sigma_2 = 0$) depends on an initial, frictionless ($\mu = \tau_c = 0$) crack-inclination angle $\gamma$. In the interval $\gamma \in (0, \pi)$, the results clearly show a concave shape and for $\gamma \in (\pi/2, \pi)$ — a convex shape of the branches. This suggests that in the last case the crack changes its convexity one more time. As we show, microcracking is the primary mechanism of the nonlinear deformation behavior at the initial stage. Crack branches are initiated at the tips of the initial crack and propagate in the direction of the major principal compressive stress. After some time, these crack wings become larger than the initial crack itself and axial crack growth is now controlled by shear splitting of the initial fault, which can be approximately modeled as split forces at the crack center; Fig. 10c. Growth of such a crack is stable as well as in the case of small wings. It led us to make a plausible conclusion that brittle and quasi-brittle crack growth in the plate under compression is always stable.
The shape of the specimens plays an extremely important role in the crack propagation and morphology. The study of fracture characteristics and morphology of cracks in brittle materials has broad applications from material science to geomechanics (Bahat [23]). Experimental studies of the brittle and/or already damaged materials, such as some rocks or construction materials cause additional difficulties. The existing methods of measurement of fracture toughness (see, for example, [24–27]) involve the laborious process of introduction of a crack to initiate fracture. Furthermore, fatigue crack growth from the initial notch cannot be recommended for brittle materials, since it increases lengths of existing microcracks and other defects that change the morphological properties of the sample. It would be desirable to determine fracture toughness using samples without initial cracks. This can be performed by monotonic loading of a disk with a circular hole in the center by a pair of compressive forces. The central hole, playing the role of the defect, initiates fracture (Staroselsky [28]).

To obtain the formula of fracture toughness for this load scheme, we take as a basis the approximate solution obtained by Murakami [24]. We asymptotically transform his expression using the fact that the size of the initial defect is of order of a grain and is much smaller than the radius of the hole. Finally, the expression has the form as follows:

\[
K_{IC} = F^* \frac{P \sqrt{R_{in}}}{\sqrt{\pi} R_{out} h} \left(1 + \frac{d}{R_{in}}\right),
\]

where \( P \) is applied load, \( R_{in} \) and \( R_{out} \) are radii of the hole and of the disk, respectively, \( h \) is the sample thickness, \( d \) is the average diameter of the grain.
and $F'$ is the correction function depending only on the ratio $\beta = \frac{R_{in}}{R_{out}}$; shown in Table 1.

Table 1: Values of the fracture-toughness correction function for the ring specimen tests.

<table>
<thead>
<tr>
<th>$\frac{R_{in}}{R_{out}}$</th>
<th>0.05</th>
<th>0.1</th>
<th>0.15</th>
<th>0.2</th>
<th>0.25</th>
</tr>
</thead>
<tbody>
<tr>
<td>$F'$</td>
<td>1.7</td>
<td>1.85</td>
<td>2.0</td>
<td>2.2</td>
<td>2.5</td>
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The crack propagation is unstable and, subsequently, the fracture toughness may be determined from the geometrical parameters and the critical value of the applied force $P$. Note that for most materials, the second term in the parenthesis may be dropped.

![Figure 12](image_url)

Figure 12: Two ring specimens after testing. (a) Two vertical symmetrical cracks have developed beginning at the edge of the hole along the diameter in the applied force direction. (b) The ring specimen containing a horizontal pre-existing crack, destroyed by two vertical symmetrical cracks developing along the diameter.

We have tested disk specimens with radii varying from 21 to 36 mm. The thickness of disk specimens of 15 mm ensures plane-strain conditions. A picture of the solid sample after the test is presented in Fig. 12a. As expected, in all our tests the crack develops symmetrically, beginning at the diameter of the inner hole away from the hole in the applied-force direction.
It is often not possible to produce a solid specimen due to disruption of structures. Typical examples of this are layered materials, composites, and weak, cracked materials. We have analyzed the applicability of our method to the analysis of such materials. The sample shown in Fig. 12b contained a crack before testing. The crack length did not exceed the half-radius of the disk. The load was applied in the direction perpendicular to this pre-existing crack. Experiments showed that the symmetrical vertical cracks in the direction of loading (Fig. 12) grow first. The horizontal (pre-existing) crack begins to develop only after the vertical cracks reach the outer disk boundary, i.e. the load is driven to the pre-existing crack at the moment of the destruction of the specimen, causing it to propagate. Thus, the possible existence of cracks in a ring sample does not influence the results. The method permits examination of materials that are not successfully tested by standard methods, for example, weak, layered rocks, such as argillite. We compare these test results with data from ASTM four-point bending of prismatic samples with an edge crack [26, 27] and the difference has not exceeded 10%.

Concluding this section we would like to emphasize that ahead of the tip of a brittle crack, a process zone contains a randomly oriented set of microcracks that start developing in the closest vicinity of the main crack tip and several of them coalesce extending the main crack. There is no statistically stable effect of microcracks on SIF of the major crack (Kachanov [29]) due to their random orientation. Thus, the crack growth is the sequence of the random micro-breakage events, which finally align the major crack trajectory in such a way as to maximize SIF for normal separation $K_f \to \text{max}$. Cracks with maximal values of SIF (Mode I) grow first and may shield other cracks in the specimen. Shear-stress components predominantly define the crack geometry and changes in $K_f$ with the crack length determine propagation stability.

5 Fatigue and environment

The service life of structural components can be governed by several modes of degradation and the dominant mode of failure is due to fatigue. Structural components have to be designed so that they can adequately endure the fatigue loading during their service life. There are two commonly recognized forms of failure: high-cycle fatigue (HCF) and low-cycle fatigue (LCF). HCF is observed under small-amplitude elastic vibrations and the part might eventually fail after $10^7-10^9$ cycles. A typical part subjected to HCF is a turbine-fan airfoil. LCF is the degradation mode characterized by high loading amplitude and, concurrently, by plastic strains. These conditions occur, for example, in areas of stress concentration of a turbine blade; (DeLuca [30]).
Fatigue damage is usually initiated at free surfaces. A rare exception from this rule is contact failure (or fretting fatigue) where the crack nucleates in a subsurface zone where shear stresses reach their maximum value. The failure is caused by extensive plastic-strain localization in cyclically stressed materials, at least for most technically important metals and alloys. Slip-band development on the surface, as shown in Fig. 13, is the source of fatigue cracks; (DeLuca [30]). Microcrack formation results from the inability of materials to relax stress concentrations in the intrusion/extrusion cavity by plastic shear. Crystallographic microcracks deepen with successive cyclic deformation. Resolved shear stress should be favorable for active crack propagation. Thus, if a primary slip system is parallel to the surface, the microcracking would be significantly delayed. There are experimental evidences that the lifetime of repeatedly electro-polished specimens is orders of magnitude longer than unpolished ones (Klesnil and Lukas [31]). This demonstrates the crucial role of intrusions/extrusions in fatigue crack nucleation.

Fatigue failure is a process extended in time, which may be conventionally divided into three stages. First, excessive plastic shear localization takes place at surface and subsurface regions. Secondly, microcracks are nucleated in the zones of stress concentrations caused by shear localization and by microdefects.

The third stage starts when a microcrack begins propagating and transforms into a macrocrack. This microcrack has to be favorably orientated and located for crack growth (Romaniv et al. [32]). Initially, microcracks propagate along shear...
bands and for polycrystalline materials crack growth is defined by maximum shear stress. With the crack growth, the trajectory turns perpendicular to the direction of principal tensile stress (Tetelman and McEvily [33]). At this stage, local phenomena at the crack tip fully control the crack propagation. Finally, when the crack reaches a critical length, then spontaneous fracture occurs.

Figure 14: SEM fractograph (magnification 5000×) of a fatigue specimen made from turbine disk superalloy and tested at 260°C.

Typical fatigue-fracture surfaces consist of (i) crack-initiation site, (ii) crack slow-propagation zone where fatigue fractures exhibit typical striations. Macro-branching of the crack and propagation in parallel planes with later rejoin forms radiating straight ridges or hackle marks perpendicular to the crack front. The load history can be reconstructed by the fractography: striations are formed in each loading cycle, (Broek [34]), resulting from slip growth at the crack tip. Formation of regular striations requires noticeable material plastic behavior; in brittle materials striations are ill-defined and might be restricted to a favorably oriented crystallographic grain [34]. Advanced structural materials such as nickel-base superalloys undergo extensive plastic deformation prior to failure; therefore, small cracks can be subjected to extensive localized damage with significant amounts of plasticity (Staroselsky et al. [35]). We have conducted high-temperature fatigue tests at a typical turbine-disk bore temperature of 260°C. The scanning electron microscope (SEM) micrograph in Fig. 14 shows the details of the fracture surface and striations at large magnification, which is typical for plastic crack propagation. In the same micrograph one can observe several hackle lines normal to the crack front. Although it is widely accepted that each striation represents one load cycle, it is quite difficult if not impossible to use them for practical measurements of crack growth rate. Overstressing generates a much wider striation than usual loading cycles. It is consistent with
the assumption that the striation spacing is a measure of the crack growth rate. Also, it provides the simple method of the fatigue-crack rate measurements. After each say thousand cycles, the specimen is overloaded for several cycles and then the loading is set back to a nominal value.

![Figure 15: Evolution of the fatigue crack front. Striation marks produced by overstress are wider than others and can be clearly counted.](image)

The overstressed wide striation lines are clearly observable as shown in Fig. 15 and the distance between them is easily measurable.

Formation of striations in vacuum is suppressed and does not take place in every cycle, (Broek [34]), but even such a neutral environment as air significantly accelerates fatigue-crack growth and facilitates new surface oxidation and, subsequently, irreversible slip formation leading to striations. The role of environmental attack can be crucial during fatigue as described in Vasudevan and Sadananda [36]. For obviously aggressive environments the fatigue and corrosion interact and enhance the total damaging effect that becomes greater than the simple superposition of these modes.
6 Corrosion and fretting fatigue

In a corrosive environment the fatigue-crack-growth behavior of a structure is the result of the mechanically induced stresses and strains and an electrochemical deterioration of the material. The basic concept is that the electrochemical process results in a pitting of the material with microcracking within the pits. As the pit deepens and the microcracks coalesce in a load environment, the stress intensity associated with the crack front increases to the point where a crack propagates from the corrosion pit. It has been established, (Scully [37]), that when a metal is treated with an electrolyte, anodic and cathodic sites are generated on the metal surface and an electrical current flows between them, resulting in a loss of metal as a result of dissolution. Despite the fact that this metal-degradation process has been extensively studied, there is still no reliable model predicting pitting kinetics and growth in metals with different microstructure subjected to different environmental corrosion attacks.

![Figure 16: Pitting attack on Al 7075 specimen after three months in NaCl solution at 50°C.](image)

The pitting process (see Fig. 16) depends on the material, corrosive environment, and possibly loading. Material parameters include the type of material and its microstructure, i.e. grain structure, texture, and the surface conditions.

It is experimentally observed (Shafig and Agarwala [38]) that pit growth is more active in the transverse direction of rolled plates, and that the majority of pits stop growing at some point while a few grow disproportionately to a large size. The role of applied stress still is not understood in the pitting process.
The process of crack nucleation and short-crack propagation may cover the significant part of structural safe life. The stress field in the vicinity of the crack controls fatigue-crack growth. This field is a combination of externally and internally generated stresses. The interaction between the two processes arises from the internal stress dependency on the corrosion. This non-linear interaction is greatest in the short-crack region when the crack is both metallurgically and physically small based on the characteristic corrosion size. The fatigue starts dominating the pitting when the crack growth rate is becoming higher than the pitting expansion. The pit-crack transition strongly depends on microstructure and electro-chemical factors as well as on the degree of oxide-induced crack closure.

Figure 17: Pit-crack transition. (a) Crack growth from a surface pit with multiple corrosive attacks along the crack trace. (b) Fracture surface showing multiple initiated cracks from pits.
Stress corrosion, pit growth and corrosion-fatigue cracking are time-dependent phenomena, because they are related to the kinetics of electrochemical reactions. These reactions are immensely accelerated under stress and cyclic loading (Shafig and Agarwala [38], Renauld and Song [39]). Corrosion occurs very intensively at the microcrack tips where the stress level is very high. The rate at which corrosion-crack growth can occur is controlled by the transport rate of the corrosive material in the crack-tip region, or the rate of the electrochemical process at the crack tip. The transport process includes transport of corrosive elements to the propagating crack tip as well as transport of the by-products of the crack-tip reactions, such as hydrogen into the metal inducing embrittlement reactions. The crack tip under stress becomes an anode with a relatively small area and the surrounding region serves as a cathode (Agarwala [40]). This electrical potential difference (up to 100 mV) between the two is a driving force to speed up the corrosion process (Ford [41]). This also intensifies the metal-dissolution process, which in turn leads to crack blunting and crack branching. The two latter processes reduce the energy density around the crack and slow down the process. Synergism of corrosion and fatigue involves two specific regimes wherein environmental effects and mechanical effects are superimposed on each other. The products of corrosion on the crack surface affect the closure and local stresses at the crack tip, and embrittlement of the material in front of the crack reduces the energy required to drive the crack, which in turn leads to intergranular or transgranular cracking. Crack propagation along weakened grain boundaries produces excessive crack branching and increases tendencies for non-planar propagation.

Corrosion and other environmental effects are amplified many-fold under surface wear and fretting. Fretting fatigue occurs when contacting surfaces are subjected to oscillating loading. It is responsible for premature failure in components such as spline joints, bolted and riveted connections. Fretting is the synergistic combination of wear, corrosion and fatigue phenomena driven by the partial slip of tribosurfaces, and is attributed to severe reductions in service lifetimes of contacting components. Fretting itself is an extremely complex degradation process consisting of a combination of mechanical and chemical attacks and by its very nature is a multidisciplinary subject (Ferris et al. [42]).
It is known from multiple fractographic observations that some similarities between the corrosion and fretting pits along the fraying surface exist in the geometry and mechanical behavior. For instance: (i) both corrosion and fretting pits act as stress raisers, (ii) both are in multiaxial stress state, (iii) both may originate microcracks under cyclic loading, and (iv) both fretting fatigue and corrosion fatigue have a transition from pitting to fatigue-crack growth, and after the cracks propagate beyond this transition the influence of the pits becomes less important but the bulk stress dominates subsequent crack propagation (Hills and Nowell [43]).

Another important observation is that a fretting scar may catalyze the local corrosion. A further degradation process is a fully coupled fretting-corrosion-
assisted failure. The process of fretting fatigue can be divided conveniently into three stages: (1) the initiation (or nucleation) phase in which damage accumulates at the fretting interface and a crack nuclei is formed; (2) a phase of short-crack growth where the microcrack propagation is determined by the microstructure of the material and/or microscopic conditions at the contact; (3) the long-crack propagation.

Analysis of the initiation process and the subsequent short-crack propagation is the issue of practical importance to the engineering designer. Crack nucleation primarily occurred at a β-phase boundary as has been shown by Anton et al. [44] for Ti-6Al-4V. For this important aerospace material, fretting-fatigue cracking starts from nucleation sites that are mostly surface connected. Sub-surface nucleation has been observed only for small fretting slip (Fig. 18). Initial microcracks propagate from the fretting scar (shown in Fig. 19) at approximately 45 degrees to the surface (shear mode) and finally coalescence. After penetration to about 50 µm, the final crack turns perpendicular to the external applied stress (Fig. 20). A typical surface damage consists of microcracks, surface pitting, and

Figure 19: Fretting scar illustrating heavy fretting surface wear (Photo courtesy of D. Anton.)

Figure 20: Long fretting-fatigue crack. Several short cracks start from surface pits. (Photo courtesy of D. Anton.)
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third-body debris. It is noted in Anton [44] that cracks did not necessarily nucleate from pits, but many surface pits have been observed. On the micrograph in Fig. 20 the traces of $\alpha$-phase can be easily seen. This crystallographic structure plays a similar role as twinning and might initiate subsurface cracking.

Analysis of the fracture morphology of the fretting damage might shed light on the still unanswered question of when and how fretting fatigue initiates and how fast it grows. This is a complex, fully coupled, multiscale physical phenomenon in which the importance of different failure mechanisms is defined by the material microstructure and loading conditions. The minimization of fretting and corrosion damage is an important practical problem in different applications from civil engineering bolts to single-crystal turbine-blade attachments.

7 Fracture morphology of single superalloy crystals

Failure of gas-turbine components is a long-standing and crucial problem affecting all design issues. In high-temperature applications, grain boundaries in polycrystals are susceptible to active creep failure. Also, new deformation mechanisms, such as new activated slip systems, grain growth, grain-boundary sliding, diffusion and oxidation take place (Stouffer and Dame [45]). The general trend in multiple attempts to obtain the best creep resistance was to reduce the number of grain boundaries (i.e. increase grain size). For illustrative purposes, let us choose the thermal-creep capability of the polycrystalline IN100 superalloy with the average grain size of 1 mm for the basis. Then a columnar alloy (for example, PWA 1422) can sustain temperatures of 25°C higher, and Ni superalloy

![Figure 21: Difference in fracture mechanisms after simple tension of Single-crystal PWA1480 specimens with two different crystal orientations (a) [001] and (b) [111].](image-url)
single crystal (for example, PWA 1480), which does not have grain boundaries at all, has material temperature capability of 50°C higher than the basic polycrystalline material. Therefore, the ultimate design goal is to have single-crystal parts operating at critically high temperatures. Currently, blades and vanes of high-temperature turbines are made from single-crystal Ni-based superalloys.

The response of a single crystal to a load is different from that of polycrystals primarily because of highly anisotropic properties that depend on the orientation relative to the crystal lattice. The yield and fracture properties are strong functions of material orientation due to the number and orientations of the prevalent active slip systems, which in turn depend on the microstructure (Swanson and Arakere [46]). Fig. 21 illustrates the dependence of fracture mechanism on crystal orientation. In order to understand the damage and fracture morphology of superalloy single crystals we start with a brief review of its microstructural properties. The material structure is defined by parallel primary dendrites spanning the casting in the direction of solidification [46]. Ni-base superalloy is a two-phase material with hard \( \gamma' \) precipitates based on the intermetallic compound \( \text{Ni}_3\text{Al} \) with \( \text{L}_1_2 \) superlattice and soft \( \gamma \) coherent FCC matrix as shown on the micrograph in Fig. 22. At low strains, only the \( \gamma \) matrix phase undergoes plastic deformation. Therefore, dislocations are concentrated in the narrow matrix channels and surround the \( \gamma' \) precipitates (Allan [48]). At higher strains, the \( \gamma' \) particles undergo shearing.

The crystallographic lattice planes are continuous across \( \gamma' - \gamma \) interfaces (Bassani [47]). At very high temperatures \( \text{Ni}_3\text{Al} \) remains ordered up to almost the melting point, which is responsible for the high-temperature strength and creep resistance of this material. Strength and other material properties are controlled by the size
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and volume fraction of the $\gamma'$ precipitates. The $\gamma'$ cubical participates usually vary in side sizes over the range of 0.3 – 0.5 $\mu$m (DeLuca and Annis [49]). From the micromechanical point of view, $\gamma'$ particles do not allow dislocation bypass, hardening the material and $\gamma$ matrix.

The deformation of a single crystal is the sum of contributions from two independent atomic mechanisms: (i) an overall “elastic” distortion of the lattice, and (ii) a “plastic” deformation due to slip that does not distort the lattice geometry. Dislocation slip in f.c.c. crystals (and in L1$_2$, as well) occurs on the twelve $\{111\}$<110> slip systems listed in Table 2. Note that for a given slip system, slip can occur in either the positive or negative <110> direction in a {111} plane.

Table 2: The twelve $\{111\}$<110> slip systems.

<table>
<thead>
<tr>
<th>(111)</th>
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<th>(T 1)</th>
<th>(T1)</th>
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<tr>
<td>01T[01T][100]</td>
<td>01T[01][100]</td>
<td>01T[01][100]</td>
<td>01T[01][110]</td>
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At temperatures higher than 430°C plastic deformation occurs simultaneously along octahedral slip systems (listed above) and along an additional six cube slip systems {100}<110>. Thus, at high temperatures, plastic deformation suppresses failure modes. The projection of the stress tensor on slip planes along the slip direction defines a scalar called the resolved shear stress. The resolved shear stress (RSS), or so-called Schmid stress, on a slip system defines its potential
activity. If the RSS reaches some critical value then shear strain along the slip system may take place. Of the potentially active systems, only some have a non-zero shear increment at a particular time, and these systems are actually active slip systems. The number of active slip systems depends on the ability of the material to accommodate a prescribed deformation and rarely exceeds two. For particular crystal orientations under tension in the [001] direction, for example a [001] crystal, the number of potentially active slip systems reaches eight and results in patchy slip and multiple cross-slip with a high rate of crystal latent hardening. At high temperatures, the plastic deformation is viscous and some slip takes place even under very low RSS (Ohnami [50]). At low homologous temperatures slip takes place by dislocation loops shearing $\gamma'$ particles. In contrast, at elevated temperatures, high dislocation density in the $\gamma$ phase with mostly $\gamma'$ bypass or looping has been observed (Milligan and Antolovich [51]).

Fatigue is the most typical mode of failure that is observed in single-crystal Ni superalloys as was pointed out in Cowles [52]. Cracks usually start from micropores or from small carbides. At low homologous temperatures ($T<0.4 T_{\text{mel}}$) crystallographic fracture is the dominant mode along \{111\} planes. At higher temperatures starting from approximately 430°C for PWA 1480 [46, 48] some non-crystallographic failure appears. After a small initial region of “non-crystallographic” growth, the cracks turn abruptly and proceed along a single \{111\} plane. With the temperature rise, the probability of non-crystallographic propagation becomes higher. At elevated temperatures, ($T > 0.6 T_{\text{mel}}$) or more specifically after approximately 800°C, cracking is predominantly non-crystallographic, similar to transgranular propagation in a polycrystals as noted in (Telesman and Ghosen [53]).

![Micrograph of fatigue crack in single crystal along \{111\} plane. River marks and radial ledges are shown clearly together with striation marks. (Kindly provided by D. DeLuca.)](image-url)
The antiphase $\gamma/\gamma'$ boundary can act as a crack-initiation site causing interprecipitate fracture. This crack-initiation mechanism typically reveals cubic $<001>$ cracking since the $\gamma/\gamma'$ interface is parallel to the (001) planes. Gas pockets filled with dendrites (Fig. 23) also play the important role of crack-nucleation sites.

At low homologous temperatures, fatigue cracks propagate predominately along {111} crystallographic planes, and failure is driven by the shear mode. Examination of fracture surfaces of PWA 1480 reveals a set of river marks (normal to the crack front) that usually is attributed to coalescence of cleavage steps and is specific for brittle fracture. These steps are formed when different parts of the crack propagate on parallel crystallographic planes and are joined [1] at the advancing crack front. However, Fig. 24 clearly shows river marks together with striation marks, which, as we have already discussed in the section devoted to the fatigue morphology, are specific for plastic failure. This characterizes the specifics of crystallographic shear-fatigue-crack growth. The micrograph in Fig. 25 shows the striation marks of the crystallographic {111} crack crossing the matrix and dendrites. As one can see from the fractography, the geometry of the striation lines and distances between them does not change significantly at dendrite crossings, which supports the conclusion that crack growth rate and crack-propagation mechanisms are the same in the matrix and dendrites.

For structural-life prediction and reliability analysis, one needs to know fatigue-crack growth rates, locations/directions and deformation of material near the crack tip. A fatigue-crack propagation law is the better the closer it reflects the...
cyclic plasticity, the material microstructure and the texture. Multiple observations suggest that a crystallographic fatigue crack belongs to the glide plane but propagates under mixed (111) shear (Mode II) and normal separation (Mode I) action (Milligan and Antolovich [51]). Therefore, the easiest crack propagation should be observed under some mixed-mode loading. Experimental techniques for measurements of such crack propagation in single crystals is based on ring specimens and some test results are described below.

Much work has been done on elaboration of laboratory test methods of fracture-toughness determination for brittle materials (for reviews see [24, 28]). However, reliable test methods that industry needs for the crack initiation in single crystals seem to be missing. Tests should provide information on crack initiation over a wide range of temperatures and environmental conditions. In order to develop design tools for predicting the life of single-crystal blades, there is a need to understand, analytically describe, and experimentally verify crack-initiation mechanisms at different conditions. Brazilian disk (BD) testing has been widely used in civil engineering and geophysics for the last century. Recently, it was proposed to use the same technique for the analysis of single-crystal superalloy material (John et al. [54]). An initial “preflaw” is placed coplanar to the (111) crystallographic plane. The specimen is compressed with “point forces” from the outside diameter. The rotation of the flaw with respect to the loading direction determines the relative components of normal separation and shearing. Obtaining an exactly oriented specimen is critical in evaluating failure results. Also, the proper fracture-mechanics analysis requires the knowledge of the precise expression for the SIF in the anisotropic material, which is not fully available to date. Currently, SIF are being calculated based on a simplified isotropic solution (John et al. [54]), which deviates from the actual stress distribution and shadows the real crack-initiation process. This method provides information regarding relative activities of different slip systems. However, the initial flow causes an additional inhomogeneity in its plane and in the “preflow” direction. Due to near crack-stress concentration the crack-propagation conditions for different angles between loading and crack axis are different, so results for different axes are not exactly correlated with active crystallographic mechanisms of failure (Peters et al. [55]). What is more important, these test results do not provide any information about crack initiation.

A natural expansion of the Brazilian disk idea is to drill a small hole in the center of the disk as it was described above in this work (see tests results in Fig. 12 for monotonic loading) with application to brittle materials. In this case we keep a homogeneous stress field around the hole and may study the process of the crack initiation depending on the specimen crystallographic axes orientation. It is also important that the stress field in such a system can be obtained with high accuracy. The ring sample with outer radius $R_{\text{outer}}$, inner radius $R_{\text{inner}}$, and thickness $h$ sufficient to realize plane-strain conditions is loaded by a pair of point forces $P$ acting along diameter. The central hole, playing the role of the defect, initiates fracture. In such testing, the crack initiation depends only on the following
measurable and controllable parameters: (1) applied load, (2) specimen geometry, and (3) crystallographic orientation.

During our preliminary tests, PWA 1480 showed rate independence at room temperature, which permits us to conduct the test at 30 Hz. A strobe light was used in order to guarantee a stable visual image. To eliminate contact failure we used 3-mm thick aluminum pads. The cyclic load ratio (R-ratio) was 0.1 for all tests. The load varies in the interval from 1000 lb to 2000 lb. We studied the crack initiation on specimens with the following parameters: single-crystal superalloy PWA 1480 with <110> orientation normal to the ring [flat] surface; external radius $R_0=1.1$" and thickness of $t=0.1$". Specimens with central holes of $0.05R_0$, $0.1R_0$, and $0.2R_0$ have been tested at different orientations with respect to the loading direction. The failure might be of two types: crystallographic and non-crystallographic. We observed both types of cracks in our tests. Crystallographic cracks at room temperature might propagate only in the {111} planes. There are only two types of intersection of {111}-type planes with {110} planes of the specimens, namely along $<\bar{1}10>$ and $<1\bar{1}0>$ directions at an angle of 35.26 degrees to the surface and along $<\bar{1}12>$ and $<1\bar{1}2>$ directions perpendicular to the surface, the so-called easy slip direction as illustrated in Fig. 26. Both types of crystallographic cracks have been observed in our tests. Traces of particular crystallographic planes and directions can be easily marked on BD and ring specimens made from a single crystal.

![Figure 26. Crystallographic orientations for the BD specimen with <110> normal (out of plane in the figure). Two possible crystallographic crack directions belonging to {111} crystallographic plane ( $\bar{1}10$ ) and ( $\bar{1}12$ ) have been observed.](image-url)
By rotating the specimen with respect to the loading direction one can investigate the crack initiation and propagation under mixed-mode loading.

Such single-crystal specimens show strong anisotropic fatigue behavior. Loading along different crystallographic axes causes cracking after very different numbers of cycles varying almost an order of magnitude as shown in Fig. 27. The weakest orientation is close to 16 degrees from the $<112>$ crystallographic direction (note that tested ring specimens have a $<110>$ plane normal). These results are in good agreement with test data of [54] for BD with a central crack. Based on this finding, we focused our analysis mostly on testing the weakest orientations. In other words, we tested specimens loaded along the direction of $\approx 16$ degrees from the $<112>$ directions in both the $<001>$ and $<1\bar{1}0>$ directions. We notice that fatigue tests for $<1\bar{1}0>$ orientation have not shown consistent results and a number of tests have been terminated after 8M cycles without noticable crack.

All these test observations suggest that crystallographic cracks start with shear banding and break the continuity, becoming a crack only if normal tensile stress is applied. Thus, we may assume that crystallographic cracks have a mixed-mode crack origin and for most of the crack starting points, $K_{\tau} \approx 2K_{\sigma}$ or normal separation should be approximately half of RSS. These effects are related to (and probably caused by) non-Schmid yield behavior in single-crystal superalloys (Qin and Bassani [56], Vitek et al. [57]), which means that due to the complex
core dislocation structure in Ni$_3$Al the dislocation mobility is controlled not only by the RSS on the primary slip system but also by other shear stress components. Most of the cracks start crystallographically in <112> or <1T0> directions as illustrated in Fig. 28. With the fatigue-crack growth, local stress distribution changes as well as ductile effects accommodate around the crack tips. As can be seen from the micrograph of Fig. 28b and in detail from Fig. 29a, a crystallographic {111}<112> crack has been stopped by excessive shear banding and later turns into a {111}<1T0> crack.

It is important to note that there exists a zone of loading directions (loading along directions approximately between <112> and <001>) which causes non-crystallographic crack initiation. A non-crystallographic crack starts at the position where maximum tension stress takes place. Crack initiation depends on stress concentration around the hole. Cracking of the specimens with a hole diameter of 20% of the outer ones starts easily, but specimens lose stability causing distortion of the fractographic picture and results as shown in Fig. 29b. Empirically, we found that the ring specimen with a hole of 10% of the disk diameter is stable and cracks reasonably fast, and so can be recommended for industrial tests.

Concluding the discussion of single-crystal ring specimen testing, we briefly list some observations: (i) crystallographic cracking of L1$_2$ at room temperature is always in the {111} plane; (ii) typically, both crack directions <112> and <110> take place and compete with each other; (iii) multiple-crack initiation, which leads to growth of a major crack is often observed; (iv) crack initiation starts at

![Figure 28: Fatigue cracks initiated at the central hole of ring specimens.](image)

(a) Specimen was loaded at 23° off <112> to the direction <111>. Two cracks in {111} <112> and in {111} <1T0> are initiated. (b) Specimen has been cracked along <112> but a shear band along <1T0> arrested the crack.
the point of mixed mode $K_{II} + \alpha \cdot K_{I} \rightarrow \text{max}$; $\alpha \sim 1/2$, and (v) the “closer” the loading line to <001>, the easier non-crystallographic crack nucleation.

Fatigue-crack growth in a single crystal is limited to a crystallographic plane, but crack direction is often different from the primary slip directions. As we have seen above, the <112> twinning directions are typical for crystallographic crack growth. One possible explanation of this fact can be deduced from the micrograph in Fig. 30 showing the {111} plane fractography of the crack in a single PWA 1480 crystal. The crack grows symmetrically in the primary slip directions <101> and <011> with the crack fronts piecewise straight and oriented at 60° to each other. The crack fronts intersect in the corners of the hexagon, align with the <112> direction and, thus, the crack growth in <112> is $\frac{2}{\sqrt{3}}$ faster than in the slip direction. In our ring tests we just observed the trace of the fastest-growing direction.

![Figure 29: Plastic shearing accompanying crystallographic cracking.](image)

(a) <110> shear band appears at crack <112> tip and stops the crack. (b) Multiple cracking has been stopped by excessive plastic deformation at high stress concentration.

Similar crystallographic crack-growth behavior has also been observed (Anton [58]) in the cubic {001} planes. In this case, the fatigue crack grew in a square-like fashion, where the side directions were <110> and <110>. Thus, diagonal or the fastest direction was <010>. Anton [58] explained this by material anisotropy and showed that the critical crack-tip-opening displacement (CTOD) reaches its largest values along <010>. As usual, both explanations – morphological and mechanical – are manifestations of the same physical
phenomenon: crack-front-preferred orientation. The morphological study of crack initiation and propagation in a high-symmetry single crystal (Ni-base superalloy) has revealed the importance of crystal orientation and a pronounced effect of dislocation activity and temperature regime on the failure mechanism and fatigue life.

8 Conclusion

An understanding of damage and failure mechanisms is fundamental to good design of high-performance structural parts. The examination of the fracture surfaces and materials in-situ damage assessment is an important tool in mechanics of materials in order to reveal a significant correlation of strength, damage and failure with microstructure. Proper materials requirements are deduced from these experimental observations. They allow us to make the next step to understanding and developing of physics-based fracture models in complex microstructures.

These quantitative observations could categorize morphological parameters of several damage mechanisms typically acting simultaneously. Damage and fracture-morphology studies allows bridging of different length scales and proper
implementation of microstructural characteristics into macroscale constitutive laws, which is the important current problem of computational mechanics.

In this chapter we have studied and reviewed different damage mechanisms. Crack initiation, growth and branching in brittle and elastic-plastic materials have been analyzed and related to loading and microstructure, improving our ability to predict crack behavior. Synergism of different degradation mechanisms of fretting and pitting corrosion has been reviewed and it was found that the failure mechanisms were material specific. We have paid much attention to the fracture morphology in single crystals of high symmetry, especially superalloys. New testing approaches effective in both macroscale and single crystals analysis have been proposed and details of cracking have been reported. We have also analyzed specifics of damage evolution and failure analysis in single crystals needed for lifetime predictions to improve turbine-blade design and reliability.

Acknowledgement

The author would like to express his sincere gratitude to Daniel DeLuca, Pratt & Whitney Fellow with whom I have had the privilege to work on single-crystal superalloy failure and who is a true artist of microphotography. I also wish to thank Leroy (Corky) Favrow (Materials Lab. Pratt & Whitney), Donald Anton (UTRC) for their help and for sharing their materials with me, and Daniel Mosher (UTRC) for fruitful discussions and help in manuscript preparation.

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