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Corrosive Wear and Fretting

► Fretting and Tribocorrosive Testing

Cottonseed Oil

► Natural Oils as Lubricants

Coulomb's Law

► Friction Coefficient
► Friction (Concepts)

Counterpart Rack

► Rack

Crack Growth in Brittle and Ductile Solids

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Synonyms

CTOD – crack tip opening displacement; DBTT – ductile-to-brittle transition temperature; LEFM – linear-elastic fracture mechanics; PSZ – partially stabilized zirconia

Definition

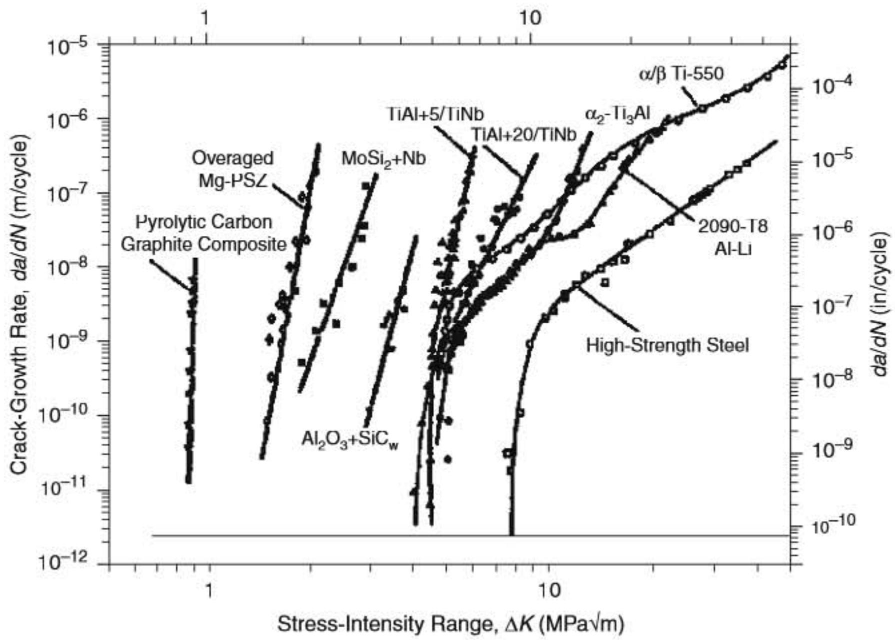
Crack growth in brittle and ductile solids pertains to the extension of incipient cracks or flaws in materials that either show limited (brittle) or extensive (ductile) plasticity prior to fracture.

Scientific Fundamentals

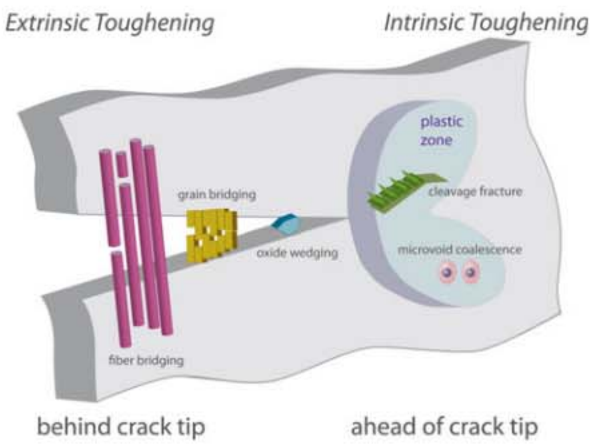
The quest for strong, fracture-resistant structural materials remains a persistent challenge in materials science. Although nominally ductile metallic materials such as steels, aluminum, titanium, and nickel-base superalloys are still used for most cutting-edge structural applications, there is a continued motivation to utilize higher strength, higher temperature materials such as ceramics and intermetallics for improved performance. Unfortunately, the widespread adoption of these materials has been severely limited by their low damage-tolerance, that is, they are brittle with generally low ductility and poor fracture toughness. In addition, it is now appreciated that these materials, like metals, are susceptible to premature failure due to the cycle- and time-dependent growth of incipient cracks.

Damage-tolerance is a material property that characterizes resistance to crack growth, the principal mechanism of which is fatigue. Fatigue involves the microstructural damage and failure of materials under cyclically varying loads. In fact, it is generally considered that more than 80% of all service failures can be traced to mechanical fatigue, whether in association with cyclic plasticity, sliding or physical contact (fretting and rolling contact fatigue), environmental damage (corrosion fatigue), or elevated temperatures (creep fatigue).

Here, the nature of crack growth, primarily by fatigue, is described for both ductile and brittle materials. It is shown that the mechanisms involved in the propagation of cracks in brittle ceramic and intermetallic materials are quite distinct from those commonly encountered in ductile metals. In particular, their crack growth rate (da/dN) behavior displays a markedly higher sensitivity to the applied stress-intensity factor, K ($K = Q\sigma(\pi a)^{1/2}$, where Q is a geometry factor of order unity, σ is the applied stress, and a is the crack size, than is observed in most metals (Fig. 1) (Ritchie 1988). However, by considering crack growth as a mutual competition between intrinsic microstructural damage mechanisms, which promote crack extension ahead of the tip, and extrinsic crack-tip shielding mechanisms, which act primarily behind the tip and retard crack growth (Ritchie 1988), a specific commonality of behavior between the crack-growth behavior of ductile and brittle materials can be found, differing only in the relative importance of the intrinsic and extrinsic mechanisms.



Crack Growth in Brittle and Ductile Solids, Fig. 1 Schematic variation of fatigue-crack propagation rate (*da/dN*) with applied stress intensity range (ΔK), for ductile (metals) and brittle (intermetallics and ceramics) materials



Crack Growth in Brittle and Ductile Solids, Fig. 2 Schematic illustration of mutual competition between intrinsic mechanisms of damage/crack advance and extrinsic mechanisms of crack-tip shielding involved in crack growth

Intrinsic and Extrinsic Mechanisms

The extension of a crack can be considered to be a result of the mutual competition of two classes of mechanisms (Fig. 2). Crack growth is promoted ahead of the crack tip by intrinsic microstructural damage mechanisms and impeded by extrinsic toughening mechanisms acting primarily behind the crack tip, which serve to “shield” the crack tip from the far-field driving force (Ritchie 1988). Intrinsic damage mechanisms in metallic materials typically involve processes that create microcracks or voids (e.g., by dislocation pile-ups or interface decohesion) in the highly stressed region ahead of the tip, leading to classical failure by cleavage, intergranular cracking, or microvoid coalescence; comparable mechanisms under cyclic loads involve the repetitive blunting and resharpening of the crack tip. Extrinsic shielding mechanisms, conversely, result from the creation of inelastic zones surrounding the crack wake or from physical contact between the crack surfaces via wedging, bridging, sliding, or combinations thereof (Fig. 2) (Ritchie 1988). Examples of *zone shielding* are transformation and microcrack toughening in ceramics and rocks (e.g., McMeeking and Evans 1982), where the in situ dilatant phase transformations, or the microcracking of precipitates/particles ahead of the crack tip, can lead to inelastic zones in the crack wake that impart closing tractions on the crack surfaces. Examples of *contact shielding* are the bridging tractions imposed across a crack by unbroken fibers, laminated layers, or a particulate phase in composite materials (e.g., Evans 1990), or the wedging of corrosion debris or fracture surface asperities during crack closure in metal fatigue (Suresh et al. 1981).

Intrinsic mechanisms are an inherent property of the material, and thus are active irrespective of the length of the crack or the geometry of the test specimen; under monotonic loads, for example, they control the driving forces (e.g., the stress intensity) to initiate cracking. Extrinsic mechanisms, conversely, act primarily in the crack wake and are thus critically dependent on crack size (or more correctly the amount of crack extension) and (to a lesser extent) on geometry; they are responsible for the development of crack resistance-curve (R-curve) behavior under monotonic loading, and thus play a prominent role in the driving forces required for continued growth of the crack. The implications of this are that where extrinsic shielding mechanisms are active, rising R-curve toughness behavior and small-crack effects are to be expected, both phenomena resulting from the crack-growth properties being dependent upon crack size. Moreover, since extrinsic mechanisms can have no effect on crack initiation (because there is no crack wake), the microstructural factors affecting (large) crack growth may be quite different from those affecting crack initiation (or small crack growth).

In general, ductile materials are toughened *intrinsically* (e.g., through mobile dislocation activity) to induce a significant plastic-zone size, although under cyclic loading extrinsic mechanisms play an important role in the form of crack closure. In contrast, brittle materials are invariably toughened *extrinsically* (e.g., Evans 1990), via such mechanisms as transformation toughening and crack bridging, the latter through interlocking grains in many monolithic ceramics, or by uncracked ligaments or unbroken reinforcement phases in composites, laminates, and many biological materials such as bone and dentin.

In terms of finding any commonality in mechanisms of crack growth in different materials, it is the specific nature and, more significantly, the relative importance of the intrinsic (damage) vs. extrinsic (shielding) mechanisms that distinguish the cracking behavior of ductile and brittle solids. For fatigue, this in turn governs the specific dependencies of the alternating and maximum stress intensities on crack-growth rates, that is, how da/dN depends upon ΔK and K_{max} (and thus how the resulting lifetime is a function of the alternating or maximum stresses), and the relationships between the thresholds for crack growth (ΔK_{th} and $K_{max,th}$) and the crack-initiation (K_o) and steady-state crack growth (K_c) fracture toughness values (Ritchie 1988). Specific mechanics and mechanisms affecting crack growth in ductile and brittle materials are described below.

Crack Growth in Ductile Materials

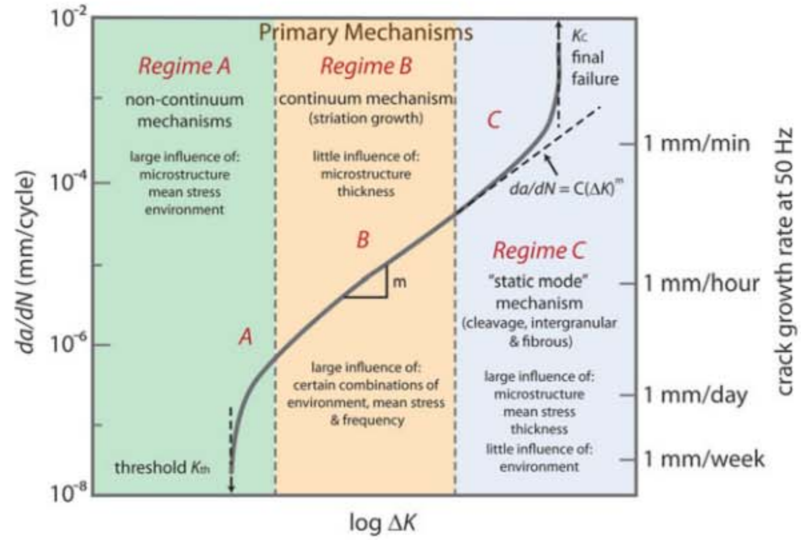
Fatigue-crack growth can occur at stress intensity K levels generally far less than the fracture toughness, K_c, in any metallic alloy when cyclic loading is applied ($\Delta K_{th}/K_c \sim 0.1$–$0.4$). Stated simply, the accumulation of damage from the cyclic plastic deformation in the plastic zone at the crack tip accounts for the intrinsic mechanism of crack growth at K levels below K_c. The process of failure itself consists of several processes involving initial damage (e.g., cyclic hardening or softening), formation of an initial "fatal" flaw (crack initiation), macroscopic propagation of this flaw (crack growth), and final catastrophic failure or instability.

Although the problem of fatigue failure was first seriously considered in the mid-nineteenth century when widespread failures of railcar axles in Europe circa 1860 prompted the first systematic investigations into the topic (Wöhler 1860), the main impetus for research on crack propagation did not occur until the mid-1960s, when the concepts of linear-elastic fracture mechanics (LEFM) and damage-tolerant design were first applied to the problem of subcritical crack growth. Such approaches recognize that all structures are flawed and that cracks may initiate early in service life and propagate subcritically. Lifetime is then assessed on the basis of the time or number of loading cycles for the largest undetected crack to grow to failure, as might be defined by an allowable strain, or limit load, or K_c criterion. Implicit in such analyses is that subcritical crack growth can be characterized in terms of some governing parameter (i.e., a crack-driving force) that describes local conditions at the crack tip, yet may be determined in terms of loading parameters, crack size, and geometry. Linear-elastic and nonlinear-elastic fracture mechanics have, to date, provided the most appropriate methodology for such analyses to be made.

The general aspects of fatigue-crack growth in ductile (metallic materials) and its description using fracture mechanics can be briefly summarized by the schematic diagram in Fig. 3, showing the variation in da/dN with the nominal stress-intensity range ($\Delta K = K_{max} - K_{min}$). In actuality, the growth rates depend upon numerous factors other than ΔK, although this is the primary variable in metal fatigue, viz.:

$$da/dN = f[\Delta K, K(\text{or } R), \nu, \text{environment}, \text{wave form} \ldots], \tag{1}$$

where the load ratio R is the ratio of minimum to maximum applied loads ($= K_{min}/K_{max}$ for positive R), and ν is the cyclic frequency. Specifically, results of fatigue-crack growth rate tests for most ductile materials display the



Crack Growth in Brittle and Ductile Solids, Fig. 3 Schematic illustration of the typical variation in fatigue-crack growth rates, *da/dN*, as a function of the applied stress-intensity range, ΔK, in metallic materials, showing the regimes of primary growth-rate mechanisms and effects of several major variables on crack-growth behavior

following characteristics: (a) a region at low values of ΔK and da/dN (less than $\sim 10^{-9}$ m/cycle) in which fatigue cracks appear dormant below the fatigue thresholds, ΔK_{TH} and $K_{max,TH}$; (b) an intermediate region ($\sim 10^{-9}$ to 10^{-6} m/cycle) of power-law behavior nominally described by the Paris equation :

$$da/dN = C(\Delta K)^{m}, \tag{2}$$

where C and m ($\sim$2–4) are material scaling constants; and (c) an upper region of accelerating crack growth (above $\sim 10^{-6}$ m/cycle) as K_{max} approaches K_c or gross plastic deformation (limit load) occurs. Similar approaches have been suggested for crack growth under large-scale yielding, where growth rates have been related to a cyclic J-integral (ΔJ) or range of crack-tip opening displacement ($\Delta CTOD$).

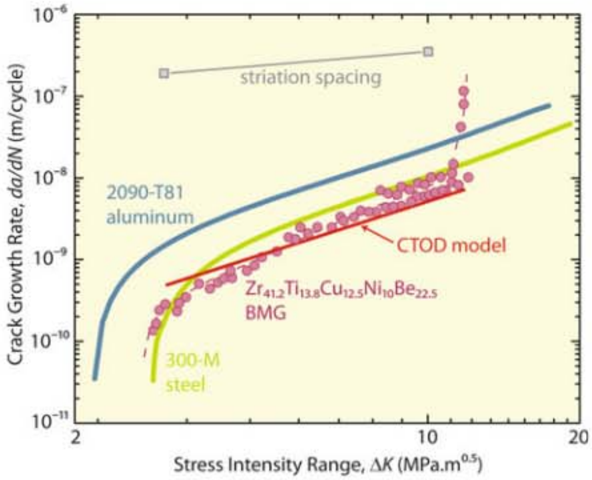
Intrinsic mechanisms: Ideally, crack growth by fatigue in ductile materials, such as polymers and metals, can be characterized by a mechanism of transgranular ductile striations, although this mode often is difficult to identify owing to the nature of the underlying microstructure or the occurrence of other cracking modes. Striations represent the result of crack-growth increments each cycle, and have been hypothesized to occur via a mechanism of opening and blunting of the crack tip on loading, followed by resharpening of the tip on unloading. Several theoretical models for such growth (often termed stage II crack propagation) have been proposed, relying on the fact that, where plastic zones are sufficiently large compared with microstructural dimensions, plastic blunting at the crack tip is accommodated by shear on two slip-systems at roughly 45° to the crack plane. Recognizing that such sliding-off is largely irreversible, new crack surface can be created during cyclic crack advance either by simultaneous or alternating slip on these two systems. This damage process is the primary intrinsic mechanism promoting crack advance.

Simple models for striation formation predict that an upper-bound estimate for the increment of crack advance per cycle should be proportional to the cyclic crack-tip opening displacement ($\Delta CTOD$):

$$\frac{da}{dN} \propto \Delta\, CTOD \approx \beta \frac{\Delta K^2}{2\sigma_o E'}, \tag{3}$$

where σ_o and E' are, respectively, the appropriate flow stress and Young's modulus, and β is a proportionality constant, of order 0.1–0.5, reflecting the efficiency of crack-tip blunting and reversibility of slip.

This approach provides a first-order description of crack-growth rate behavior in the mid-range of growth rates (regime B in Fig. 3), as shown for example by recent results on a bulk amorphous Zr–Ti–Cu–Ni–Be metal where crack advance occurs by a striation mechanism (Fig. 4) (Gilbert et al. 1999), although it is an insufficient description at high growth rates and in the near-threshold regime. At high growth rates, as $K_{max} \rightarrow K_c$ (regime C in



Crack Growth in Brittle and Ductile Solids, Fig. 4 Fatigue-crack propagation in a bulk amorphous metal (metallic glass), $Zr_{41.2}Ti_{13.8}Cu_{12.5}Ni_{10}Be_{22.5}$, showing fatigue-crack propagation rates scaling with the $\Delta CTOD = 0{:}01\ \Delta K^2/\sigma_o E'$ (where σ_o is the flow stress and E' is the appropriate Young's modulus) and corresponding striation spacings. Also shown for comparison are growth-rate results for polycrystalline metals, namely 300-M high-strength steel and 2090-T81 aluminum–lithium alloy (Reproduced from Gilbert et al. 1999)

Fig. 3), the simple Paris law (2) underestimates measured growth rates due to the occurrence of monotonic fracture mechanisms (static modes) which replace, or are additional to, striation growth. Such mechanisms include cleavage, intergranular cracking, and microvoid coalescence, and their presence results in growth rates that are markedly sensitive to microstructure and K_{max} (or R). Conversely, at very low growth rates, where $\Delta K \rightarrow \Delta K_{th}$ (regime A in Fig. 3), the Paris law overestimates measured growth rates and behavior becomes markedly sensitive to K_{max}, microstructure, and environment; the K_{max}-dependence in this regime, however, results primarily from crack closure. At such near-threshold levels, the scale of local plasticity (i.e., the plastic-zone size, r_y) approaches microstructural size-scales, and measured growth rates become less than an interatomic spacing per cycle, indicating that crack advance is not occurring uniformly over the entire crack front. Crack-growth mechanisms in this regime (typically where r_y is smaller than the grain size) generally are faceted, often being referred to as *microstructurally sensitive* or *crystallographic* fatigue, and reflect more of a single shear mode of crack advance with associated mode II plus mode I displacements, particularly in coarse planar-slip materials.

Extrinsic mechanisms: Although the primary mechanism motivating fatigue-crack extension in ductile materials (i.e., crack-tip blunting and resharpening) is intrinsic and controlled principally by ΔK (or more precisely the local plastic strain range), extrinsic crack closure mechanisms act in the crack wake to oppose this. Such wedge shielding (Ritchie 1988) results from local deformation, fracture and chemical processes that induce physical contact between the mating crack surfaces at positive loads during the fatigue cycle. Elber originally proposed that closure arises from the constraint of surrounding elastic material on the residual stretch in material elements previously plastically strained at the tip (plasticity-induced closure) (Elber 1970). Since the crack cannot propagate while it remains closed, the net effect is to reduce the nominal (applied) ΔK value to some lower effective (local) value ΔK_{eff} actually experienced at the crack tip:

$$\begin{aligned} \Delta K_{eff} &= K_{max} - K_{cl}, \quad (K_{min} \leq K_{cl}) \\ &= K_{max} - K_{min}, \quad (K_{cl} \leq K_{min}), \end{aligned} \tag{4}$$

where K_{cl} is the stress intensity to close the crack. There are, however, several mechanisms of closure that assume greater importance at near-threshold levels, where *CTOD*s are small and approach the dimensions of the wedge. These processes rely on wedging mechanisms inside the crack from corrosion debris, fracture surface asperities, or, in the case of environmentally assisted fatigue, fluid inside the crack, as reviewed by Suresh and Ritchie (1984).

Crack closure arising from crack-surface corrosion deposits (oxide-induced closure) is promoted in oxidizing environments at low load ratios. Notable examples are the crack surface oxides and calcareous deposits formed during corrosion fatigue in structural steels tested, respectively, in water and sea water, and the chromic oxides formed during creep fatigue in Ni-based superalloys. Simple quantitative modeling, based on the concept of a rigid wedge inside a linear-elastic crack, suggests that such closure depends upon the thickness d of the oxide film and the location of its peak thickness from the crack tip $2z$ (Suresh and Ritchie 1984):

$$K_{cl} \approx \frac{E' d}{4\sqrt{\pi z}}, \tag{5}$$

implying that deposits in the immediate vicinity of the crack tip have a dominating influence in the development of closure. A more general source arises from the wedging action of fracture surface asperities, where *CTOD*s are small and where significant crack-tip shear displacements occur. Such roughness-induced closure is promoted at near-threshold levels, particularly where crack advance is strongly crystallographic—as in coherent-particle-hardened (planar-slip) systems such as underaged Al alloys and Ni-based superalloys or in duplex microstructures where the crack can be made to meander from frequent crack deflection. Notable examples of where crack deflection (around microstructural features) and the resulting closure have led to excellent crack growth properties are found with dual-phase steels, β-annealed Ti alloys, and Al–Li alloys. The magnitude of the roughness-induced mechanism depends upon the degree of fracture surface roughness and the extent of the mode II crack-tip displacements. From simple two-dimensional geometric modeling, the closure stress intensity at the point of first asperity contact is given (Suresh and Ritchie 1984) as

$$K_{cl} \approx \sqrt{\frac{2\gamma u}{1+2\gamma u}} K_{\max}, \tag{6}$$

where γ is a measure of surface roughness, that is, ratio of height to width of the asperities, and u is the ratio of mode II to mode I displacements.

Because crack-closure mechanisms involve a wedging process that effectively raises the minimum stress intensity, the potency of the shielding depends upon the magnitude of the crack-tip opening displacement compared with the dimension of the wedge. Since *CTOD*s are much larger at high R, the closure effect is markedly diminished at high load ratios, that is, it is controlled primarily by the value of $K_{\max}$, particularly at lower ΔK levels, where closure results predominantly from oxide or asperity wedging. Thus, in the fatigue of ductile materials, it is the ΔK component of the driving force that controls the intrinsic damage mechanisms ahead of the crack tip, whereas the $K_{\max}$-component controls the extrinsic shielding mechanisms behind the tip. The exception to this is behavior at high stress intensities as $K_{\max}$ approaches K_c or the limit-load instability (Regime C in Fig. 3), where additional static fracture mechanisms (e.g., cleavage, microvoid coalescence) occur; these mechanisms are primarily $K_{\max}$ controlled.

Crack Growth in Brittle Materials

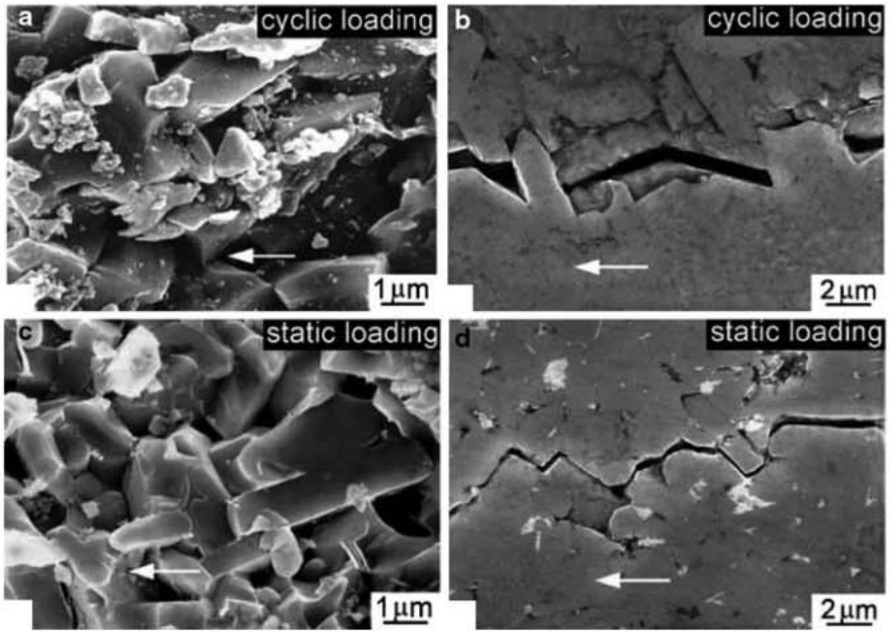
By designing microstructures that develop such shielding processes, monolithic and composite ceramics can now be processed with toughness up to an order of magnitude higher than was available 25 or so years ago. It is perhaps ironic, however, that whereas glasses and untoughened ceramics are essentially immune to cyclic fatigue, the generation of a nonlinear stress–strain curve in toughened ceramics results in their increased susceptibility to fatigue failure under cyclic loading. The characteristics of cyclic fatigue in ceramics, however, appear to be quite different from those of metal fatigue:

(a) Unlike in ductile materials, and with the exception of phase-transforming ceramics such as PSZ, fatigue cracks in ceramics do not appear to initiate naturally; crack initiation is invariably associated with some pre-existing defect.
(b) Again, unlike metal fatigue where there is a characteristic fracture mode for cyclic loading (i.e., striation growth), which is quite distinct from that for monotonic loading, the morphology of fatigue fracture surfaces in ceramics is almost identical to that under monotonic loads, although more debris is often present on the fatigue fractures. Indeed, the crack path and crack-advance mechanism appear to be identical under static and cyclic loading (Fig. 5).
(c) Akin to metals, microstructure can have a marked effect on fatigue-crack growth rates in ceramics. However, if the growth-rate data are normalized with respect to the fracture toughness (i.e., by characterizing da/dN in terms of $K_{\max}/K_c$), the microstructural effects are essentially removed. This implies that, unlike metals, the microstructural influences on fracture by fatigue are similar to those for overload fracture.
(d) The sensitivity of growth rates to the stress intensity is extremely high; specifically, the exponent m in the simple Paris equation (2) can take values as high as ~15–50 and above, as shown by the growth-rate curves for alumina, PSZ, and pyrolytic carbon in Fig. 1. The very high exponents, however, result from a particularly marked sensitivity of growth rates to $K_{\max}$, rather than ΔK per se. This can be appreciated by expressing the growth-rate relationship in terms of both $K_{\max}$ and ΔK (Dauskardt et al. 1992), viz.:

$$da/dN = C'(K_{\max})^n (\Delta K)^p, \tag{7}$$

where, compared to (2), C' is a constant equal to $C(1-R)^n$ and $(n+p)=m$. In a typical brittle ceramic, for example, SiC–whisker reinforced Al_2O_3, the exponents n and p are ~10 and 5 (Dauskardt et al. 1992), respectively; this is to be compared with values of $n=0.5$ and $p=3$ for metal fatigue of a nickel-base superalloy (Van Stone 1988).

Mechanistic aspects: Mechanistically, it is clear that fatigue-crack advance in ceramics is conceptually different from that in metals. Again, we can involve the intrinsic versus extrinsic concept to explain this distinction. Essentially, there are two possible classes of

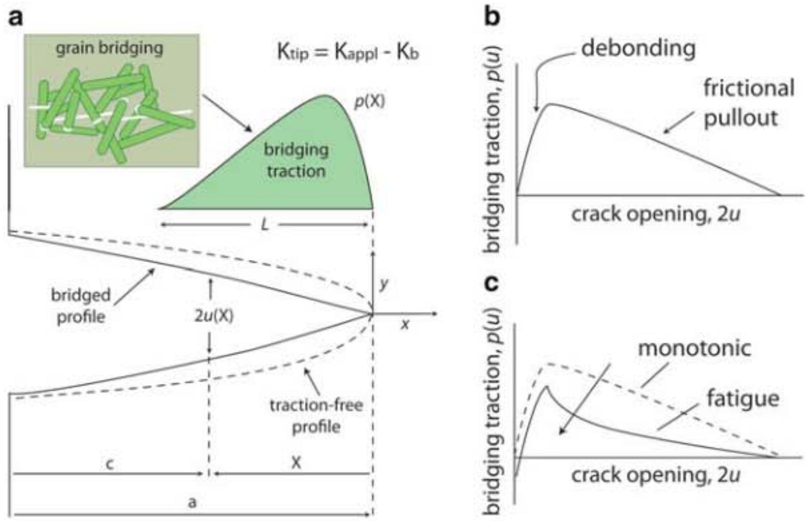


Crack Growth in Brittle and Ductile Solids, Fig. 5 Fractography of ceramic fatigue in an in situ-toughened silicon carbide (ABC-SiC) at ambient temperatures, showing the fractography and crack trajectories under (**a, b**) cyclic and (**c, d**) static loading. Note the nominally identical intergranular fracture surfaces in (**a**) fatigue and (**c**) on the R-curve, and associated crack-tip shielding by grain bridging (**b, d**) in the crack wake. Due to repetitive opening and closing of the crack, such bridging leads to the creation of more debris and surface damage on the fatigue surfaces in fatigue. *Horizontal arrow* represents direction of crack growth

fatigue mechanisms (where failure is associated with a dominant crack):

(a) Intrinsic mechanisms where, as in metals, crack advance results from damage processes in the crack-tip region, which are unique to cyclic loading.
(b) Extrinsic mechanisms, where the crack-advance mechanism ahead of the crack tip is identical to that for monotonic loading, but the unloading cycle promotes accelerated crack growth by degrading the degree of crack-tip shielding behind the tip (Ritchie 1988).

Whereas the cyclic processes in metal fatigue are predominantly intrinsic in nature, the cyclic fatigue processes in ceramics are extrinsic. The mechanism by which the crack advances is thus identical under cyclic loading to what it would be in a single overload cycle; this clearly is consistent with the marked dependency of growth rates on K_{max} (rather than on ΔK), and a similarity in fracture surface appearance under cyclic and monotonic loading. The cyclic loading conversely acts to diminish the shielding (i.e., the R-curve toughening) in the crack wake. This can take various forms, depending upon the prevailing shielding mechanism; examples include the premature fatigue failure of the metallic reinforcement phase in ductile-phase toughened materials (Rao et al. 1992); a reduction in the effect of fiber bridging in fiber-reinforced ceramic-matrix composites (Rouby and Reynaud 1993); and a decreased bridging capacity of a wake zone of interlocking grains in grain-bridging ceramics, such as coarse-grained Al_2O_3, grain-elongated Si_3N_4, and in situ toughened SiC (Gilbert and Ritchie 1998, Dauskardt 1993). The latter mechanism is particularly relevant to monolithic ceramics where the decay in crack bridging under cyclic loading can be associated with the relative motion of the grains as the crack proceeds intergranularly (Fig. 5); this can be modeled in terms of cycle-dependent sliding wear degradation of frictional grain bridges (Dauskardt 1993). Specifically, the stress in the grain bridges rises rapidly with displacement during debonding of the grain, followed by a gradual decrease where frictional pullout of the grain occurs (Fig. 6); both processes contribute to the toughness, although in general most energy is dissipated during frictional pullout. Under cyclic loads, the repetitive opening and closing of the crack results in a decrease in the toughening capacity of the bridging zone by reducing the grain-bridging stress, that is, accumulated damage at the grain/matrix interface

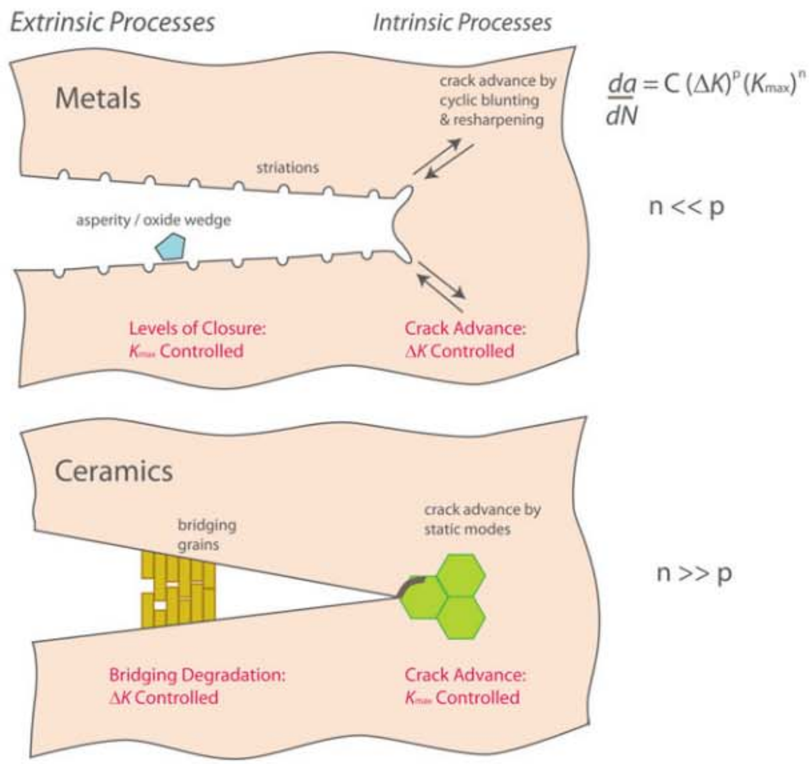


Crack Growth in Brittle and Ductile Solids, Fig. 6 (**a**) Crack-tip shielding from bridging by interlocking grains (inset), showing (**b**) the variation in bridging tractions *p(X)* due to debonding and frictional pullout, and the corresponding crack-opening profile as a function of the crack-opening displacement (2*u*). (**c**) Under cyclic loading, the bridging is observed to decay (i.e., *p(u)* is progressively decreased)

ahead of the tip and reduced frictional sliding resistance of partially debonded grain/matrix interfaces from frictional wear causes premature grain debonding and a reduced frictional pullout stress. Such reduced crack-tip shielding under cyclic loads can be verified by measuring the crack-opening profiles of cracks grown under monotonic and cyclic loading; such experiments show how cracks tend to open more under cyclically varying loads to approach their traction-free profile (Gilbert and Ritchie 1998).

Intermetallics: Like ceramics, intermetallics generally display only very restricted mobile dislocation activity at low homologous temperatures (below their DBTT) due to their complex, and invariably ordered, crystal structures; consequently, they generally have only limited ductility and poor resistance to crack growth. To toughen intermetallics, both intrinsic and extrinsic toughening approaches have been attempted, and to varying degrees have been successful. For example, the intermetallic compound Nb_3Al can be toughened through the addition of a ductile phase such as Nb, with the intermetallic matrix cracking preferentially to the metallic reinforcement (Bloyer et al. 1998). In fact, using a high-aspect-ratio reinforcement (e.g., a Nb layer in an arrester laminate), the crack-initiation toughness may be enhanced intrinsically by the necessity of the crack in the intermetallic matrix renucleating across the metal phase. With subsequent crack extension, the uncracked Nb ligaments left in the crack wake can induce additional crack-growth (or R-curve) toughness, in this case extrinsically from the crack bridging created by the zone of intact ductile ligaments behind the crack tip.

In many respects, the mechanical behavior of intermetallics can be considered to be intermediate between metals and ceramics, ranging from compounds with some ductility, such as α_2-Ti_3Al, that display metal-like characteristics, to the very brittle Nb_3Al and $MoSi_2$ that are ceramic-like (below the DBTT). Specifically, with respect to crack growth, intrinsic damage mechanisms associated with crack advance do appear to operate in the more ductile intermetallics, as in metals; in contrast, the mechanism by which the crack extends in the cyclic fatigue of brittle intermetallics, such as Ni_3Al and $MoSi_2$, remains intergranular or transgranular cleavage and thus is identical to that occurring under monotonic loads, as in ceramics (e.g., Rao et al. 1992). Moreover, whereas the intrinsic toughening mechanisms, such as crack renucleation, do not degrade under cyclic loading, the extrinsic toughening mechanisms, such as crack bridging, can suffer severe degradation. A notable example is ductile-phase reinforced intermetallic composites such as Nb-reinforced γ-TiAl, which due to extensive crack bridging by the uncracked ductile phase can display significantly higher toughness (by a factor of three or greater) than the constituent matrix (e.g., Rao et al. 1992). However, the



Crack Growth in Brittle and Ductile Solids, Fig. 7 Schematic illustrations of the intrinsic and extrinsic mechanisms involved in cyclic fatigue-crack growth in metals, and ceramics, showing the relative dependencies of growth rates, *da/dN*, on the alternating, ΔK, and maximum, K_{max}, stress intensities

improvement in crack-growth resistance is far less obvious in fatigue, simply because the ductile phase fails prematurely; indeed, the fatigue-crack growth properties are rarely much better than that of the unreinforced matrix. Thus, similar to ceramic materials, cyclic loading can act to reduce the potency of extrinsic toughening mechanisms in impeding crack advance, which makes these materials susceptible to premature fatigue failures.

Differences and Similarities between Crack Growth in Ductile and Brittle Materials

Compared with metallic materials, where there is an extensive database and degree of understanding of crack growth and fatigue failure (e.g., Davidson and Lankford 1992), there is far less information on the cyclic crack-growth resistance of ceramics and intermetallics. As noted, it is now clear that cyclic loading induces a progressive degradation in the toughening (or shielding) mechanisms behind the crack tip, leading to a local elevation of the near-tip driving force. It is this cyclic suppression of shielding that is considered to be the principal source for the susceptibility of brittle materials to cyclic fatigue failure. By contrast, the propagation of fatigue cracks in metallic materials involves primarily intrinsic damage processes occurring ahead of the crack tip, that is, involving crack advance by progressive blunting and resharpening of the crack tip; this is clearly a mechanism that is distinct from fracture under monotonic loads. Additionally, shielding, in the form of crack closure (wedging) mechanisms, can act in the crack wake. Since the physical mechanisms of crack advance and crack-tip shielding are quite different in ductile and brittle solids, the dependencies on the alternating and mean loads, specifically ΔK and K_{max}, are also quite different.

A schematic illustration highlighting these differences is shown in Fig. 7. In ductile metals, the dominant dependence of ΔK is a consequence of the intrinsic crack-advance mechanism; the smaller K_{max} dependence results primarily from its effect of the crack-opening displacement, which in turn controls the degree of crack wedging due to crack closure in the wake. Thus, for ductile metals $p \gg n$ in (7). In brittle ceramics, conversely, growth rates are principally a function of K_{max}, since the crack-advance mechanism is identical to that under static loading; the much weaker ΔK dependence here arises from the cyclic-induced degradation in shielding in the wake. Thus, for ceramics $n \gg p$. In intermetallics, fatigue properties are intermediate between these two extremes, such that generally $n \sim p$. However, in very brittle intermetallics such as $MoSi_2$, where there is no intrinsic cycle-dependent crack-advance mechanism, $n > p$, whereas in more ductile materials, such as the $(\gamma + \alpha_2)$ TiAl alloys, intrinsic fatigue damage mechanisms, similar to those in metals, clearly exist, and $p > n$.

Key Applications

Design and Life Prediction

Information on crack growth is used primarily for the design of new higher toughness materials and most importantly for the prediction of lifetimes using damage-tolerant design concepts. For metallic structures, the latter "lifing" procedures are well established for safety-critical applications and rely on the integration of data relating crack-growth rates (da/dN or da/dt) to the applied stress intensity (ΔK or K_{max}) in order to estimate the time or number of cycles, N_f, to grow the largest undetectable initial flaw, a_i, to critical size a_c, viz:

$$N_f = \frac{2}{(m-2)\pi^{\frac{m}{2}} C(Q\Delta\sigma)^m} \left[a_o^{(2-m)/2} - a_c^{(2-m)/2} \right], \quad (8)$$

where $\Delta\sigma$ is the applied stress range, Q is the geometric factor in the K-solution, C and m ($\neq 2$) are the scaling constants in the crack-growth relationship (e.g., (2)).

The use of such life-prediction methods for structural components fabricated from brittle materials, however, is complicated by the marked sensitivity of their crack-growth rates to the stress intensity; specifically the approach can be difficult to implement in practice due to the large values of the exponent m. Since the projected life is proportional to the reciprocal of the applied stress raised to the power of m, a change in applied stress by, for example, a factor of two can lead to life projections that vary by more than six orders of magnitude. Essentially, because of the high exponents, the life spent in crack propagation in brittle materials is extremely limited or infinitely large, depending upon whether the initial stress intensity is above or below the fatigue threshold. An alternative approach is to design on the basis of a threshold for no crack growth, for example, the ΔK_{th} or $K_{max,th}$ thresholds, although even these approaches are not always conservative due to uncertainties in the definition of their values in brittle solids.

Cross-References

► Crack Growth in Noncrystalline Solids
► Crack Initiation in Brittle Solids
► Cyclic Loading and Cyclic Stress
► Temperature Effect on Fatigue
► Fatigue
► Rolling Contact Fatigue in Rail – Insulated Rail Joints (IRJ)
► Stress Intensity Factors
► Stress-Life Theories

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