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ASM Handbook, Volume 11, *Failure Analysis and Prevention*
B. Miller, R. Shipley, R. Parrington, D. Dennies, editors
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Fracture Appearance and Mechanisms of Deformation and Fracture

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FAILURE is most commonly defined within the context of engineering components and structures as an inability to perform the intended function. Failures manifest through many mechanisms but are most often associated with fracture in one or more forms. Interrogation, or investigation, of fracture patterns and fracture surfaces is often a critical step in conducting a failure investigation, even when the fracture mechanism itself may not be difficult to identify or particularly meaningful in the context of the broader investigation. A good example of such an occurrence is the single-event overstress fracture of a component. Although the fracture mechanism (overstress) may be obvious to investigators, critical information necessary to the broader understanding of the failure sequence may lie within the interrogation of the physical evidence. Magnitude and direction of loading, interaction with other components, damage sequencing in systems, and other valuable information may be reliant upon the investigation and reconstruction of fracture patterns. In this way, inspection and analysis of fracture surfaces may be critical in the overall failure analysis even when the component studied does not exhibit any deficiencies but rather has been subjected to circumstances well beyond its intended function. Automobile and aircraft accidents are some examples of circumstances where such fracture investigations may be critical to understand a broader engineering failure.

Note that some non-fracture-failure scenarios may ultimately lead to fracture. Wear processes, for example, can ultimately lead to fracture by galling and/or fretting fatigue. Other examples include fatigue crack initiation at surface pits from corrosion, cyclic loading in a corrosive environment (stress-corrosion fatigue), and elastic buckling. Elastic buckling may cause parts to contact, causing seizure of a rotating system, but it may also lead to plastic buckling and ultimately to fracture.

The purpose of this article is to introduce the subject of fractography and aspects of how it is used in failure analysis. Fractography is the science of revealing loading conditions and environment that caused the fracture by a three-dimensional interpretation of the appearance of a broken component. If the specimen is reasonably well preserved, and if the analyst is knowledgeable, the fracture appearance reveals details of the loading events that culminated in fracture. An understanding of how cracks initiate and propagate microscopically to cause bulk (macroscale) fracture is an essential part of fractography. The ability to accomplish this resides in interpretation of fracture surface features at both the micro- and macroscales. It is important that examination of the fracture surface and adjacent component surface be done starting at low magnification with sequential examination of features of interest at increasing magnification. It is only in this way that significant features are identified as to location on the macroscale fracture surface. In other words, potential explanations for cause for failure must be consistent with both macroscopic and microscopic features. By understanding the fracture processes of the material involved, particularly at the microscale, interpretations of larger fracture patterns can be completed even when microscopic evidence is obliterated. One example is fatigue crack propagation analysis where striations (microscopic signatures of crack propagation) may have been damaged due to postfracture exposure. In such a case, macroscopic evidence indicating the nature and type of fatigue loading (beach marks, ratchet marks, etc.) may provide useful evidence to the analyst.

The ultimate purpose of fractography and the other methods of failure analysis is the determination of the (technical) root cause of failure, which may arise from various conditions, such as inappropriate use, an unanticipated operating environment, improper prior fabrication, improper or inadequate design, inadequate maintenance or repair, or combinations thereof. Possible root causes also include design mistakes such as inadequate stress analysis, alloy selection, improper mechanical/thermal processing, improper assembly, and failure to accommodate an adverse operating environment. Fractography provides a unique tool to determine potential causal factors, such as:

- Whether a material was used above its design stress
- Whether the failed component had or did not have the properties relied upon by the design engineer
- Whether a flaw or discontinuity was critical enough to cause failure at a time or loading level unanticipated by design
- Whether inspection methods were properly employed
- Whether features were present that indicate the dynamics of a fracture event (quasi-static versus dynamic/impact loading)

In this article, the basic types of fracture processes (ductile, brittle, fatigue, and creep) are described briefly, principally in terms of fracture appearances (as sometimes affected by the microstructure). More in-depth coverage on specific types of fracture (processes such as ductile and brittle fracture, fatigue, creep, and complex environmentally assisted cracking from stress-corrosion cracking and hydrogen embrittlement) can be found in other articles in this Volume. Articles on the fractographic appearances of polymeric and ceramic materials are also included. (Fractography of electronic components is addressed in detail in Ref 1.)

In summary, the following are key features in distinguishing between monotonic versus fatigue fracture and ductile versus brittle fractures (on either a macroscale or microscale):

- *Monotonic versus fatigue fracture:* Beach marks and striations indicate fatigue, but their absence does not confirm fracture from monotonic loads. Fracture surfaces from fatigue do not always reveal beach marks and fatigue striations.
- *Macroscale ductile versus brittle fracture:* Macroscale ductile fracture is revealed by obvious changes in cross section of the

fracture part and/or by shear lips on the fracture surface. Macroscale brittle fractures have fracture surfaces that are perpendicular to the applied load without evidence of prior deformation. Macroscale fracture surfaces can have a mixed-mode appearance (brittle-ductile or ductile-brittle). The brittle-ductile sequence is more common on the macroscale, while the appearance of the ductile portion is typically microscale in a ductile-brittle sequence.

- *Microscale ductile versus brittle fracture:* Microscale ductile fracture is uniquely characterized by dimpled fracture surfaces due to microvoid coalescence. Microscale brittle fractures are characterized by either cleavage (transgranular brittle fracture) or intergranular embrittlement.

Fracture Surface Information

Correct interpretation of fractographic features is a critical part of a failure analysis to determine the root technical cause for failure. Unfortunately, it is possible to misinterpret a fracture surface, and seldom can a unique cause for failure be associated with a single specific fractographic feature. The process of failure analysis, then, should use the total information available to reach a root-cause conclusion. Macroscale examination by itself is often inconclusive in identifying a unique set of conditions causing failure, and likewise, microscale examination without supporting macroscale examination and/or microstructural examination can lead to incorrect conclusions.

Both the macro- and microscale appearances of fracture-surface features reveal information regarding how and sometimes why fracture occurred. Fracture features can provide information about:

- Crack initiation site and crack propagation direction
- Mechanism of cracking and the path of fracture
- Load conditions (monotonic or cyclic)
- Environment
- Geometric constraints that influenced crack initiation and/or crack propagation
- Fabrication imperfections that influenced crack initiation and/or crack propagation

In the latter case, it is very important to make the distinction between a manufacturing imperfection and a manufacturing flaw (or defect). Manufactured components may contain geometric and material imperfections or nonconformities, but whether an imperfection caused a failure is critical in the determination of root cause. Manufacturing imperfections are not necessarily defects, and in many (if not all) situations, quantitative analysis should be considered to determine whether an imperfection is actually a root-cause flaw responsible for failure. Fabrication imperfections are discussed in more detail in the article "Mechanisms and Appearances of Ductile and Brittle Fracture in Metals" in this Volume.

Macroscopic Features

Macroscopic features typically help identify the fracture initiation site and crack propagation direction. The orientation of the fracture surface, the location of crack initiation site(s), and the crack propagation direction should correlate with the internal state of stress created by the external loads and component geometry. When the failed component is in multiple pieces and chevrons are visible on the fracture surface, analysis of crack branching (crack bifurcation) (Fig. 1) (Ref 2) can be used to locate the crack initiation site. Fracture initiates in the region where local stress (as determined by the external loading conditions, part geometry, and/or macroscopic and microscopic regions of stress concentration) exceeds the local *strength* of the material. Thus, variations in material strength and microscale discontinuities (such as an inclusion or forging seam) must be considered in conjunction with variations in localized stress that is determined by applied loads and macroscopic stress concentrations (such as a notch, hole, or other change in cross section).

The fracture surface orientation relative to the component geometry may also exclude some loading conditions (axial, bending, torsion, monotonic versus cyclic) as causative factors. For example, crack initiation is not expected along the neutral axis of a component loaded in bending or torsion, even if a significant material imperfection is present at that location because no normal stress acts along the neutral axis. (There is a shear stress at this location in bending, but in a homogeneous material, it is usually too small to initiate fracture. That may not be the case for a laminated structure loaded in bending.) Alternatively, brittle torsion failure is readily identified at the macroscale in cylindrical sections because of the unique helical nature of the fracture surface (Fig. 2) (Ref 3).

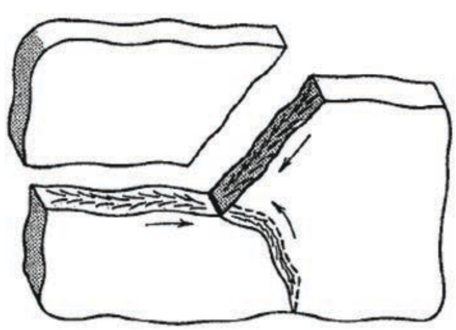

Fig. 1 Schematic view of a component that has fractured in multiple pieces. If chevrons are visible on the fracture surface, the sequence of crack formation can be used to obtain the crack formation sequence and the location of the initiation site. Source: Ref 2

In off-axis or bending fractures, the fracture plane is often generally perpendicular with the direction of maximum principal stress, providing information about the type and direction of loading. As the fracture progresses, the surface can curve as the principal stress plane orientation changes. As the fracture continues to the compression side of the component, the principal stress orientation changes, often resulting in a lip or compression curl in the final fracture region (Fig. 3).

Surface roughness and optical reflectivity also provide qualitative clues to events associated with crack propagation. For example, a dull/matte surface indicates microscale ductile fracture, while a shiny, highly reflective surface indicates brittle cracking by cleavage or intergranular fracture. In addition, when intergranular fracture occurs in coarse-grained materials, individual equiaxed grains have a

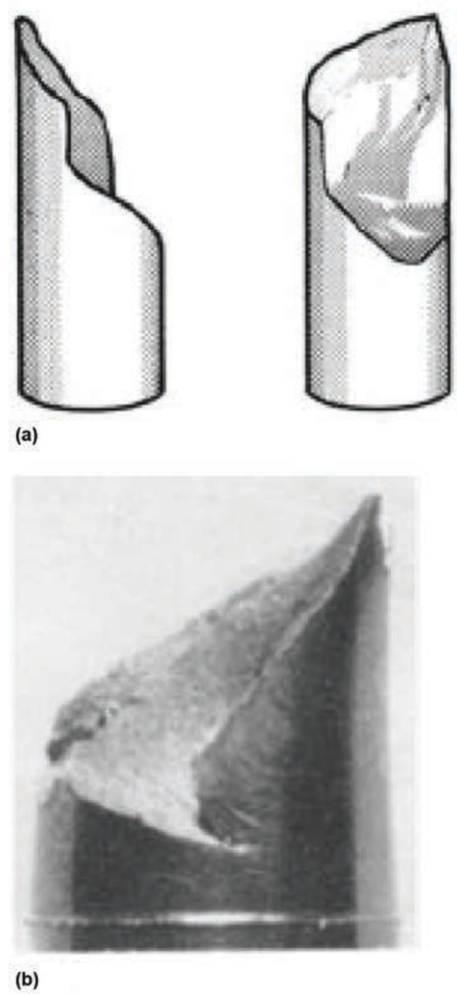

Fig. 2 Wolf's ear helical fracture due to torsion loading. (a) Schematic of brittle torsion fracture of chalk. (b) Helical tensile fracture of oxygen-free high-conductivity copper bar prestrained in torsion to a shear strain of 4.3×. Source (b): Ref 3

distinctive rock-candy appearance that may be visible with a hand lens. Surface roughness also provides clues as to whether the material is high strength (smoother) or low strength (rougher) and whether fracture occurred as a result of cyclic loading. The surfaces from fatigue crack growth are typically smoother than monotonic overload fracture areas. The monotonic overload fracture of a high-strength quenched and tempered steel is significantly smoother overall than is the overload fracture of a pearlitic steel or annealed copper. Also, fracture surface roughness increases as a crack propagates, so the roughest area on the fracture surface is usually the last to fail (Fig. 4) (Ref 4). Fracture surface roughness and the likelihood of crack bifurcation also increase with magnitude of the applied load and depend on the toughness of the material (Fig. 5) (Ref 5).

Radial marks (Fig. 6) (Ref 6) and chevrons (Fig. 7), if present, are macroscopic surface features that indicate the region of crack initiation and propagation direction. They are common and dominant macroscopic features of the fracture of wrought metallic materials but are often absent or poorly defined in castings. The "v" of a chevron points back to the initiation site, and a sequence of "v's" across the fracture surface indicates the crack propagation direction. Radial marks typically fan out from the initiation site, although Fig. 8 shows an exception to this.

Figure 8 (Ref 7) illustrates the importance of using fracture surface roughness in conjunction with radial marks to identify the initiating location for fracture. There is a light-colored region around the perimeter of the specimen, and there is a small dark region slightly in from the surface at the 9 o'clock position. Light-colored surface regions are often associated with surface hardening. If no information is available except the photograph, it may be (incorrectly) concluded from the radial markings that the dark region was the location of some material imperfection that initiated fracture. Typically, it is common for fracture initiation to occur within a relatively small region, where the center of the radial-fan markings provides a strong indication of the crack initiation region.

The center of the radial-fan markings is usually at or near the fracture initiation site, but this interpretation would be incorrect in the case of Fig. 8. The specimen shown in Fig. 8 was in fact prefatigue cracked and then loaded in monotonic tension to failure. Cracking initiated within a wide arc at the surface (at positions from approximately 11 to 4 o'clock), and the radial lines grew together rather than fanned out as the crack

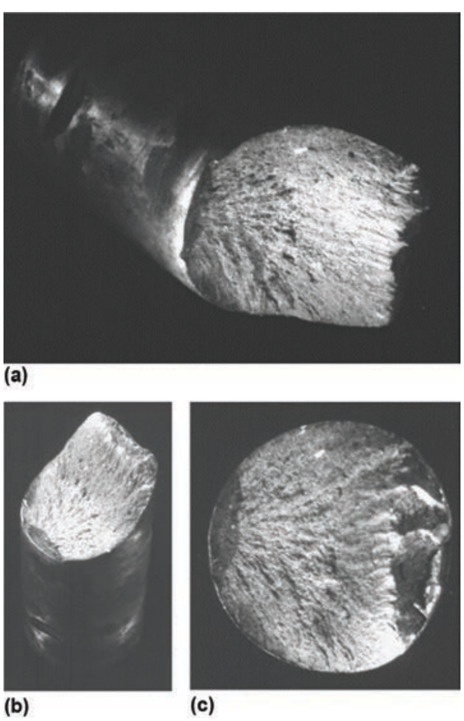

Fig. 3 Compression curl of bend fracture surface. (a, b) Angle views. (c) Top view showing radial marks emanating from the origin

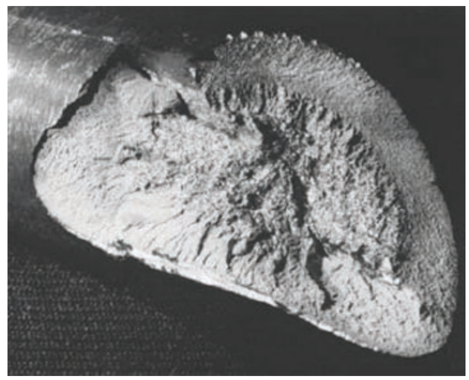

Fig. 4 Change in surface roughness due to crack propagation. Fracture surface roughness increases with distance of propagation, crack propagation rate, and decreased strength level. This component failed in fatigue. Crack initiation was on a longitudinal plane visible at the top in a surface-hardened region. The crack then propagated on a helical plane in torsion. Note the change in surface roughness as the crack propagates from the surface-hardened region at the top into the core and finally into the hardened case at the bottom of the photograph. The roughest region on the fracture surface is the final overload region at the bottom. Source: Ref 4

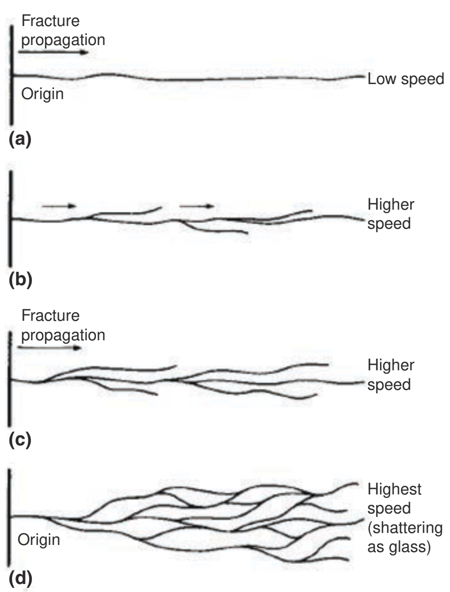

Fig. 5 Change in crack bifurcation with magnitude of the load and speed of crack propagation. (a) Low speed (low load) and high toughness. (b) Higher speed (higher load) and high toughness. (c) High speed (high load) and low toughness. (d) Higher speed (higher load) and low toughness. Source: Ref 5

Fig. 6 Macroscale radial marks. The fan array of the radial marks points back to the crack initiation site and is created under conditions of rapid crack propagation. Source: Ref 6

Fig. 7 Chevrons on the fracture surface of an induction-hardened axle fabricated from 1541 steel. The V-shaped chevrons point back to an initiation site marked by the arrow at the top of the figure. Component shows fatigue crack growth initiating at the arrow, creating the circular-shaped region at the top. Overload then occurred, and fast fracture created chevrons in the hardened case but not in the tougher core. Note also the increase in surface roughness in the pearlitic core as the overload crack grows. Source: Ref 6

propagated. The dark region at 9 o'clock is not an imperfection but rather contained within a rougher region with surface steps, indicative of the last region of fracture. Surface roughness increases as the crack propagates, and the rough area surrounds the dark spot. The spot appears dark because of the angle of the illumination. The deciding issue is the surface roughness, because it indicates that the later material to fail was near the dark region and not at the surface. Typically, it is common for fracture initiation to occur within a relatively small region (in which case the center of the radial markings would fan out from the small region of crack initiation). However, as demonstrated in Fig. 8, it is possible for crack initiation to occur over a wider region.

Fracture patterns are also indicative of the state of stress operative at the time of fracture. Although beyond the scope of this article, analytical tools such as finite-element analysis are often useful in correlating fracture patterns to the stress state. Similarly, computed tomography (CT) and other metrology techniques such as laser scanning can be quite useful in fracture pattern assessments. A typical example of the application of such tools is described in Appendix 1 of this article, along with a cursory discussion of fracture mechanics principles, and understanding of which is often required to properly interrogate fracture surfaces and patterns.

Microscopic Examination and Appearances

Microscopic examination can help identity imperfections that initiate cracking, the path of fracture (intergranular or transgranular), and the microscale mechanisms of cracking (i.e., microvoid coalescence, cleavage, or fatigue). Localized directions of crack growth also can be determined from the river lines of cleavage fracture surfaces. In the case of microvoid coalescence, the shape of the dimples on the fracture surface can be correlated with loading conditions (i.e., equiaxed dimples from tensile loading; elliptical dimples from shear or torsional loading).

Microscale fractographic features help identify the microscopic mechanism(s) causing fracture. Figure 9 (Ref 8) is a summary of the possible microstructural features associated with the basic types of external load conditions (overload, fatigue, and environmentally assisted sustained-load cracking). As indicated in the figure, a dimpled fracture surface is uniquely associated with the microscale mechanism of microvoid coalescence, which typically is associated with macroscopic ductile fractures. However, macroscale brittle fractures can also occur when plastic deformation is limited to a small volume of material and not macroscale visible, while the fracture process is still microvoid coalescence. This is the case when the ductile fracture mechanism of microvoid coalescence is constrained to a plane-strain fracture mode (referred to as plane-strain microvoid coalescence) or occurs preferentially in the limited region adjacent to the grain boundary (resulting in a dimpled intergranular fracture surface). These types of ductile and brittle fractures are discussed in more detail later in this article.

In a more general sense, the microscale features of fractures in crystalline materials can be described as either transgranular (TG) or intergranular (IG). Transgranular crack propagation is caused by competing mechanisms of ductile crack nucleation, growth by slip deformation mechanism, and brittle cracking by cleavage. (As described later in this article, twinning is a TG mechanism of plastic deformation. Deformation twinning provides a limited amount of ductile deformation but also provides an alternative for initiation of cleavage cracks. What can result from twinning deformation is cleavage crack nucleation at the intersection of mechanical twins, for example, as discussed further in the article "Mechanisms and Appearances of Ductile and Brittle Fracture in Metals" in this Volume.)

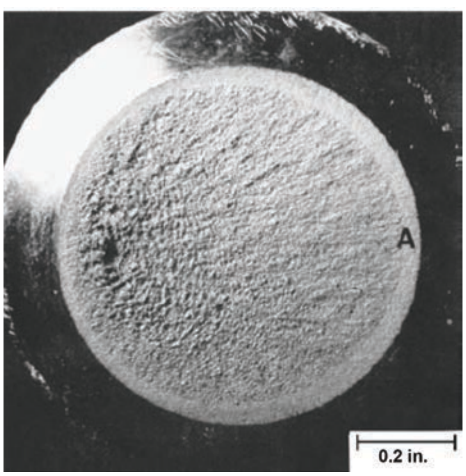

Fig. 8 Macroscale brittle fracture in tensile loading. A light ring is visible around the outside circumference. A faint radial pattern is visible from approximately 11 to 4 o'clock and running toward a dark spot near 9 o'clock. The roughest area on the fracture surface is near the dark spot (see text for discussion). Source: Ref 7

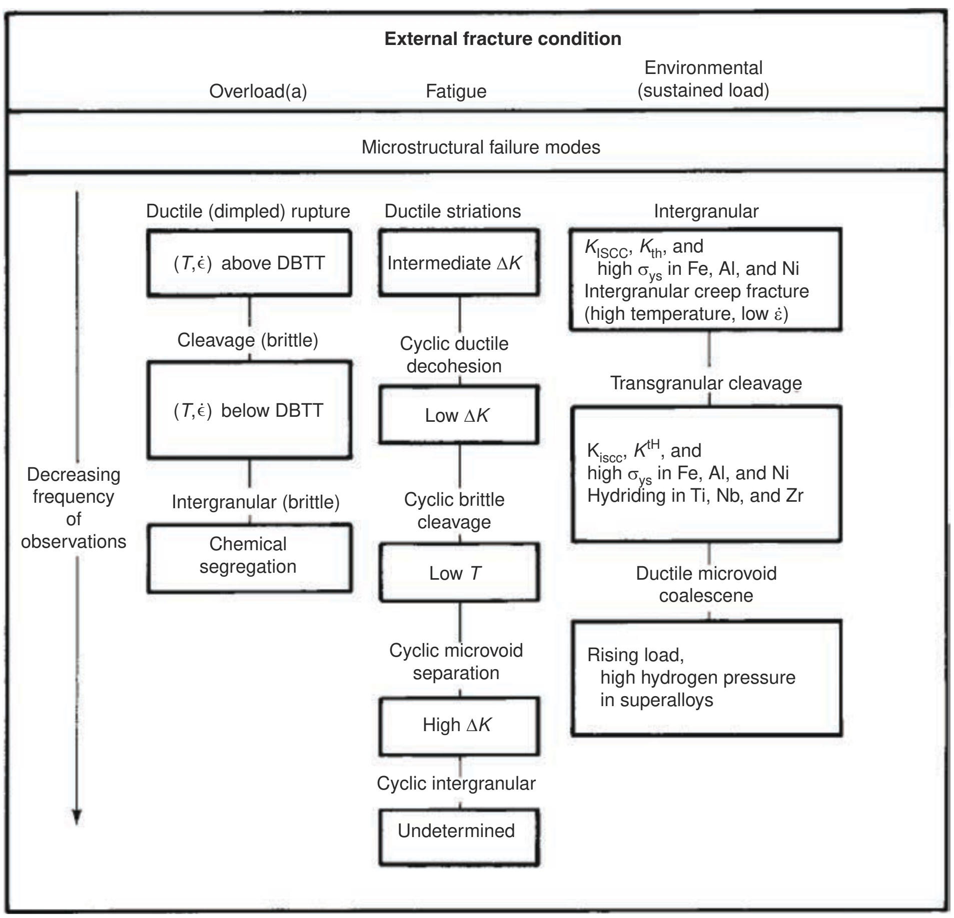

Fig. 9 Observed microscopic fracture mechanisms for different loading conditions and environments. T, temperature; $\dot{\epsilon}$, strain rate; DBTT, ductile-brittle transition temperature; ΔK, stress-intensity factor range; K_{ISCC}, stress-corrosion cracking threshold; K_{th}, threshold stress-intensity factor; σ_{ys}, yield strength stress. Note (a): See Fig. 14 and discussions for conditions of macroscale ductile and brittle fracture. Adapted from Ref 8

These two mechanisms of TG cracking have distinct appearances on the microscale, characterized by a dimpled fracture surface for ductile TG fractures (Fig. 10) and the distinctive river lines of cleavage for brittle cracking (Fig. 11). The occurrence and appearances of these TG cracking mechanisms are influenced by crystal structure, microstructure, loading rate, and temperature, as briefly discussed later in this article and in more detail in the article "Mechanisms and Appearances of Ductile and Brittle Fracture in Metals" in this Volume.

Intergranular fracture is clearly distinguished from TG fracture, but unlike TG fracture, the microscopic appearances of IG fractures are not uniquely associated with a specific microscale mechanism. (A fracture surface can contain both IG and TG fracture.) Preferential cracking in (or near) the grain boundaries may be related to various types of IG mechanisms (which may be diffusion-related processes such as creep void formation, hydrogen void formation, or impurity segregation in the grain boundaries). Sometimes the facets of IG fracture are almost featureless, perhaps containing only the presence of a second phase when they occur truly in the grain boundary. However, very small dimples can be seen on grain-boundary facets. Dimpled IG fracture can occur from void formation in the grain boundaries or from the mechanism of microvoid coalescence in the region adjacent to the grain boundary.

The possible influence of the microstructure or the environment sometimes can be determined more easily by microstructural examination, especially with specimens taken perpendicular to the fracture surface and containing the fracture surface in edge view. Specific examples include identification of IG fracture and the possible role of inclusions and/or second phases in influencing the direction of crack propagation. In summary, microscopic examination of the fracture surface can but may not always or uniquely provide information regarding:

- Whether crack propagation (i.e., the fracture progression mechanism) is ductile or brittle
- Whether loading was axial, bending, or torsion and monotonic or cyclic
- Whether fracture may have occurred at a high fraction of the melting point (high homologous temperature, T_H)
- Whether the environment played a role in the fracture
- Whether the thermal processing history of the material was improper

The presence of oxidation products may indicate elevated-temperature service, and the presence of surface discoloration may indicate a corrosive service environment. Thicker surface deposits from liquid on the surface often crack in a distinctive manner as they dry and result in mud cracks (Fig. 12, 13). These mud cracks in the surface deposit may indicate the possibility of an environmentally induced fracture, that is, stress-corrosion cracking. Unfortunately, mud cracks can also be created by caustic cleaning residue from attempts to clean the fracture before providing the specimen to the analyst. The potential of altering the fracture surface during the investigative process reinforces the need for thorough documentation of the fracture during all stages of the failure analysis process.

Ductile and Brittle Behavior

Perhaps most importantly, the question of whether a fracture is ductile or brittle is almost always addressed in a failure analysis. *Ductile* and *brittle* are terms often used to describe the amount of macroscale plastic deformation that precedes fracture. The presence of brittle fracture is a concern, because catastrophic brittle fracture occurs due to the elastic stress that is present and usually propagates at high speed, sometimes with little associated absorbed energy. Fracture occurring in a brittle manner cannot be anticipated by the onset of prior macroscale visible permanent distortion to cause shut down of operating equipment, nor can it be arrested by removal of the load, except for very special circumstances.

Note that the terms *ductile* and *brittle* also can be and are applied to fracture on a microscopic level. At the macroscale, ductile fracture by the microscale ductile process of microvoid formation and coalescence is characterized by plastic deformation and expenditure of considerable energy, while microscale brittle fractures by cleavage are characterized by rapid crack propagation with less expenditure of energy than with ductile fractures and without macroscale evidence of plastic deformation. The point is that the terms *ductile* and *brittle* are used to describe both appearance (macroscale behavior) and mechanism (microscale behavior). The macroscale view of ductility is neither more nor less correct than the microscale definition for the fracture mechanism.

The specific meaning of *ductile* and *brittle* may carry different connotations depending on background, context, and perspective of the reader. It is therefore important to clearly identify whether a ductile or brittle fracture is being described in terms of macroscale appearance or microscale mechanisms. Also note there is no universally accepted dividing

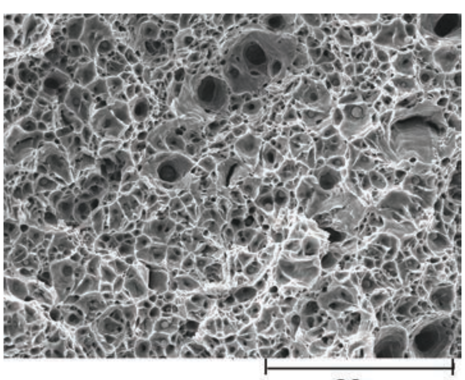

Fig. 10 Dimpled rupture created by microvoid coalescence. Courtesy of Engineering Systems, Inc.

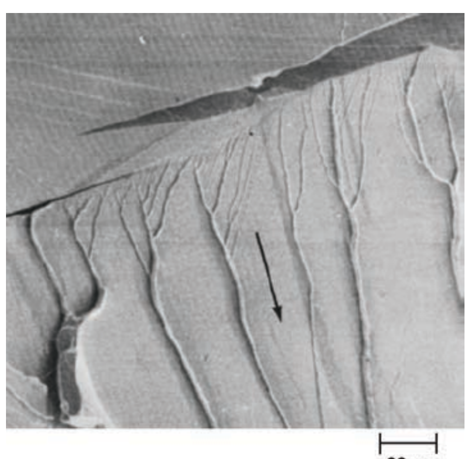

Fig. 11 River lines on a cleavage fracture surface. Direction of growth is parallel to the direction of crack coalescence, as indicated by the arrow. Cracks must reinitiate at a boundary containing a twist (mode III) deformation component

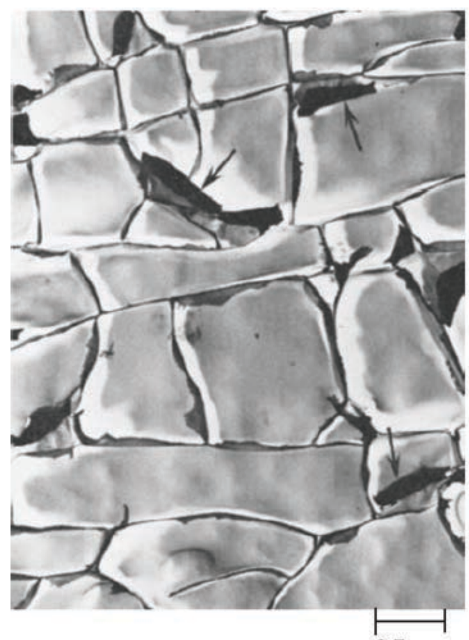

Fig. 12 Mud cracks on the surface of an intergranular fracture in 7079-T651 aluminum that failed under stress-corrosion cracking conditions in a 3.5% chloride solution. Transmission electron microscopy replica

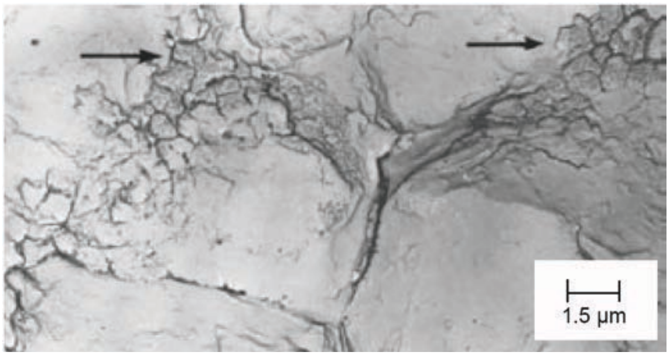

Fig. 13 Mud cracks on the fracture surface of a quenched and tempered 4340 steel exposed to a marine environment. Transmission electron microscopy replica

line for macroscale ductile and brittle behavior in terms of strain at fracture nor in terms of energy absorption. For example, large fracture strain is desirable for forming operations, and materials selection may be based on the relative ductility observed during tensile testing.

Another set of criteria may apply in structural design, where analytical expressions to determine allowable loads are based on whether failure is ductile or brittle. Some (arbitrary) value of tensile elongation or reduction in area is required to define whether a (ductile) distortion energy yield criterion or a (brittle) maximum normal stress or maximum shear criterion (perhaps modified by a normal stress term, such as the Coulomb-Mohr model) is used in design. Ductile behavior also is often associated with high energy absorption at fracture, and adequate toughness or ductility may be evaluated and defined by impact data, where criteria to determine whether the fracture is ductile or brittle involve some minimum level of absorbed energy at the service temperature of interest.

The macroscale definition of ductile versus brittle behavior also may be misleading about material behavior. For example, when subjected to large compressive hydrostatic loads, brittle materials may behave in a ductile manner. The fracture strain of ductile materials increases with an increase in loading conditions containing a large compressive hydrostatic component relative to the deviatoric component of stress and decreases with an increase in the tensile-hydrostatic stress component. It is also possible for ductile fracture to require little energy for initiation or propagation if strain-hardening capacity is low.

From the perspective of safe design, materials that are inherently ductile but can behave in a brittle manner in service require the most caution. Many engineering materials are inherently ductile and some are inherently brittle, but those behaviors can be altered. Possible reasons for brittle behavior of ductile materials include loading conditions and the internal state of stress created by the part geometry and the geometry of any imperfections in conjunction with the operating environment (chemically reactive and/or high or low temperature). The inherent ductile behavior of metallic material also can be drastically reduced by improper heat treatment (e.g., incipient melting, temper embrittlement, improper age hardening) or by processing (hydrogen embrittlement due to plating baths). Therefore, it is necessary to understand:

- Why some materials are inherently ductile or brittle
- How part geometry, operating conditions, or mechanical/thermal processing may alter that behavior

The inherent ductility or brittleness of materials is addressed later in this article in the section "Structure and behavior."

Observation of Plastic Strain

Smaller amounts of plastic deformation may be determined via careful measurement if the surfaces of the component are relatively smooth. The ability to see a neck in a tensile specimen depends on the amount of strain hardening and, to some extent, the amount of strain-rate hardening. If there is no hardening to force the neck to grow along the length of the specimen, plastic flow via slip can occur without visual evidence. That is, there may be microstructural evidence or microscale fractographic evidence of plastic deformation, but it occurs over a sufficiently small volume that it is not visually apparent.

In some instances, small amounts of plastic deformation may be visible at the macroscale, such as the twisting of extrusion marks around the axis of the component (torsion loading). Two halves of a bending fracture can often be brought into close proximity to determine if a small amount of plastic bending has occurred (for example, by placing the two components on a flat surface). This is a helpful technique in the examination of threaded cylindrical sections. However, it is of *extreme* importance that two fracture surfaces not be brought into actual physical contact. *Doing so can destroy microscale fractographic information.*

Preferred methods for quantification of plastic strain in a fractured component are the use of three-dimensional laser scanning or CT scanning. High-resolution scanning of a fracture surface and surrounding regions allows for the realignment of fracture surfaces in a component and the determination of plastic strain, both locally and globally, as long as original or reference geometry is available. A case study illustrating this method is presented in Appendix 1 of this article.

Sometimes plastic strain can also be seen by examination of the surface of the component adjacent to the fracture. Plastic strain will result in a roughening of the surface if the grain size is very large. Conversely, the presence of a large grain size may be visible (detected) by roughening of the surface for a component with a distortion of the original geometry. In extreme cases, the roughening occurs over the complete section, not just the area immediately adjacent to the fracture surface, and is then described as "orange peel."

It is also possible to use scanning electron microscopy or atomic force microscopy to obtain three-dimensional surface maps showing surface profiles. Hull (Ref 9) describes a technique using surface profiles that can be used to identify small-scale plastic strain. Profilometer traces are obtained from matching regions on each half of the fracture surface. If the two traces cannot then be brought into alignment, it is likely that there has been some plastic deformation associated with fracture. If a piece has dropped out of the surface, there may be no matching, but neither has there been any plastic deformation.

It is also important to clarify whether the term *ductile* refers to:

- Plastic strain accumulated prior to the nucleation and growth of a crack
- The process of crack nucleation
- The process of crack growth

As an example of the necessity for careful description, consider two tensile specimens fabricated from the same alloy, but one is in the annealed condition and the other prepared from cold rolled material. The annealed material is expected to show fracture in the necked region, but no or minimal necking is expected in the cold-worked material. The microscopic mechanism of fracture in the annealed material is a ductile mechanism (microvoid coalescence), and macroscopic deformation preceded fracture. Thus, there is little confusion in describing the fracture as ductile on both the macroscopic and microscopic scales of observation.

However, what about the cold-worked material? There was plastic deformation (presumably compressive) during manufacturing, and the presence of prior cold work may explain the absence of the macroscopic necking.

Metallographic observation and/or hardness testing would determine the material condition and clarify the effect of previous cold working on the fracture appearance, which could be either ductile or brittle at the microscale.

Macroscopic Ductile and Brittle Fracture Surfaces

As previously noted, there is no universally accepted dividing line between ductile and brittle behavior at the macroscale in terms of strain at fracture, nor is there a defined dividing line in terms of energy absorption. The macroscale fracture appearance that occurs depends on the microstructure (strength and ductility) of the material and the degree of constraint associated with the presence of a cracklike imperfection.

Under plane-stress conditions, a fracture is typically considered brittle at the macroscale if it is oriented orthogonally to the maximum normal stress (condition 4 in Fig. 14). A fracture is typically considered to be macroscopically ductile when the fracture surfaces are oriented at an angle of approximately 45° to the maximum normal stress. A fracture surface displaying both types of planes can be described as a mixed-mode fracture.

The local state of stress created by a load on a component geometry may cause crack propagation (i.e., critical fracture) that results in a fracture surface with a macroscale appearance; that is:

1. Totally ductile
2. Totally brittle
3. Initially brittle, then ductile
4. Initially ductile, then brittle
5. Mixed mode (ductile and brittle)

In the latter two cases (4 and 5), the ductile appearance may not be directly visible at the macroscale. Fractures that are initially ductile then transition to brittle (case 4) are usually associated with rising-load ductile tearing, or the initial ductility may be inferred by transverse strain at the crack tip. The size of the plastic zone may be microscale in this case. Mixed-mode ductile and brittle cracking (case 5) would be inferred due to the presence of an intimate mixture of cleavage and microvoid coalescence at the microscale or by the presence of shear lips at the macroscale.

Note that some of the aforementioned criteria are based on macroscopic conditions or appearances and do not consider the microscopic mechanisms (i.e., slip, twinning, viscous flow, cleavage) that cause fracture. A fracture may appear to be macroscopically brittle, but the cracking process may occur by a ductile mechanism. Examples in which the cracking mechanism is ductile but for which there is no or little visual macroscopic distortion include monotonic loading of a component containing a cracklike imperfection (plane-strain microvoid coalescence fracture induced by part and crack geometry), long-life cyclic loading, and elevated-temperature failure (IG creep fracture). These examples are discussed in subsequent sections of this article, but the point is that the terms *ductile* and *brittle* should be used carefully with respect to the scale of observation or the description of fracture mechanisms. The distinction is important, because macroscopic brittle fractures can occur from the microscopic mechanism of ductile cracking.

Constraint and Macroscopic Fracture Appearance

Constraint is created by longer cracks, thicker sections, and a decreased crack tip radius. If the material is inherently brittle (a steel below the ductile-brittle transition temperature), crack initiation is expected at or near the preexisting cracklike imperfection, and the crack is expected to propagate in a microscale brittle manner. When the material has some inherent ductility, the fracture process is influenced by component and crack geometry creating various fracture surface features. The purpose here is not to discuss microscopic details of fracture initiation and crack propagation but rather to characterize the macroscopic appearance. The features to be considered are:

- Crack blunting and crack propagation on a plane of maximum shear stress
- Loss in constraint due to crack propagation with a macroscale transition from plane-strain flat fracture (normal to the load) to plane-stress slant fracture
- Mixed-mode fracture and incomplete constraint resulting in shear lips and crack-arrest lines
- Creation of constraint by subcritical crack growth resulting in a fracture surface predominantly flat after a small initial ductile region (which may not be visible on the macroscale)

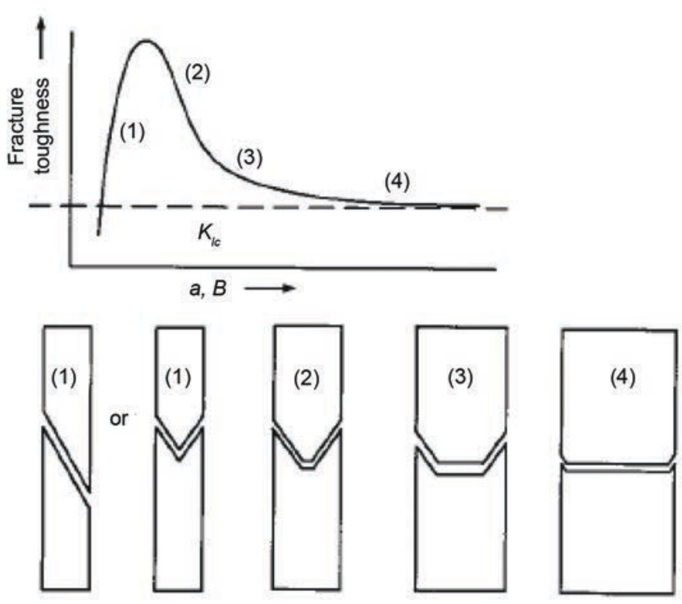

Fig. 14 Schematic of variation in fracture toughness and macroscale features of fracture surfaces for an inherently ductile material. As section thickness (*B*) or preexisting crack length (*a*) increases, plane-strain conditions develop first along the centerline and result in a flat fracture surface. With further increases in section thickness or crack size, the flat region spreads to the outside of the specimen, decreasing the widths of the shear lips. When the minimum value of plane-strain toughness (K_{Ic}) is reached, the shear lips have very small width.

Plane-Strain Microvoid Coalescence

As previously noted, ductile cracking by microvoid coalescence can result in a macroscale brittle fracture when the cracking is constrained by the geometry of the part and/or crack. With geometric constraint, plastic strain may be concentrated and lead to fracture without visible macroscale deformation. The microscale cracking mechanism is ductile, but geometric constraint limits macroscale distortion. This type of fracture may best be referred to as plane-strain microvoid coalescence, following the previous definition of macroscale brittle fracture and also characterizing the microscopic process of cracking. The geometry of the part and/or crack is thus one factor that may influence the macroscale deformation of the fracture process (distinct from the microscale mechanisms of cracking, which are discussed later in this article).

Shear Lips and Crack-Arrest Lines

Consider first the effects of section thickness for an intermediate value of crack length and a sharp crack tip. For thin sections there is little constraint imposed by a stress concentrator, so that the fracture process occurs essentially under conditions of plane stress, resulting in complete slant fracture (condition 1 in Fig. 14). As the section thickness increases, constraint, which is higher along the centerline than at the free surfaces, becomes sufficiently large to create plane-strain conditions and result in flat fracture (condition 4 in Fig. 14). The slant fracture surfaces (conditions 2 and 3 in Fig. 14) are described as shear lips, or, alternatively, the fracture can be described as mixed mode. Orientation of the shear lips may be used to identify the crack initiation location, which is helpful because chevrons or radial marks may not be present. The direction of crack propagation is parallel to the shear lips.

Further increases in section thickness spread constraint toward the sides of the specimen, decreasing the width of the shear lips and ultimately resulting in a fracture that is essentially 100% flat (condition 4 in Fig. 14). (There is still a vanishingly small shear lip unless the material is inherently brittle.) The crack length and or section thickness required to obtain essentially flat fracture (i.e., plane-strain fracture) can be estimated from:

$$a, B \geq 2.5\left(\frac{K_{Ic}}{S}\right)^2$$

where a is crack length, B is section thickness, K_{Ic} is plane-strain fracture toughness, and S is nominal stress.

When the fracture surface is essentially flat, a quantitative assessment of fracture toughness or stress level at the time of fracture can be obtained (see Appendix 2 of this article). When small shear lips are present on the flanks of the specimen, assumptions of plane-strain loading are less accurate, and the stress intensity at the time of fracture is greater than that indicated by K_{Ic}. However, Hertzberg (Ref 2) has proposed a procedure (Appendix 2) to estimate the toughness and/or stress level based on the width of the shear lip.

As constraint increases behind the notch, the through-thickness stress increases. This may lead to splitting of a plate near midthickness, especially if inclusions are concentrated in this region and have a high aspect ratio parallel to the width direction or if the material is heavily banded (say pearlite-ferrite banding in a steel, Fig. 15) (Ref 10). When there is less constraint at the notch tip from those situations described previously or for an inherently tougher material, crack blunting becomes significant and can lead to ductile tearing on a plane of high shear stress rather than on the plane of maximum normal stress.

Even when there is significant constraint at the notch tip, a small amount of plastic tearing can occur on the plane of maximum normal stress in conjunction with crack tip blunting (Fig. 16) (Ref 11). Then, depending on the degree of constraint, subsequent crack propagation can be microscale ductile or brittle. This limited microscale ductility provides a second quantitative evaluation of toughness and stress at fracture by relating the crack tip opening displacement to toughness. The analytical relations are described in Appendix 2 of this article. A discussion of the appearance is also in the article "Mechanisms and Appearances of Ductile and Brittle Fracture in Metals" in this Volume.

Depending on the level of constraint and fracture toughness when fracture initiates, the stored elastic strain energy may or may not be sufficient to drive the crack completely across the specimen. A common situation in laboratory testing is that the crack "pops in" the specimen; that is, a small crack suddenly forms under plane-strain conditions with a concurrent drop in load. The load then rises, and crack propagation continues by ductile tearing. The process can repeat more than once, leaving telltale crack-arrest marks on the fracture surface (Fig. 17) (Ref 11).

Note that the crack-arrest marks indicate crack tunneling along the centerline and that they are matte in appearance compared to the generally shiny reflective surface. The curvature of the arrest lines delineates the crack front and indicates the direction of crack propagation. *These crack arrest lines should not be confused with beach marks in cyclic loading nor with chevrons in monotonic loading*. Chevrons created on a flat fracture surface point back to the crack initiation site; arrest lines point in the direction of crack propagation. Microscale examination shows that the arrest lines are created by a change in fracture mechanism, not due, for example, to crevice corrosion as for beach marks in a steel. Microscale examination of the fracture surface shows that the highly reflective regions of the fracture surface are created by cleavage or quasi-cleavage, while the thin-arced arrest regions failed by microvoid coalescence.

Structure and Behavior

Fractographic features are related to not only the geometry, loading conditions, and service environment but also to the inherent properties of the material as controlled by its submicro-, micro-, and macroscale structure. The combination of alloy composition, microstructure, macrostructure (segregation, banding, and fibering), service loading conditions (monotonic, cyclic, uniaxial, multiaxial, and so forth), service environment (chemically aggressive, low temperature, and high temperature), and the possibility of both geometric and material imperfections created by the processing history provide a large number of conditions that influence the tendency to fracture. For purposes of discussion, it is convenient to distinguish between crystalline, noncrystalline, and

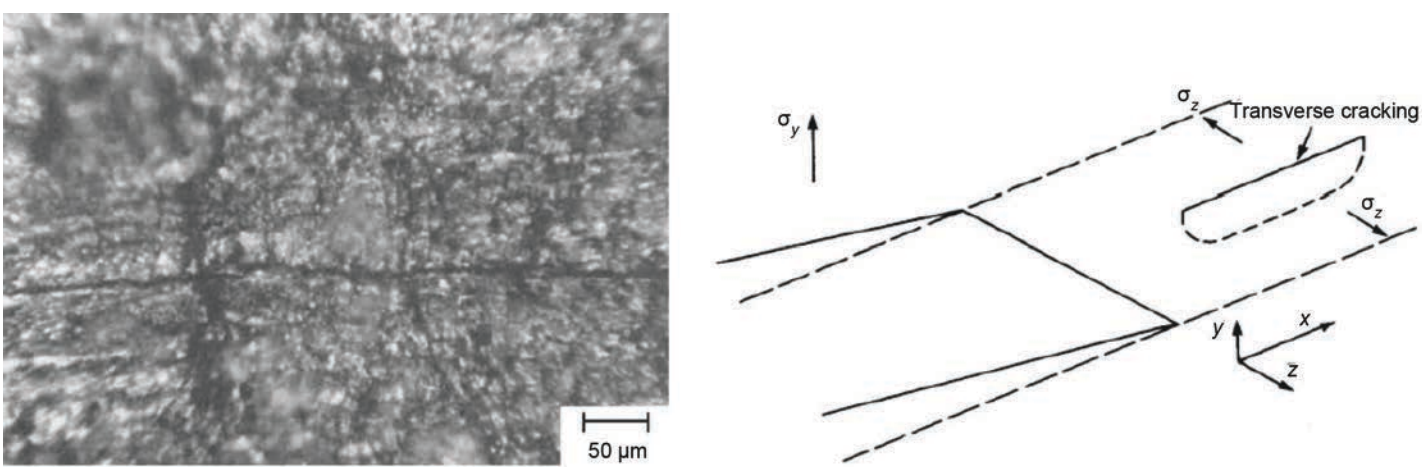

Fig. 15 Centerline cracking in a plate containing a crack-like defect. Constraint in the thickness direction created by the crack-like defect causes a transverse stress (σ_z). This stress sometimes causes transverse cracking. Source: Ref 10

partially crystalline materials and, in some cases, to consider the behavior of mechanical mixtures (composites, aggregates, and mechanical alloys).

Under monotonic loading, the competing processes of ductile fracture by deformation mechanisms (slip and possibly twinning) and brittle fracture by cleavage are influenced by crystal structure, microstructure, loading rate, and temperature. These mechanisms are discussed followed by a section on the appearances of fatigue fractures and then the sources of crack initiation. Many fracture mechanisms also have associated names, which are descriptive (e.g., stress-corrosion cracking, corrosion-enhanced fatigue, temper embrittlement, liquid metal embrittlement, etc.) and, when used correctly, imply and sometimes explain causes for failure. However, to reiterate, results obtained from only microscale examination may not indicate a unique cause for fracture.

Fig. 16 Ductile tearing on a plane of maximum normal stress at the tip of a compact tension specimen. Material is O1 tool steel. Source: Ref 11

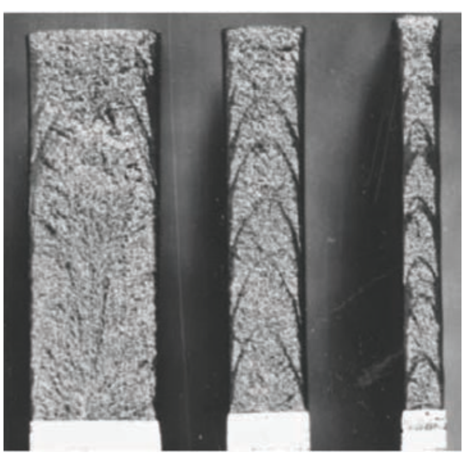

Fig. 17 Crack-arrest lines on edge-notched tension specimens. Material thickness: 13 mm (½ in.), 10 mm (⅜ in.), and 6 mm (¼ in.). Note the distance for first arrest, which increases with section thickness, and note that the arrest lines are not closed along the centerline in the 13 mm (½ in.) thick specimen, indicating full constraint at that location. Source: Ref 11

Atomic-Level and Microscopic Structure

Two important submicroscopic variables are the type of bonding between atoms (ionic, covalent, metallic, and van der Waals) and whether the material in question is or is not crystalline. Metallic materials are metallically bonded and are typically crystalline. Ceramic materials are predominantly ionically bonded but may show some covalent bonding. They can show three-dimensional order (crystalline), two-dimensional order (laminar or layered structures), or no order (amorphous).

Polymeric materials are typically amorphous or partially crystalline. In organic polymeric material, carbon atom backbone chains and pendant atoms on the chain are covalently bonded. The individual polymeric chains are then either van der Waals bonded or may be cross linked, that is, covalently bonded. Covalent (and ionic) bonds are typically high-strength bonds, while the van der Waals bond is weak. Polymers may or may not be oriented (i.e., whether there is alignment of the carbon backbone chains). Seldom are polymeric materials completely crystalline, and crystallinity decreases with complexity of the pendant atom groups (steric hindrance), chain branching, as well as with increasing molecular weight.

The chains, unless specific procedures have been undertaken to cause alignment, are extensively kinked and interwoven (Fig. 18) (Ref 12). Behavior of polymeric materials also depends strongly on whether the service temperature is above or below the glass transition temperature, where the molecule dramatically stiffens. The elastic moduli and strength increase below the glass transition temperature, while the engineering strain to fracture decreases. Additionally, several mechanical properties of polymeric materials depend on the average molecular weight.

Qualitatively, strength and modulus are increased as crystallinity increases, while ductility is usually reduced. In contrast to modeling of metallic material behavior, it is uncommon to describe behavior of polymeric material in terms of dislocation models and/or microscale slip and twinning processes.

Deformation and Fracture

Plastic deformation in crystalline material at low homologous temperatures is a consequence of TG microscopic deformation by slip and/or twinning in the crystalline lattice and, at higher homologous temperatures (for example, $T_H \sim 0.4$), by slip and viscous

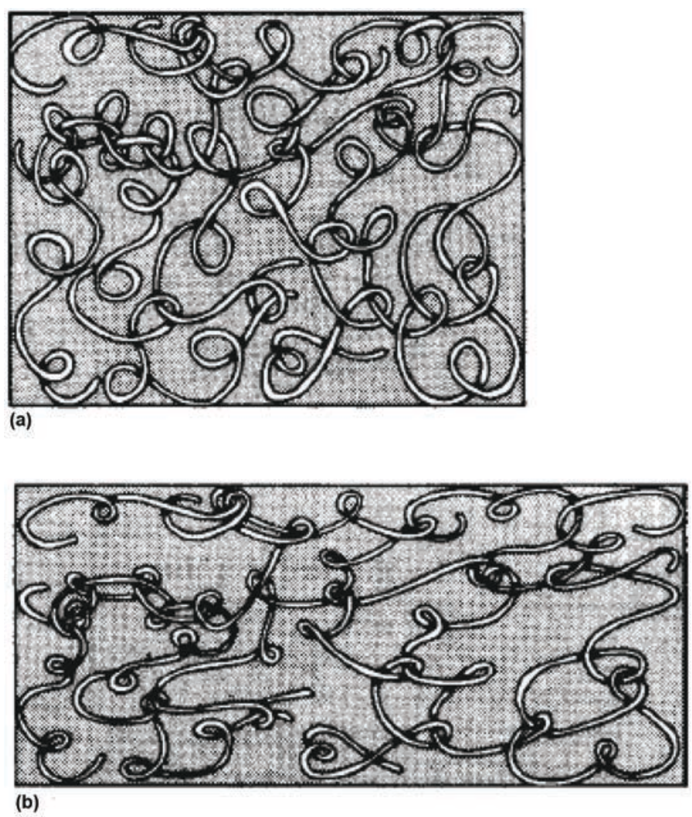

Fig. 18 Schematic picture of spaghetti-bowl appearance of an unoriented amorphous polymer. (a) Prior to plastic strain. (b) After plastic strain; twisting and kinking are reduced and the polymer chains become oriented in the direction of plastic strain. Source: Ref 12