1. Introduction and fracture overview

In this chapter, a fundamental approach to fracture will be followed in order to develop a few central ideas about why any given material is ductile or brittle, and how the basic lattice defects of the material cooperate to give rise to these two general classes of behavior. The analytical discussion will tend to focus on relatively brittle phenomena, partly because any idealization of fracture-leads one first to the consideration of *brittle fracture*, and partly because the analysis for brittle fracture is better understood, and therefore more accessible. Nevertheless, the more ductile aspects of fracture, though more difficult to analyze quantitatively, will be kept in view throughout, because of the inherently ductile tendencies of most metals.

There are two extreme poles of material failure under stress. In the first, fig. 1, the material *necks down* under tension like taffy (single crystals of pure copper are an example) in a continuous plastic manner, until the last atoms in the narrowing neck come apart and the material splits into two pieces. This type of failure is usually termed *ductile rupture*. In the other case (fig. 2), the material suddenly and catastrophically fractures with no discernible plastic flow either preceding or during the failure event. Examples of the latter are glass and silicon at low temperature. These two poles are useful to keep in mind, because they illustrate how two crystal or lattice defects — dislocations and atomically sharp cleavage cracks — are responsible, ultimately, for the mechanical failure of crystalline materials. In the first instance, only dislocations are involved, while in the second only a sharp crack participates in the failure event. Materials scientists are not accustomed to thinking of brittle cracks (fig. 3) as one of the fundamental crystal defects which have proven so useful as constructs for understanding the structure-sensitive material properties. However, in this chapter, we will put such cracks on the same conceptual footing as dislocations because, like dislocations, they answer one of the two most fundamental questions in the mechanical behavior of materials. Dislocations were postulated to explain why the yield strength of a solid can be much lower than the theoretical shear strength of the atomic bonds of the solid. Likewise, cracks provide the answer to why fracture occurs in materials at average stress levels well below the theoretical tensile strength of the atomic bonds of the material. The reason is that a pre-

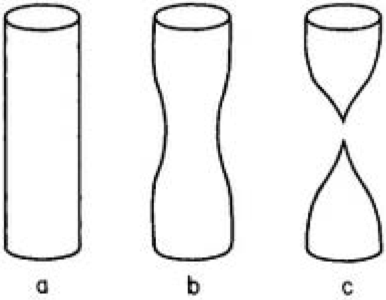


Fig. 1. A rod of ideally ductile material when pulled develops a region of plastic instability which finally thins uniformly down to a sharp point.

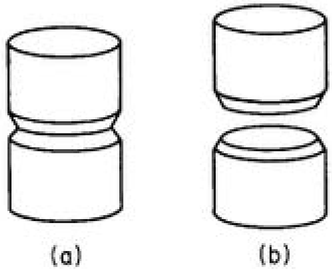


Fig. 2. An ideally brittle material when pulled separates suddenly by cleavage with no prior or simultaneous deformation. A notch is shown which localizes the plane of fracture.

existing crack in the material magnifies or concentrates the low applied stress to a value at the tip of the crack which may be many orders of magnitude higher. When this magnified stress at the tip reaches the maximum atomic bond strength, the crack will propagate through the solid and fracture it. From this chapter, we hope the reader will find that the otherwise very confusing subject of mechanical failure is best understood by viewing it as an interplay between, and sometimes a competition between dislocations and cracks.

The *stress concentration* referred to around a slit crack is shown in fig. 4 where the analogy of the magnetic lines of force around a similar slit in a magnet is striking. Physically, the crack must act as a stress concentrator, because every horizontal plane (line in two dimensions) in the figure must carry the total force imposed on the two external surfaces. Since the cut line representing the crack means there is less material on the cleavage plane defined by the cut to carry the applied force, the force per unit area (or stress) on this plane outside the crack must be higher. The stress is especially concentrated near the source of the "force transmission difficulty" at the crack tips, and

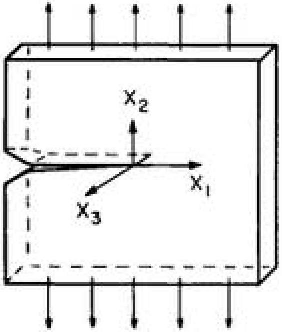


Fig. 3. A brittle crack in a material is viewed as a mathematical cut in the medium across which the atom bonds have been cut. For mathematical simplicity the crack line will be assumed to be straight and to lie along the x_3 axis. The cleavage plane is the negative x_3x_1 plane. External stresses are also exerted on the external surfaces of the material in an arbitrary manner in order to hold the crack open.

References: p. 2287.

Fig. 4. Stresses are concentrated near the tip of a crack in a way suggestive of magnetic lines of force around a slit. In this diagram, the lines are computer-generated to simulate crack stress lines quantitatively (Courtesy Prof. P. NEUMANN.)

indeed we will show that for a mathematically sharp cut, the stresses become singular at a tip. If the tip is rounded, the maximum stress is finite, and if the tip has a radius of atomic size, the maximum average stress which can be sustained in the material is that which begins to break the bonds at the tip in a progressive fashion as alluded to above (fig. 5).

An interesting contrast between dislocations and cracks follows from the fact that, unlike a dislocation, a crack cannot exist in a material without the presence of "external" stresses. This is easily seen from fig. 5, where the atomic bonds across the cleavage plane will close up the slit unless a counteracting set of external forces is applied. This "external" stress may of course be supplied by misfitting grains, precipitates, etc. in the solid, as well as by actual external forces applied to the surface of the specimen.

Although the atomically sharp brittle crack will play the dominant conceptual role in this chapter, other kinds of cracks also exist in materials. In particular, we will find in the next section that in ductile materials, a sharp crack will open up under appropriate conditions into a wedge shape, with a finite opening angle at its tip. However, it is the sharp crack which possesses the well known dynamical properties of cleavage, while the

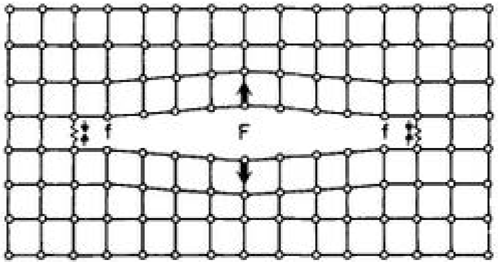


Fig. 5. Diagram of a crack in a two-dimensional square lattice with external forces exerted at the center of the crack. Bonds at the two crack tips are nonlinear. The nonlinear attractive forces at the tip are labeled f, and the external applied forces at the center of the crack are labeled F.

wedge-shaped crack and its derivatives are associated with a complex form of ductile failure. Thus, the sharp crack is the only one which we shall elevate to the status of fundamental crystal defect.

Cracks are formed in materials in a variety of ways by any mechanism which causes a sufficiently high stress, locally. Examples are high-density dislocation structures formed near surfaces during fatigue, stress concentrations in brittle particles due to the strain incompatibility at the interface between the particle and the deforming metal matrix, grain-boundary triple points, etc. In their early stages these cracks in many cases are sharp.

The materials scientist will find in studying this chapter that the problem of fracture deeply involves several other disciplines, and that this fact is both the cause of stimulation because of the breadth of ideas involved, and also daunting because these ideas often involve unfamiliar subjects. In particular, fracture has from the first involved continuum fracture mechanics on a deep level. The fundamental reason why the mechanics of the problem cannot be ignored is that a crack cannot exist in a material without an external stress, and the way in which this stress is transmitted to the crack tip is the subject of *continuum mechanics*. The necessity of the external force is related to the long-ranged $1/\sqrt{r}$ stress field surrounding the crack, which means that the crack is not a localized defect in the same sense as a vacancy or even a dislocation is. Fortunately, for long cracks, the atomic configuration around the tip depends only on a quantity termed the *stress intensity*, K, and not on other mechanical and geometric details, but the way the plastic zone interfaces to the immediate region around the crack tip is both a materials problem and a mechanics problem. Likewise, fracture deeply involves surface physics and chemistry, because the tip region of the crack is an incipient surface, and reactions at the tip with external environments involve ideas and processes which are brought over from those fields. The concept of *fracture mechanics* is defined in § 9.1.

Finally, the question of the structure of the crack tip, which determines such fundamental properties as whether dislocations can be emitted from it, or how it interacts with external chemical species, depends upon the kind of atomic bonding involved in the material. In spite of the large amount of work and substantial progress which has gone into understanding the metallic bond, we still are not in possession of bonding laws or

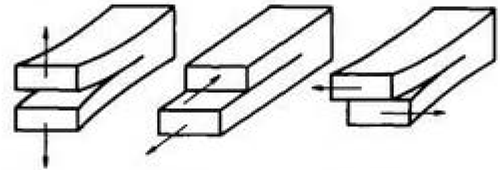


Fig. 6. The three prototype modes of fracture. In each case, the crack is formed by making a planar slit in the material, with the crack line along the edge of the cut. In mode I, the opening mode, the force is exerted normal to the cleavage plane. In mode II, the force is in the plane of the crack normal to the crack line. In mode III, or antiplane strain, the force is in the plane along the crack line. Modes II and III are shear cracks.

extensive use of this analysis to describe results for cracks in a generic sense. Of course, most actual cracks are not pure cases; the fracture surfaces are not flat, and the crack fronts are not straight. Thus, a crack may be produced which is primarily mode II or III, but with enough mode I present to separate the cleavage plane. Needless to say, curved cracks possessing nonplanar fracture surfaces are much more difficult and often impossible to handle mathematically, at least analytically, so that the pure modes listed above will be featured almost exclusively in this discussion.

2.2. Fractographic observations

A desirable steel exhibits high *toughness*, which means that it requires a large energy to pull it apart in fracture. Such a steel has a fracture surface which exhibits a non-crystallographic "fibrous" quality on a microscopic scale as shown in fig. 7 The surface is composed of microdimples, which are the result of holes forming ahead of the main

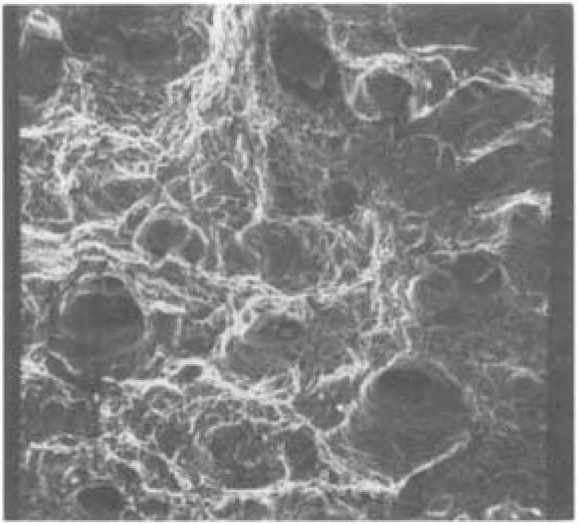


Fig. 7. Ductile fracture surface, showing the final dimpled structure of hole growth. Precipitate particles are visible in some of the dimples which served as nucleation sites for the holes. (Courtesy C. INTERRANTE.)

References: p. 2287.

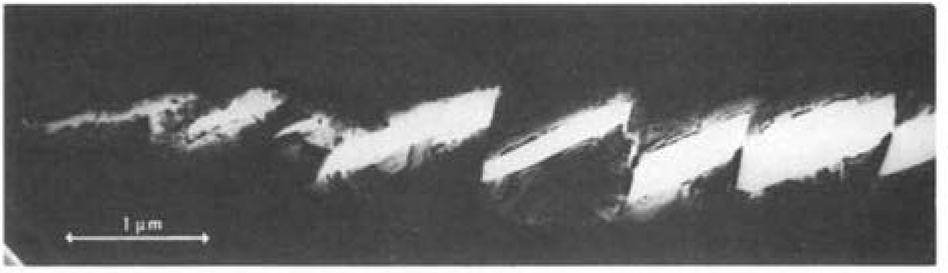


Fig. 8. Side view of a growing fracture in a thin film of gold. Holes are shown nucleating ahead of the crack; see LYLES and WILSDORF [1975]. (Courtesy H. G. F. WILDORF).

crack as shown in fig. 8. These holes are thought to initiate in practical alloy steels primarily at the site of precipitate particles in the matrix as illustrated in fig. 7. Often the large voids in the medium are connected by bands of intense shear, as shown in fig. 9. Within these shear bands, void formation also occurs, but typically at a much smaller average size than in the case of the larger precipitates. The initiating sites of the smaller voids are at the time of writing still a matter of controversy. In some cases, they are associated with precipitates — but in this case particles of very small size (100–1000 Å) (GARRISON and HANDERHAN [1989]). However, holes have been observed by GARDNER *et al.* [1977] to nucleate in very pure metals without the help of precipitates, and the small voids in impure steels may be produced homogeneously also. Figure 10 illustrates two forms of relatively brittle fracture in a high-strength steel under embrittling condi-

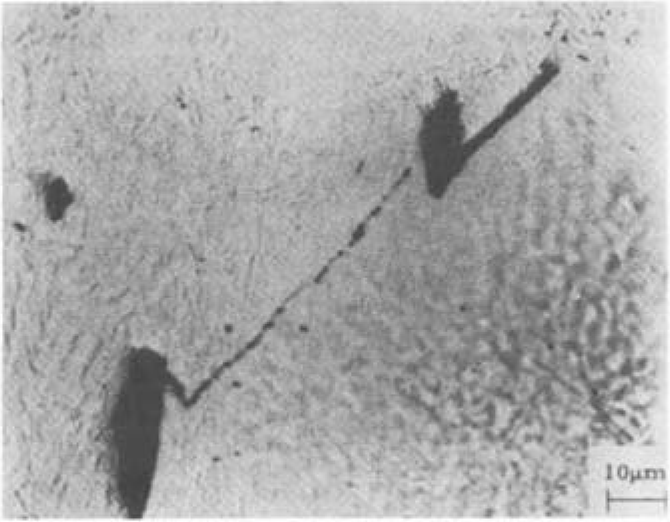


Fig. 9. Hole growth at precipitates in ductile fracture. Between the large holes, shear bands develop consisting of large numbers of much smaller holes. (After COX and LOW [1974]).

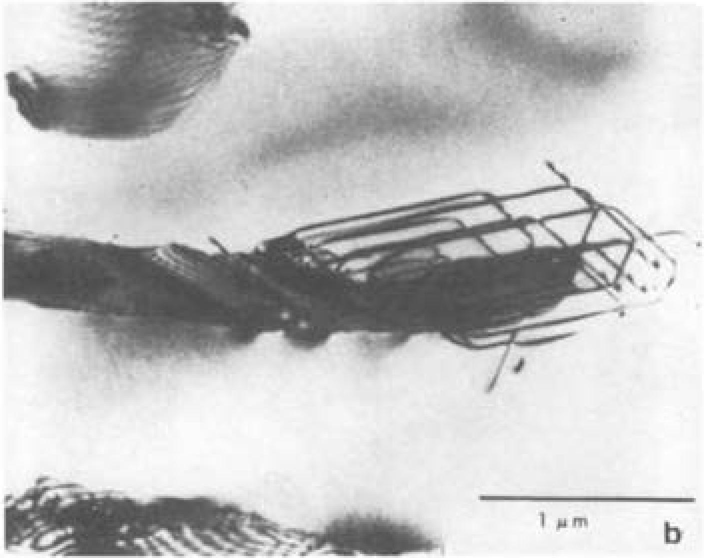
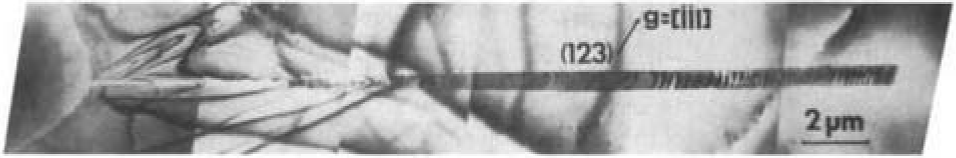


Fig. 11. (a) An electron microscope picture of a fully brittle crack in Si. In the photo, the crack is seen edge on. (b) Cracks formed in Si at 500°C are associated with dislocations as shown (see LAWN *et al.* [1980]). (Photos courtesy B. HOCKEY.)

immobile, by an indentation technique. In the figure, the crack is shown in cross-section. No dislocations whatsoever are present, proving that completely brittle atomically sharp cracks are possible. If cracks are injected into the same material at a temperature of 500°C, where the dislocations are mobile, then considerable dislocation activity is evident (fig. 11b).

This behavior is a particularly simple example of a type of brittle/ductile transition



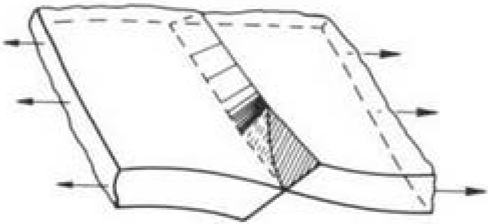


Fig. 13. (a) Dislocations emitted from the tip of a crack in a foil of copper. The crack has grown from the left. Immediately ahead of the crack is a region free of dislocations. The pile-up of dislocations is seen extending on the cleavage-slip plane to the right. In b) the crack-dislocation geometrical configuration is illustrated. (After KOBAYASHI and OHR [1981], courtesy S. M. OHR.)

are obtained for Ni and steel). In this case, the foil "slides off" on its slip plane which is at an oblique angle to the plane of the foil. When the foil locally shears completely through its thickness, a cracklike artifact is generated (a mode III crack of zero length) which grows by dislocation emission and translation. This figure illustrates that the crack has a substantial region near its tip (termed an elastic enclave, or a dislocation-free zone) where no dislocations are present.

2.3. The basis for fracture science

The illustrations selected above exemplify the major experimental findings to be used as a basss for a science of fracture. The reader is referred to vol. 9 of "The Metals Handbook" (SHUBAT [1974]) and to vol. I of the "Fracture" series (LIEBOWITZ [1958]) for a compendium and atlas of fractographic examples which go far beyond what we can display here. The illustrations demonstrate clearly enough, however, that fracture in materials is most often a very complex phenomenon which depends upon a large number of material and external variables, and that the task of sorting out the fundamental phenomena which are critical to the fracture event is not an easy nor straightforward one. However, primarily on the basis of these observations, we shall state a number of qualitative conclusions as a point of departure for the sections which follow:

(a) In practice, cracks are most often flaws of very complex structure which only approximate the three idealized two-dimensional modes of fig. 6. However, because of the necessity of using models which can be analyzed and described in accessible mathematics, we are led to the study of these idealized modes.

(b) The two major poles, of brittle and ductile fracture, are discernible in the fractography of metals (figs. 7 and 10).

(c) The ubiquitous presence of dislocation sources in metals assures that copious dislocations are generated to accompany the fracture process in all cases, even when the material is intrinsically brittle. Only in such exceptional (nonmetallic) cases as Si at low temperatures, where dislocation sources in the medium are nonexistent, are ideally brittle fractures found (fig. 11). Thus, even for metals which can cleave, ductile behavior is possible because of the elastic interactions with the dislocations. This ductility will be highly dependent on the dislocation mobility and source density (and thus on the microstructure of the material), and the material will exhibit an unstable tendency for transition between the ductile and brittle forms. In contrast to the kind of ductility which is controlled by the atomic bonding character of the crack tip, which we have termed intrinsic ductility, this latter form of ductility will be termed extrinsic ductility, because it depends on the material microstructure, as well as load rate and temperature.

(d) Although the extreme pole of ductile failure is the macroscopic rupture by necking depicted in fig. 1, a more localized form of ductile failure is also possible which has the macroscopic appearance of a fracture. Pure forms of such ductile fracture occur when dislocation emission at the crack tip generates a mode I wedge-shaped crack (fig. 12) or a mode III shear crack (fig. 13), or a sharp crack is rounded by the action of a ductile matrix. In all these cases again, the dislocation interactions with the underlying "crack" should be a key to the overall behavior. However, the criterion for dislocation emission, in the form of a local "fracture criterion" at the crack tip, will involve different physical ideas from the fracture criterion for a sharp cleavage crack as envisioned in the last paragraph. Such a treatment of ductile fracture has been developed for the mode III shear crack, and will be presented in the following sections, but the wedge case has not yet been analyzed in these terms.

(e) In practice, however, ductile fracture is characterized by void nucleation and growth as discrete events ahead of the main crack (figs. 7–9). The larger voids are initiated first at precipitates in practical alloys. Shear bands of intense localized shear associated and coincident with sheets of very small voids then connect the large voids. Final failure occurs by purely plastic rupture processes operating between the voids and the main crack. Clearly, this (mixed) form of failure is very complex, and we shall only be able to address it in a qualitative way, and only after development of ideas conforming to the simpler pure forms of brittle and ductile fracture.

(f) A given material can be influenced to respond over a range of the ductile-brittle axis by varying the external environment (fig. 12), the strain rate, or the temperature. Internal metallurgical variables also have the same effect. Thus, we shall be very sensitive in the following sections to the features in the models to be developed, which may lead to the breakdown of either the brittle or ductile poles in favor of the opposite pole of behavior.