PRINCIPLES OF FAILURE ANALYSIS

Ductile and Brittle Fracture

Revised by
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This lesson starts with a discussion of what is meant and implied by the presence of “ductile” or “brittle” fracture in a broken or cracked part. There is a discussion of both macroscale and microscale fractographic features. Macroscale features typically identify the fracture-initiation site, crack-propagation direction, and the fracture-surface orientation, correlating with the nominal loading conditions. The microscale features correlate with the microstructure and identify the mechanism(s) by which fracture has occurred. Microscale features help identify environmental conditions at the time of fracture (e.g., stress-corrosion cracking) as well as heat treating imperfections (e.g., tempered martensite embrittlement).

There is a discussion of pre-existing geometric and microstructural imperfections that may be defects and therefore cause a component to fail. If an imperfection is a defect, the location of the defect may change the crack-initiation site from that expected based on the nominal loading conditions.

Where appropriate, there is also discussion of the importance of microstructural examination to help identify the cause of undesirable cracking. The role of the microstructure in controlling the strength, ductility, and toughness of crystalline metallic materials is fundamental to an understanding of the response of a fabricated component to loads and deformations and ultimately the cause of failure. This lesson presupposes some exposure to, but not expertise in, the examination and interpretation of microstructures. It is assumed that the reader has at least been exposed to the examination and interpretation of microstructural features. Three important microstructural features affecting mechanical behavior are the grain size of the matrix phase, the location of any second-phase particles, and the presence of entrapped foreign particles (inclusions) in the microstructure. This lesson emphasizes the macroscale and microscale features associated with ductile and brittle fractures.

Upon completion of this lesson you should be able to:

- Understand the relationship between fracture-surface orientation and external loading conditions (axial, torsion, bending)
- Understand how to identify macroscale features that identify the crack-initiation site and the crack-propagation direction
- Understand how to identify common crack-initiation mechanisms
- Understand how to identify the microscale crack-propagation mechanisms as ductile or brittle
- Understand the causes for brittle fracture in a component that when tested as a nominally smooth cross section tensile specimen shows ductile behavior
- Understand the errors in heat treating that can cause a material to fail in a brittle manner
- Understand the sources of environmentally induced embrittlement
Ductile and Brittle Fracture Terminology

One important issue in characterizing a fracture is whether the fracture is ductile or brittle. If there is visible permanent change in shape associated with a fractured component, an observer without a materials background would likely describe the fracture as ductile based on visual evidence of plastic deformation prior to fracture. A person with a materials background might just as logically examine the microstructure and, if that examination showed evidence of plastic deformation, also be inclined to describe the fracture as ductile. A macroscale fracture surface without any visual evidence of plastic deformation is likely to be described as brittle by persons with either background. However, microstructural examination of material immediately adjacent to the fracture surface or direct examination of the fracture surface at increased magnification may reveal the presence of local plastic deformation. Similarly, long-life fatigue fracture typically occurs at nominal stress levels less than the yield strength so there is no visible macroscale distortion. This failure condition is often described as brittle on the macroscale. However, the microscale mechanism for fatigue crack initiation and growth in materials showing some tensile ductility is due at least in part to plastic deformation.

Some means of describing the fracture appearance that does not lead to possible confusion based on the scale of observation or background of the observer as to whether the fracture is ductile or brittle and that results in an accurate description of the conditions that resulted in fracture is necessary. One such procedure is to describe the macroscopic appearance as ductile, brittle, or ductile plus brittle, based on whether the fracture surface(s) is (are) on a plane of maximum normal stress (brittle) or high shear stress (ductile) and to describe the local scale mechanism of fracture as ductile or brittle based on the microscale appearance and the microscale mechanisms that operated to produce the fracture. Fracture surfaces perpendicular to a tensile stress are brittle on the macroscale, whereas fracture surfaces inclined to a tensile stress (and therefore on planes of high shear stress) are ductile. Adoption of this terminology can prevent misinterpretation and disagreements between persons of varying backgrounds.

Additional misinterpretation derives from the use of the term mode,* used in multiple ways in discussions of failure. This lesson uses the term mode in only two ways: (1) to distinguish between micromechanisms of fracture (e.g., transgranular versus intergranular, slip versus cleavage) and (2) to distinguish between different macroscale loading conditions (axial, bending, shear).

The determination of the mechanism causing fracture is normally accomplished by examination of the fracture surface at high magnification, usually in the scanning electron microscope (SEM).** Supporting evidence can and should be obtained by examination of the microstructure. Metallographic specimens showing the fracture surface in edge view

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*“Mode” is defined with reference to macroscale loading conditions in ASTM E 399.
**Early fractographic studies were often done by the preparation of a replica of the fracture surface and examination of the replica in the transmission electron microscope (TEM). The literature contains many of these fractographs. In such instances, the image is reversed. The TEM is still of use for the examination of features that exceed the resolution limits of the SEM.
reveal the microstructure immediately adjacent to the fracture surface. Microstructural examination is especially useful in identifying the presence of, and the cause for, intergranular fracture.

At the macroscale, a fracture surface may change orientation as a crack propagates (Figure 1), changing from ductile to brittle or from brittle to ductile (more common). For some loading conditions, there may be multiple crack-initiation sites on planes of both high normal stress and high shear stress. Microscale examination will show that fracture on planes of high shear stress occurred by a ductile mechanism, but fracture on planes of high normal stress may be either ductile or brittle at the microscale. The familiar cup-and-cone fracture associated with tensile loading (Figure 2) is one case where fracture initiates in the flat portion of the cup (macroscale brittle) and subsequently propagates on a shear plane (macroscale and microscale ductile). Microscale observation, however, shows that fracture in the flat portion of the cup occurs by a ductile mechanism as does the fracture on the walls of the cup.

Ductile fracture is typically assumed (sometimes incorrectly) to be associated with high energy absorption and brittle fracture with lower energy absorption, with high energy absorption being associated with “good” behavior. For the fracture process to require high energy, the material must

![Figure 1. Slant and flat fracture surfaces. (a) Flat and macro brittle (plane strain). (b) Macro brittle changing to macro ductile with loss of constraint. (c) Macro brittle changing to general net section yield. Plastic flow on all four shear planes. Source Reference 1, p 129.](image)

![Figure 2. (a) Ductile cup-and-cone fracture with necking in a tensile specimen. (b) Brittle fracture. Source: Reference 2.](image)
be ductile under the loading conditions, have high strength, and have strain-hardening capacity. The non-heat-treatable wrought aluminum alloys (e.g., 11xx and 3xxx) have moduli of toughness (ultimate tensile strength × tensile elongation; an approximation of the toughness of the material when tested as a smooth-cross-section tensile specimen) of 760 to 1930 MPa (110 to 280 ksi), whereas the less ductile aluminum-silicon casting alloys have moduli of toughness of the order of 1035 to 2070 (150 to 300 ksi). Thus, neither group of materials has a very high modulus of toughness, especially in comparison to a medium-carbon steel in the hot-rolled or quenched-and-tempered condition. The presence of a cracklike defect in the wrought aluminum alloys does not dramatically lower the toughness at fracture, and the unmodified aluminum-silicon casting alloys fail in a brittle manner at the macroscale. However, the presence of a cracklike defect can dramatically change the work required to cause fracture in steels. Steels can be described as notch sensitive, whereas the wrought aluminum alloys are not notch sensitive.

In order for a failure analysis to be successful in determining the cause for failure, it is usually necessary to perform multiple types of tests on the material as described in Lesson 1. Conclusions as to the cause of failure should be based on the total information available.

Testing at a minimum typically involves examination of the fracture surface, examination of the macrostructure and microstructure, and some procedure(s) to evaluate mechanical properties. Hardness testing is one way to obtain mechanical property information and can be done in conjunction with microstructural examination. Macroscale hardness tests (Rockwell, Brinell) can be used to estimate nominal strength (yield and ultimate), but they do not provide information about ductility and toughness. Microscale hardness testing can be used to determine possible changes in properties near the surface of the component and is especially useful in evaluating surface-treated steel components to check for decarburization and adequate case depth.

The information obtained from each type of evaluation ideally provides positive information pointing to a single cause for the failure and results ideally from self-consistent, positive information from all tests. Unfortunately, this is not always the case. Sometimes test results from a given procedure are inconclusive or do not point to a single cause for failure. A case in point is the identification of failure due to cyclic loading (fatigue). In instances in which cyclic loading is interrupted and then reinitiated, loading often leaves characteristic features (known as “beach marks”) visible at the macroscale on the fracture surface. Secondly, in some materials, cyclic loading also leaves characteristic markings at the microscale on the fracture surface (“fatigue striations”). However, it is possible that only one set of markings is present or that neither set of markings is present. Therefore, there may not be any positive fractographic information indicating cyclic loading.

The cause for failure has not been identified if conflicting information is obtained from two different procedures. Common situations involving the inability to obtain information or draw conclusions include:
- Obliteration of microscale features due to oxidation of the fracture surface
- Abusive handling of the fracture surface after failure
- Improper storage of the fracture surface
- Mechanical damage of the fracture surface during the fracture event
- Results from a specific test does not provide information indicating a single cause for failure

Macroscale examination will provide information indicating whether the fracture is ductile or brittle on the macroscale, and it almost always identifies the fracture-initiation site. The orientation of the fracture surface relative to the component geometry together with the crack-initiation site provides information regarding the loading conditions causing failure (Lesson 2). This is extremely important because it may indicate loading conditions different from that assumed in the design process. Macroscale examination often provides information as to whether an observed imperfection was or was not responsible for degrading the strength or ductility of the material. It may identify whether the mechanism of crack propagation changed during the life of the part. It may or may not distinguish between monotonic and cyclic loading. It may provide information regarding the grain size of the material. Changes in fracture-surface appearance can in some instances be related to the magnitude of the loads causing fracture.

Microscale examination of the fracture surface will sometimes distinguish between static loading and cyclic loading. It will identify the microscale mechanisms causing monotonic-loading fracture. It will sometimes identify the action of environmental variables affecting fracture. It will identify the fracture as occurring across the grains (transgranular) or between the grains (intergranular).

Metallographic examination of the macrostructure will identify the presence of manufacturing imperfections such as shrinkage cavities and porosity, as well as the presence of proper grain flow during fabrication, the presence of incomplete fusion during welding, the presence of laps or seams, and the nominal volume fraction of resolved inclusions. Macroscale examination will provide information regarding the general inclusion density and the presence of materials-processing defects such as shrinkage porosity, seams, and laps. Microscale examination will sometimes provide information regarding loading and environmental conditions at the time of fracture and identify the grain size of the material. It may provide information regarding serious errors in heat treatment. Examination of the microstructure adjacent to the fracture surface will distinguish between transgranular and intergranular fracture.

Evaluation of the microstructure may provide information as to whether the mechanical properties assumed in the design are consistent with those in the failed component. This is important because it often provides the basis to determine whether the material properties were those assumed in the design process and/or whether the properties were degraded by the service conditions.
These examination procedures and the results obtained from them are discussed in more detail throughout this lesson.

In order for the failure analysis to be successful, the root cause must be identified. Many times, material imperfections are observed in a failed component, but, unless failure can be shown to occur because of the imperfection, the imperfection is not a defect and the cause of failure. Figure 3 (railroad coupler) shows macroscale porosity in a component, but failure did not initiate at the imperfection. Therefore, this imperfection is not a defect.

One common thread that runs through many failure analyses is the absence of any (or very much) visible macroscale plastic deformation associated with a fracture surface when the same material shows extensive plastic deformation when tested as a smooth tensile specimen. There are multiple causes for this behavior.

Macroscale brittle fracture may occur because:

- The loading conditions do not permit the material to flow plastically.
- The state of stress (large triaxial tensile stresses) in the material does not permit plastic deformation in a sufficiently large volume of material that it is visible at the macroscale.
- The service environment embrittles the material.
- The material is inherently brittle (perhaps due to faulty heat treating).

Brittle fracture can occur in service without prior plastic deformation at the macroscale (although the material may have been plastically deformed during fabrication) so that there is no warning that fracture is imminent. This may result in catastrophic failure. Ductile tensile overload failures
typically provide some warning that failure is imminent. Proper maintenance procedures will then cause replacement of the part so that fracture is averted. For example, excessive deflection of a shaft can cause accelerated bearing wear that is typically indicated by noisy operation. Permanent set in a spring will cause its replacement. However, overload buckling failures due to elastic stress may provide little warning.

Triaxial tensile stresses inhibit plastic deformation and also elevate the stress at which plastic deformation begins. If the stress to cause plastic deformation is increased sufficiently, it can exceed the stress required to cause fracture. Therefore, it is necessary to understand how triaxial tensile stresses can be developed in a component.

Figure 4(a) shows an element of material subjected to a single tensile load. If the load causes the yield strength to be exceeded, plastic deformation occurs. The material becomes longer in the direction of the load and contracts in the two directions perpendicular to the load. In Figure 4(b), the block of material is subjected to two tensile loads. Plastic deformation is still possible. The material now extends in two directions and contracts in the third unloaded direction. However, if three tensile loads, all of the same magnitude, are applied to the material, the material cannot deform plastically if the volume remains constant; that is, the sum of the plastic strains must be zero (Figure 4c). The loading conditions in Figures 4(a) and (b) are described as plane stress (because all of the load directions lie in a common plane).

A general state of stress acting on the element has two parts: a hydrostatic component and a nonhydrostatic component. The hydrostatic component is defined as the mean value of the three normal stresses. The state of stress in Figure 4(c) is then pure hydrostatic loading.

Consider now a body containing a cracklike defect (Figure 5). There is stress concentration behind the notch, and, because material wants to

![Figure 4. Change in shape of an element of material subjected to loads. (a) Uniaxial loading. Plastic extension parallel to the load, contraction in two directions perpendicular to the load. (b) Biaxial tension gives plastic extension in two directions parallel to the applied load and contraction in the third unloaded direction. (c) Pure hydrostatic loading. If all three loads have the same magnitude, the material cannot plastically deform.](image)
contract perpendicular to the stress in the \( y \) direction, the material behind the notch would like to assume the geometry shown. However, there are no loads applied to the surface of the crack, so whatever state of stress exists inside the body, the material adjacent to the notch has no tendency to contract. The material in this region therefore exerts a restraining force on the stressed material.

The actual behavior of the material when stressed to fracture depends on the size of the unloaded area (section thickness and crack length). Figure 6 shows that as the section thickness or the crack length increases, the stress at yield is increased, and the tensile elongation at fracture is reduced. A similar effect occurs as the radius of the crack tip is reduced. Measurement of the lateral contraction (\( z \) direction) of these specimens in Figure 5 shows that as the size of the unloaded (notched) area is increased,

**Figure 5.** Constraint in an axially loaded member that contains a crack-like defect. As the notch severity increases (thicker material, longer crack, sharper crack tip radius), the stress at fracture increases but the strain at fracture decreases.

**Figure 6.** Tensile stress-strain curves for a specimen containing a notch of increasing severity (increase in constraint). Note that the yield strength and tensile strength increase and the fracture strain decreases as notch severity increases.
the lateral contraction becomes very small. One can then say that the specimen failed in **plane-strain loading** conditions. That is, because there is minimal lateral contraction, the strain directions lie in a plane. The loss of lateral contraction (plastic strain) results in **constraint** in the specimen.

Although there is minimal strain in the thickness direction, there is a stress in that direction. This tensile stress must exist to prevent the material from contracting. Similarly, no force was applied in the $x$ direction (perpendicular to the crack front). If the specimen does not contract in the $x$ direction due to the load in the $y$ direction, there must again be an internal stress in the $x$ direction to prevent contraction. Consequently a large **hydrostatic stress** can exist in the interior of the specimen behind the notch if the thickness or crack length is large. This large hydrostatic stress then causes the material to behave in a brittle manner at the macroscale.

The change from macroscale ductile (plane-stress) to brittle (plane-strain) behavior can be predicted and quantified based on the **stress-intensity factor**, $K$. The stress-intensity factor is given as:

$$K = S \sqrt{\pi a} \ [Y]$$

(Equation 1)

where $S$ is the nominal stress on the specimen, $a$ is the crack length, and $Y^*$ is a geometry correction factor that is tabulated in handbooks similar to the tabulation of stress-concentration factors. For example, if the ratio of the crack length to the specimen width is less than about 0.13, the correction factor $Y$ can be taken as 1.12 for a single-edge notch in a plate component. For a through-thickness-center-line crack in the plate, $Y$ can be taken as unity for a crack length-to-specimen width ratio less than 0.4. Correction factors are given for common geometries in many books on mechanical behavior. The most complete reference is Tada et al. (Reference 3).

If $K$ at failure is calculated from the loading data and plotted against either the initial crack length, or the section thickness, Figure 7 results.

*Figure 7. Variation in stress intensity at fracture versus section thickness ($B$) or crack length ($a$). At large crack length or section thickness $K$ at fracture becomes independent of these variables and becomes a material property, the plane-strain fracture toughness ($K_{IC}$).*

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*Note: three standard symbols are used for the geometric correction factor “$Q$,” “$Y$,” and “[$f(a/W)$]” that are tabulated in the literature. $Q$ is used for partial thickness surface and embedded flaws. Both $Y$ and $f(a/W)$ are used for other geometries. However, some care is required in using the tables to determine whether the $r$ in Equation 1 is or is not included in $Y$. 

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The figure shows that $K$ at fracture becomes constant after some minimum thickness or crack length. Measurements on the broken specimen show that $K$ becomes constant when the lateral contraction behind the notch reaches a minimum value, so that the deformation becomes plane strain. This critical value of $K$ at fracture is known as the **plane-strain fracture toughness** ($K_{IC}$). $K_{IC}$ is a material property and depends on alloy composition and microstructure.

Toughness measured in this way predicts behavior of the material when it contains a cracklike defect, whereas the modulus of toughness (ultimate tensile strength $\times$ tensile elongation) is a measure of toughness when the cross section does not contain a stress concentrator. Significantly, many materials are notch sensitive so that there is a significant difference in energy absorbed at fracture between notched and unnotched specimens.

If the geometry of the crack and component creates plane-strain loading conditions or if the material is inherently brittle, the value of the stress causing fracture can be calculated if $K_{IC}$ and the size of the cracklike imperfection are known:

$$S = \frac{K_{IC}}{Y\sqrt{\pi a}}$$

(Equation 2)

This is important because it permits evaluation of a failed component to determine if it was loaded above the design stress.

Alternatively, if the magnitude of the **service load** is known, it is possible to determine whether the material was improperly processed and, as a result, had a fracture toughness less than that assumed in the design.

$$K_{crit} = SY\sqrt{\pi a}$$

(Equation 3)

If fracture occurs for $K_{crit}$ less than $K_{IC}$, the mechanical properties are not those assumed in the design. That is, the microstructure (including inclusion count) and/or alloy composition is different from that assumed.

Finally, if the loading conditions and $K_{IC}$ are known, a critical crack length can be calculated and used in conjunction with **nondestructive evaluation** to determine if a component contains imperfections that exceed the critical crack length.

$$a_{crit} = \frac{(K_{IC})^2}{S^2Y^2\pi}$$

(Equation 4)

If the measured value of the crack length on the fracture surface is less than $a_{crit}$ as calculated from the material properties and loading conditions, either the applied loads or the fracture toughness is different from that assumed in the design. Critical crack lengths are discussed in more detail later in this lesson.

In the case where the cracklike defect goes completely through the section thickness, it is possible to estimate the section thickness and/or
required crack length to cause macroscale brittle fracture. Plane-strain (macroscale brittle) fracture will occur if:

\[ a_B \geq 2.5 \left( \frac{K_{IC}}{S} \right)^2 \]  

(Equation 5)

where \( a \) is the crack length, \( B \) is the section thickness, \( S \) is the nominal stress, and \( K_{IC} \) is the plane-strain fracture toughness. This criterion is conservative in that plane strain will always be met when these conditions are satisfied but may be met for a smaller thickness.

Materials having a body-centered-cubic (bcc) arrangement of atoms (including steels) and some materials having a hexagonal-close-packed (hcp) lattice show a typical S-shaped curve of toughness versus temperature, whereas materials having a face-centered-cubic (fcc) lattice show a gradual decrease in toughness with decreasing temperature (Figure 8). Values of interest on the curve include the maximum toughness on the upper shelf, the temperature at which the toughness curve begins to rise dramatically, and the minimum toughness at low temperature. Before 1960 (and to some extent still today), the most common way of measuring fracture toughness was the use of pendulum impact tests such as the Charpy and Izod tests (discussed later in this lesson). These tests are still useful, especially to evaluate the change in toughness with changes in microstructure or composition. However, the Charpy and Izod tests do not discriminate differences in toughness below the toe of the curve and therefore do not discriminate between different sizes of imperfections. Additionally, it is not possible to extract a value of the stress at which fracture occurred in these pendulum impact tests. Therefore, it is not possible using pendulum impact data to determine whether a component was loaded above the design stress or to determine whether an imperfection in the material was of sufficient size or in a critical location to become a defect. The list of possible imperfections is quite lengthy, but it includes geometric imperfections such as surface nicks and gouges, metallurgical imperfections such as lack of penetration in a weldment, shrinkage porosity in a casting, and quench cracks. There is a discussion of crack-initiation sites in a later section of this lesson.

**Thermal Effects**

![Figure 8. Variation in fracture toughness with temperature for (A) bcc materials (including steels) and (B) fcc materials (including aluminum alloys and austenitic stainless steel).](image)
The fracture mechanics approach can often permit determination of critical stresses and/or crack lengths.

The curve in Figure 9 shows the behavior of three different steels or the same steel processed in three different ways. Correct mechanical/thermal processing usually does not result in room temperature corresponding to upper-shelf toughness (curve A). An exception is the 5 and 9% Ni ferritic steels. Curve B is typical of many properly processed materials, whereas curve C would be typical of an improper material selection or improper microstructure. Improper heat treating can result in shifting the toughness curve from A to B or A to C so that the fracture toughness at room temperature is low. As carbon content increases, the trend is to shift the curve from A or B to C and also to decrease the upper-shelf toughness.

Data Trends for Steels

The fracture toughness of steels varies widely depending on many composition and microstructural variables. However, the general trend is for fracture toughness to decrease as the yield strength increases (Figure 10). Qualitatively, for a medium-carbon steel (heat treatable), the fracture toughness of tempered martensitic microstructures is higher than that of bainitic microstructures, which in turn is higher than that of pearlitic microstructures. However, improper tempering of martensitic microstructures may reduce fracture toughness to very low values. The toughness of hot-rolled pearlitic microstructures can be improved with a normalizing heat treatment (grain-size reduction) or a stress-relief anneal. Most interest in measuring $K_{ic}$ has been for higher-strength steels used in critical design situations, whereas steels for less-critical applications are more commonly evaluated via Charpy impact testing. Examination of data in Reference 4 shows that the room-temperature value of $K_{ic}$ seldom exceeds 130 MPa$\sqrt{m}$ (120 ksi$\sqrt{in}$.), and, more importantly, the square of the ratio of fracture toughness to yield strength seldom exceeds 2 for the materials shown in Figure 11. However, there are exceptions and $K_{ic}$ for lower-carbon (0.2% C) pearlitic steels can exceed 220 MPa$\sqrt{m}$ (200 ksi$\sqrt{in}$.). If the squared ratio of toughness to yield

![Figure 9](image-url)

Figure 9. Variation in fracture toughness for three different steels or the same steel with three different microstructures. RT, room temperature. See text for discussion.
Figure 10. Typical variation in $K_{ic}$ with yield strength. Source: Reference 4.

Figure 11. Correlation between $(fracture\ toughness/yield\ strength)^2$ and Charpy impact strength on the upper shelf. Source: Reference 4.
strength is 2, Equation 5 indicates that the critical crack length or section thickness at room temperature is:

$$a_B = 2.5(2) = 5 \text{ in.} = 125 \text{ mm}$$  
(Equation 6)

Thus, brittle fracture is not expected except for large sections or long cracks. At high strength levels, the required section thickness or crack length to obtain brittle fracture can be quite small, even at room temperature. For example, for SAE 4340 heat treated to 51 HRC, reference values of the yield strength and toughness are 1517 MPa (220 ksi) and 57 MPa√m (50 ksi√in.) so that the critical crack or section thickness is:

$$a_B = 2.5 \left( \frac{57}{1517} \right)^{\frac{1}{2}} = 0.0035 \text{ m} = 0.139 \text{ in.}$$  
(Equation 7)

At high strength levels or after improper heat treatment of steels, the critical crack length can be as small as 50 μm (2 mils). Because the critical crack length can be so small, it is not surprising that brittle fracture can be initiated by small manufacturing imperfections (a too-sharp keyway or fillet, threads, and splines), scratches, and gouges due to improper handling or a poor weld bead contour.

The ductile-brittle transition temperature (DBTT) associated with the bcc lattice is sensitive to the deoxidation practice used as well as to the nominal composition and microstructure, including grain size. For hot-rolled pearlitic steels, a reduction in grain size can decrease the DBTT dramatically, with changes of 26 to 56 °C (50 to 100 °F) possible. Rimming-grade steels have a higher DBTT than an alloy of the same composition that has been killed (aluminum deoxidized). The lowest toughness is associated with untempered martensite. Properly tempered, a quenched-and-tempered martensitic microstructure optimizes the combination of yield strength and fracture toughness and provides the lowest DBTT temperature. However, it is possible to temper a steel improperly, resulting in a degradation of toughness and an increase in the DBTT as discussed in the section on heat treating imperfections later in this lesson. Upper bainite has poorer toughness and a higher DBTT than lower bainite.

The most common technique to strengthen traditional steels involves an increase in the carbon content. Unfortunately, this approach also raises the DBTT and lowers the upper-shelf toughness. Sulfur and phosphorus are two trace-level impurities that also raise the DBTT. Most alloying elements raise the DBTT (less dramatically than phosphorus and sulfur), exceptions being nickel and manganese. Nickel additions up to about 8% are effective in lowering the DBTT (23xx and 25xx steels).

Manganese has a beneficial effect because it removes sulfur from solid solution as manganese sulfide (MnS). It is common to control the manganese-to-carbon ratio from about 3-to-1 to 7-to-1 for this purpose. Higher manganese contents are not used because of the lowering of the martensite start temperature and associated heat treating problems due to retained austenite. Additionally, as discussed later, a loss of toughness and an increase in the DBTT is associated with the presence of trace levels of
arsenic, antimony, tin, and phosphorus in conjunction with manganese in quenched-and-tempered steels. Steels having less than 0.5% Mn are less susceptible to this latter type of embrittlement.

For wrought, hardenable, high-strength aluminum alloys (2xxx, 7xxx), the toughness \((K_{IC})\)-to-yield strength ratio varies from about \(\frac{1}{3}\) to \(\frac{1}{2}\), so the required section thickness for plane-strain fracture is about 7 to 15 mm (0.28 to 0.6 in.). Non-heat-treatable, wrought aluminum alloys (1xxx, 3xxx, 5xxx) as used commercially do not fail in plane strain and have high fracture toughness. However, the aluminum-silicon casting alloys (heat treatable and non-heat treatable) do not have high fracture toughness. Toughness is improved by modifying these alloys to change the morphology of the silicon phase from platelike to spheroidal, but \(K_{IC}\) does not typically exceed 11 to 16.5 MPa\(\sqrt{m}\) (10 to 15 ksi\(\sqrt{in.}\)). Optimal toughness in the hardening alloys is obtained by overaging (e.g., the T73 temper).

Primary metalworking practice—rolling, extrusion, forging, and so forth—typically results in a nonhomogeneous microstructure in which microscale bands of varying composition are parallel to the direction of greatest flow (banding). Additionally, ductile inclusions have high aspect ratios after forming, and there may be strings of oxide “beads” in the flow direction. Bands of varying composition may result in a variation of microstructural constituents in the bands (Figure 12). The net effect is to cause considerable variation in ductility and fracture toughness (and fatigue crack propagation rates) with loading direction. This directionality has received considerable attention in the steel industry and has resulted in calcium ladle additions to minimize the directionality. Ladle additions of calcium to electric furnace steel reduce the number of stringer-type MnS and oxide inclusions, and such additions also permit “shape control” resulting in spheroidal inclusions (complex sulfide-aluminates). For sufficiently thick material, reduction in area obtained from a tensile specimen

### Aluminum Alloys

#### Directionality of Properties

**Figure 12.** Microscale segregation (banding) in steel results in microstructural bands of pearlite and ferrite steel after slow cooling from above the upper critical temperature. Banding is minimized with faster cooling rates. Banding leads to variation in mechanical properties with specimen orientation. Source: Reference 5.
taken parallel to the thickness direction can be compared to that from a specimen taken parallel to the rolling direction or transverse direction to evaluate the degree of anisotropy. Alternatively, Charpy or compact tension specimens can be used. For conventional melt practice, the difference in upper-shelf Charpy energy for a crack propagating in the rolling direction versus a crack propagating in the transverse direction can be as much as a factor of 5 (Reference 6).

If the orientation of the fracture surface of plate specimens of increasing thickness is considered, it is observed that specimens having a short crack or small section thickness fail on a plane of maximum shear stress (Figure 13a) (macroscale ductile). Very thick specimens or specimens having a long initial flaw fail on a plane normal to the applied load (macroscale brittle) (Figure 13c). Intermediate values of section thickness produce a multiple-orientation fracture surface in which the interior region is perpendicular to the load, but the regions closer to the edge fail on a shear plane (Figure 13b). The slant fracture regions near the surface are termed shear lips. The presence of shear lips is an important macroscale fractographic feature because it identifies the direction of crack propagation. The crack runs parallel to the length of the shear lips. That is, a crack propagating perpendicular to the applied load toward a free surface creates a shear lip that has length and width. The direction of crack propagation in this case is perpendicular to the length and parallel to the width of the shear lip. Additionally, presence of shear lips indicates incomplete constraint, so that the stress intensity at fracture is greater than $K_{IC}$. As discussed below, there are additional fractographic features that identify the specific site of crack initiation.

When sections are small enough that the local stress field at an imperfection extends to the outer surface of the material, the crack plane may diverge from its initial orientation. Figure 14 shows the effect of inadequate height of the specimen so that net section yield occurs. A similar curvature of the crack plane is obtained for an inadequate component width.
Brittle Fracture in a “Ductile” Material

Figure 14. When the specimen has inadequate height (or width) so that the volume of material in which the stress distribution predicted by the stress-intensity factor extends to the surface of the specimen, the crack plane curves as shown. Source: Reference 7, p 199.

Two closely related macroscale fractographic features that identify the location of crack initiation are radial marks and chevrons. These features are created by microscale brittle fracture (cleavage). A third feature that also indicates crack growth direction can be described as a “ridged pattern,” which is formed by a ductile process. However, in addition the term “ridged pattern” is used to describe surface waviness that is created by microstructural features (i.e., alternating plates of cementite and ferrite in pearlite. In some cases, the term “ridge pattern” is used to generically describe any of these patterns. Figure 15 shows a ridged pattern created by a ductile process (and should not be described as a radial pattern). Figure 16 shows a set of chevrons. The fanlike array of radial marks points back to the initiation site and the “V,” or arrowhead, of the chevrons also points back to the initiation site.

The coursness and definition of the ridge pattern depends on the general strength level of the material and the test temperature. It has been proposed that the pattern is developed where there is “rapid” crack propagation (Reference 50). Compare the ridge pattern shown in Figures 15, 17, and 18. The coarse ridged pattern in Figure 15, 18(b), and 18(c) is created by plastic flow, where as the fine ridged pattern in Figure 18(a) is created...
by cleavage. When cracks propagate faster in the interior of a section than at the surface by a brittle mechanism, the result is chevrons (Figure 16). When crack propagation is faster at the surface than at the interior, only one side of the “V” is present, and the feature is the (a) set of radial lines. Ridge patterns formed by ductile processes are visible with the naked eye. Radial patterns (brittle) can usually be seen with the naked eye, but sometimes a 5× or 10× hand lens is required.

Figure 18 shows the fracture surface of a set of steel tensile specimens broken at successively higher temperatures. At the lowest temperature (below the DBTT), fracture is predominantly by cleavage creating a fine
Brittle Fracture in a “Ductile” Material

Figure 17. Faint ridge pattern in a 250 grade maraging steel (ultimate tensile strength = 1725 MPa, or 250 ksi). Specimen originally loaded in fatigue to produce crack in light colored region around the periphery. Specimen then loaded axially. Fracture initiated from 11 o’clock to 4 o’clock and grew toward the dark region on the left at about 9 o’clock. Note the increase in roughness as the crack propagates. Source: Reference 9, p 22.

Figure 18. Fracture surface appearance of steel tensile specimens at increasing temperatures. The fracture surface consists of three zones; an inner fibrous zone nominally perpendicular to the specimen axis, a “radial” zone containing a ridged structure, and a shear zone surrounding the radial zone. Depending on the temperature, the size of these zones changes and zones may disappear. There are accompanying changes in the reduction in area. (a) Tested at −160 °C (−256 °F). (b) Temperature not given. (c) 80 °C (176 °F). (d) 160 °C (320 °F). Source: Reference 10.

radial pattern as in specimen (a); there is essentially no shear lip, no reduction in area, and no fibrous zone. The specimen in (b) shows a well-developed ridged pattern, a small fibrous zone, and a small shear lip zone. There is still little reduction in area. The specimen in (c) shows a course ridged pattern, plus a greater reduction in area and a larger fibrous zone than the specimen in
(b). Finally, in specimen (d) the ridge pattern has disappeared, the reduction in area is large, and the fracture surface consists of a central fibrous region (largest of the four specimens) and a large shear zone.

Identification of the fracture-initiation site is an important step in determining cause for failure, and documentation of the fracture-initiation site is important in preparation of a report. Therefore, it is important that photographs show the presence of these marks. They may not be visible if incorrect lighting is used. Qualitatively, the smoother the fracture surface, the more oblique the lighting on the specimen must be to create the necessary relief contrast. One procedure that often works well is to use one light perpendicular to the specimen together with a high-intensity lamp at an oblique angle. By moving the high-intensity light around the specimen, an optimal location to best bring out the ridge pattern can be obtained. More than one lighting arrangement may be necessary to identify clearly all of the information visible on the fracture surface. It is critical to have positive, nonconfusing evidence to support stated conclusions.

The location of the crack-initiation site sometimes provides information as to whether failure occurred because of loading above the values used in design or whether the material contained imperfections that caused the material to fail at or below the design loads. Constraint develop first at midthickness of an edge or center-notched member for axial and bending loading. Consequently, in a specimen containing a similar cracklike defect, the crack-initiation site is located near the centerline in opening-mode loading. If the fractured component shows crack initiation very close to one side, the implication is that some other condition moved the crack-initiation site away from the location predicted by macroscale stress analysis. This could be a second local stress concentrator superimposed on the macroscale, cracklike defect or could be a region containing a microstructural imperfection. Alternatively, the loading conditions may not have been simple mode I.

Several tests have been developed to determine the effect of temperature on fracture toughness. They include the Charpy (three-point bending) and Izod (cantilever bending) pendulum-impact tests, the slow-bend Charpy test, the drop-weight test (which defines the nil-ductility temperature), and the Robertson test. Each test has advantages and disadvantages in correlations with service performance including section size effects, difficulty and expense of specimen preparation, testing, and so forth. This lesson discusses only the Charpy test, but discussions of the other tests are available in the literature.

The pendulum-impact test has been used for many years to evaluate the toughness of materials, especially steels. In this test, a notched steel specimen is impacted by a falling weight. Either absorbed energy, percent shear fracture, or lateral expansion are measured as a function of the test temperature. If impact energy is measured, two data points of interest are the temperature at which the toughness first starts to rapidly increase (often indexed at 13.5 J, or 10 ft \( \cdot \) lbf) and the toughness on the upper shelf. Unfortunately, the ability of the test to distinguish differences in behavior due to composition and/or microstructures at low tem-
Brittle Fracture in a “Ductile” Material

Loading Rate

The curves of all steels show 1.35 to 6.78 J (1 to 5 ft-lbf) of absorbed energy at low temperatures. However, the 13.5 J (10 ft-lbf) (or sometimes 20.3 J, or 15 ft-lbf) transition temperature and the upper-shelf toughness are sensitive to microstructure and composition. Consequently, this test is a standard way to evaluate differences in toughness for these variables.

It is difficult to use data from the Charpy and Izod tests to predict the fracture stress or fracture toughness of a failed component having a different geometry. The loading rate in the test is high relative to the loading rate in many service conditions, and the crack geometry relative to the section size is often significantly different from that of the failed component. Furthermore, there is no way to extract from the data the stress at the time of fracture. However, the stress-intensity factor shows the same variation with test temperature as the pendulum-impact data, and results from this test do provide a value of stress at the time of fracture if the fracture occurred under essentially plane-strain conditions. (Values of the stress at fracture when plane-strain conditions are not met can also be obtained, but are beyond the scope of this lesson.) Also, stress-intensity factor data do not lose their sensitivity at low temperatures. There are empirical correlations in the literature to convert Charpy data to fracture toughness data ($K_{IC}$) for structural grade steels. The collected results of that work and other correlations are available in Roberts and Newton (Reference 11) and in Hertzberg (Reference 16).

Another variable of importance in determining toughness of a material is the loading rate. An increased loading rate raises the yield strength of the material and also increases the DBTT (Figure 19). The shift in DBTT tends to decrease as the yield strength of the steel increases.

![Figure 19. Change in fracture toughness with loading rate for an A572 steel. $\beta$ is the cutoff for plane-strain fracture. Source: Reference 12, p 118.](image)
The preceding discussion shows the importance of loading conditions, specimen geometry, and test temperature in causing a change from macroscale ductile to brittle fracture. If cracklike imperfections are present in the material, geometric constraint can develop high triaxial stresses near the defect and result in macro brittle fracture. Cracklike imperfections may be geometric in nature (changes in cross section resulting in regions of high stress concentration, nicks, and gouges in the part). They may also be laps and seams that formed during prior plastic forming, or they may be metallurgical in nature such as quench cracks.

Two other variables of importance are the service temperature and the loading rate. Larger sections and higher loading rates increase the temperature at which the ductile-brittle fracture transition occurs. As discussed previously, two other important variables in controlling the toughness and the DBTT are composition (especially nickel, manganese, carbon, sulfur, phosphorus) and microstructure.

Single crystals of metallic materials may deform plastically by two processes: slip and deformation (mechanical) twinning. They may fracture due to continued plastic deformation due to slip or fracture in a brittle way due to an elastic stress by cleavage. Slip and deformation twinning are shear processes and occur on specific crystallographic planes in specific directions when a shear stress reaches a critical value (Figure 20). Cleavage also occurs on a specific crystallographic plane, but occurs when a sufficiently high normal stress is obtained on the cleavage plane.

**Summary**

Microscale Deformation and Fracture Mechanisms

**Figure 20.** The differences in surface topology created by differences in atom shear for (a) slip and (b) twinning. Source: Reference 13, p 205.

**Figure 21.** Cleavage cracking in a molybdenum single crystal at room temperature. Source: Reference 13, p 438.
The very small amount of dislocation motion, and therefore, plastic deformation, that must occur to grow a cleavage-crack nucleus to critical length is neglected here.

The term “twin” refers to a particular atomic arrangement and stacking sequence of atoms. This arrangement may be obtained by either mechanical (deformation twins) or thermal processing (annealing or growth twins). Growth twins are common in materials having a fcc lattice except aluminum and its alloys. They do not form in bcc or hcp lattices. Figure 22 shows mechanical twining in a 26Cr-1Mo ferritic stainless steel that was explosively loaded (expansion forming). Note that these deformation twins have the shape of a lens. This is in contrast to annealing or growth twins that have parallel straight sides (Figure 23a).

There are three common arrangements of atom sites in metallic materials—fcc, bcc, and hcp (Figure 24), and, in metallic materials, there is usually only one atom per lattice site. Metals and alloys that have a fcc lattice (aluminum, copper, nickel, gold, silver, and their alloys, austenitic stainless steels) deform easily by slip but do not mechanically twin (Figure 25) except possibly under extreme conditions of impact loading rates at low temperature, nor do they fracture by cleavage (Figure 8). Therefore, brittle fracture by cleavage in a benign environment is not a possible fracture
Figure 23. (a) **Annealing twins** in 70–30 cartridge brass revealed by contrast etching. Annealing twins are parallel sided and generally extend completely across the grain. (b) Same material after annealing and then cold rolling 20%. Note the curvature in some of the twins. Courtesy of Gene Stansbury.

Figure 24. The three most common crystal structures of metallic materials. (a) Face-centered cubic (fcc). (b) Body-centered cubic (bcc). (c) **Hexagonal close packed (hcp)**.

Figure 25. Prevalence of growth twins, deformation twins, and cleavage with crystal lattice, temperature, and loading rate.
Some researchers have argued that cleavage does occur in highly nitrogenated austenitic stainless steels and in austenitic stainless steels subjected to SCC conditions. Metals and alloys having a hcp lattice (zinc, cadmium, magnesium, titanium, zirconium, beryllium, and their alloys) deform easily by slip, but also mechanically twin easily. Additionally, they may fracture by cleavage at relatively low loading rates and at moderate fractions of the melting point (e.g., room temperature). (Some of these materials deform so easily by deformation twinning that artifact twins can be introduced in metallographic specimens by the polishing pressure.) Metals and alloys having a bcc lattice (iron, niobium, chromium, molybdenum, tantalum, tungsten, and their alloys) deform by slip, deformation twinning at higher strain rates and/or lower temperatures than the hcp materials, and they may also cleave, especially at low temperatures and high loading rates. A specimen of ingot iron will mechanically twin at room temperature with a hammer blow and will mechanically twin at lower temperatures at strain rates encountered in tensile testing (e.g., 0.01 in./in./min). Although the lattice of the matrix phase of common engineering alloys is fcc, bcc, or hcp, second phases created by alloying tend to have less symmetrical lattices and are typically brittle and fail by cleavage. Cleavage cracking in second phases then provides a potential mechanism for crack initiation and propagation in two-phase alloys: cleavage-crack initiation in large second phases and propagation by ductile crack coalescence through the matrix. This fracture mechanism is observed in some aluminum alloys.

The relative magnitude of the required shear stress to cause slip or twinning and the magnitude of the normal stress required to cause cleavage depends on several factors including crystal lattice, temperature, and strain rate as noted above. An understanding of these factors is important because it precludes improper identification of microscale and macroscale fracture mechanisms and associated fractographic features. For example, fcc austenitic stainless steels do not fail by cleavage in a benign atmosphere (Figure 25). There is a particle/phase size dependence for cleavage-crack initiation. Larger particles are more prone to cleave than smaller particles. This is likely because more dislocations can pile up at a grain boundary in a larger-grained matrix or within larger phases to create the cleavage-crack nucleus.

The temperature dependence of deformation by slip is high in the bcc materials as well as in hcp titanium and zirconium and lower in the fcc materials and other hcp materials. A large temperature dependence suggests that an increased loading rate can cause fracture to change from a ductile to a brittle mechanism at a given temperature (Figure 26). The temperature dependence of twinning is also probably high, but the normal stress required for cleavage is not a strong function of temperature.

Polycrystalline materials (most, but not all commercial alloys) may deform or fracture by the above mechanisms (transgranular, or TG, deformation or fracture), but they may also deform and fracture in the grain boundaries (intergranular, or IG, deformation or fracture). The conditions required for mechanical twinning are more severe (higher loading rates
Figure 26. Variation in stress to cause slip and to cause cleavage versus temperature for (a) fcc materials and (b) bcc materials. Also shown is the effect of stress concentration and/or loading rate on raising the slip stress. The flow stress for fcc materials cannot be increased sufficiently to cause cleavage. An increase in the flow stress for bcc materials raises the DBTT.

and/or lower temperatures) than for single crystals. In contrast to slip, the microscale plastic strain resulting from twinning is small (less than 10% in thehcp alloys, larger in bcc irons). Furthermore, twinning density in the microstructure is often relatively low. Twinning deformation then does not typically result in macroscale visible permanent deformation in polycrystalline material. In addition, the stress required for twinning in polycrystalline materials is sufficiently high that a significant amount of slip normally precedes the onset of twinning.

The presence of deformation twinning on the fracture surface does not necessarily support a conclusion that the material failed at a low temperature or that it was loaded at a high strain rate. This is because a brittle crack propagating by cleavage (and therefore at a high strain rate) can initiate deformation twinning in material immediately in front of the advancing crack. This leads to a characteristic microscale feature on the fracture surface, described as tongues (Figure 27). Alternately, if a metallographic

Figure 27. Tongues on a cleavage fracture surface. Armco iron broken at \(-25^\circ\text{C} (-13^\circ\text{F})\). The light band shows where cleavage followed a twin matrix interface. The black meandering line is a shear step through the thickness of the twin. Transmission electron microscopy plastic carbon replica. Magnification: 3000×. Source: Reference 15.
specimen taken perpendicular to the fracture surface shows evidence of deformation twinning away from the fracture surface (e.g., 15 or 20 grains), it is more likely that the material was in a low-temperature environment and/or subjected to a high loading rate.

As noted above, microscale brittle fracture by transgranular cleavage is not common in fcc alloys. Cleavage, especially in larger-grained materials, can leave smooth, highly reflecting fracture surfaces and can be recognized visually or at low magnification. Sometimes these same smooth, highly reflecting surfaces are revealed in environmentally assisted cracking conditions (see below) and are described as “cleavagelike” in fcc materials.

The presence of plastic deformation immediately adjacent to the fracture surface can sometimes be determined by microstructural observation. Most fcc alloys with the exception of aluminum alloys contain annealing twins. The sides of annealing twins are straight in an annealed material, but they become curved after a small amount of cold work (e.g., 5%) (Figure 23b). This is a sensitive way to detect the presence of plastic deformation, more sensitive than relying on grain shape which requires a larger plastic strain (e.g., 20 to 30%). Additionally, microstructural examination is a simple way to determine whether fracture is transgranular or intergranular.

Commercial alloys typically have a multiphase microstructure. They also contain inclusions that form as a result of deoxidation practice and grain-size control. As a consequence, the processes of deformation and fracture are more complex in these alloys. The second phases in these alloys, with the exception of MnS in steels, typically have complex and/or lower symmetry crystal structures that make them prone to cleavage with little inherent ductility. These second phases may crack under load, providing potential crack nuclei and multiple crack-propagation mechanisms. Inclusions, because of their role in providing crack-nucleation sites for ductile fracture, are discussed separately.

Intergranular fracture is not a common fracture mode in properly processed material in a benign environment. (It can occur, however, at room temperature in high-carbon steels having a pearlitic microstructure.) Intergranular fracture is the common fracture mode for stress-rupture conditions at elevated temperature \( T > 0.4 \ T_{mp} \). With the exception of high-carbon steels just mentioned, IG fracture is often associated with both improper processing of material and some service environments. It may also be associated with precipitation of inclusions during initial solidification or after partial solution and reprecipitation in the grain boundaries after annealing at a high temperature (say in overheated forgings); MnS, nitrides, and carbides are involved. Intergranular fracture has also been reported as a fracture mechanism in stage I fatigue fracture (Reference 16), and it should also be expected for cyclic loading in those microstructures that contain grain-boundary precipitates in conjunction with solute concentration gradients at the boundary or steels embrittled by trace element impurities (Reference 16).

Noncreep IG fracture can be associated with three fundamentally different causes:
Brittle, second-phase particles and/or films in grain boundaries
Fracture where no film is visible and due to impurity atom segregation at the grain boundary
Environmentally induced fracture where there is neither a grain-boundary precipitate or solute segregation

The causes for IG fracture can be categorized as follows (Reference 17): (1) environmentally assisted fracture, (2) errors in heat treatment, and (3) normal practice. Relative to the latter, normal practice could include, for example, fracture in the case of a carburized steel (due at least in part to the large prior-austenite grain size found in this region). Environmentally assisted fracture could be just behavior at elevated temperature, but it could also include interaction of the material with the environment. This would include hydrogen embrittlement, liquid metal embrittlement, oxidation or reduction of second phases in the grain boundaries, radiation embrittlement, and stress-corrosion cracking. These topics are specifically considered later in this lesson. Some specific examples of IG fracture include:

- Grain-boundary carbide films due to eutectoid divorcement in low-carbon steels and grain-boundary hypereutectoid cementite in carburized or hypereutectoid steels
- Iron nitride grain-boundary particles or films in nitrided steels
- Temper embrittlement in heat treated steels due to segregation of phosphorus, antimony, arsenic, or tin
- Grain-boundary carbide precipitation in stainless steels ("sensitization")
- Improperly hardened, high-strength aluminum alloys resulting in coarse, grain-boundary precipitates and a denuded region adjacent to the grain boundary
- Overheating of material during hot forming—can result in excessive grain size and also partial liquation or eutectic melting; can also result in partial solution of second phases and/or inclusions that reprecipitate during cooling
- Embrittlement of molybdenum by interstitials (carbon, nitrogen, oxygen)
- Embrittlement of copper by antimony
- Reduction of Cu₂O in tough pitch copper by hydrogen
- Hydrogen embrittlement by grain-boundary absorption of hydrogen
- Stress-corrosion cracking (sometimes)
- Liquid-metal embrittlement (LME)—e.g., mercury in brass, lithium in 304 stainless steel
- Solid-metal embrittlement (SME)
- Embrittlement due to a large prior austenite grain size in quenched steels (and therefore in the heat-affected zone of a weldment)
Tensile testing of a smooth ductile member is a convenient way to introduce the correlation between material behavior and the resulting appearance of the fracture surface. When a specimen is loaded, it initially deforms elastically and then, at larger loads above the yield strength, deforms by a combination of elastic and plastic deformation. If the load is increased still further, it reaches a maximum value and then decreases to some lower value where fracture occurs (Figure 28). Careful observation will show that below the maximum load, the diameter of the specimen does not change with position along the gage length. (The small-scale change in cross section at Lüder bands is neglected.) However, once the maximum load is exceeded, the specimen diameter is no longer constant along its length, and the specimen is said to have necked (Figure 2). Necking in a fractured, axially loaded member indicates gross overloading.

The actual shape of the load-elongation curve after plastic deformation initiates is controlled by both strain hardening and strain-rate hardening. Strain hardening causes the flow stress to increase with an increase in strain; strain-rate hardening causes the flow stress to increase with an increase in strain rate. Qualitatively, the larger the difference between the yield stress and the tensile strength, the greater the strain-hardening capability. Therefore, materials cold worked prior to testing have little strain-hardening capacity.

The ability to see the neck in the specimen depends on the amount of strain hardening and of strain-rate hardening. Consider first just strain hardening. As the specimen extends, it decreases in cross section (geometric softening), so the actual stress on the cross section is increased. At the same time, the plastically deformed material is strain hardened so that its flow stress is increased. Up to the maximum load, strain hardening is greater than geometric softening, and deformation remains uniform along the length of the specimen. Beyond the load maximum, the increase in flow stress of the geometrically softened material does not increase sufficiently.
to stop flow in the smallest-diameter region, that is, the neck. Flow in regions contiguous to the neck essentially stops. Therefore, the ability to see a neck in a tensile specimen depends on strain hardening. If there is no hardening, necking starts when plastic deformation starts and the strain is not distributed along the length of the specimen.

The strain at the onset of necking correlates well with the magnitude of the strain-hardening exponent \( n \) in \( \sigma = K\varepsilon^n \). If that exponent is small, the length of the neck is decreased. In common metallic engineering materials that are not cold worked the amount of strain hardening is usually large enough to cause the neck to have a length of the order of the diameter of a cylindrical specimen, and it is therefore readily visible.

Most, but not all, materials also strain-rate harden when tested at room temperature. Once necking initiates in the tensile specimen, the diameter in the neck is smaller than the diameter outside the neck, so the strain rate is higher inside the neck than outside the neck \( \varepsilon = (1/\Delta t)(\Delta A/\Delta t) \). The flow stress of this material is therefore increased above that of the material outside of the neck causing the neck to grow along the length of the specimen. Some materials strain-rate soften when tested at room temperature, and this can cause a dramatic change in the appearance of the fracture surface (see below). Strain-rate hardening typically initially increases as service temperature increases above room temperature, but then it decreases at high fractions of the melting point where dynamic recovery and recrystallization occur.

If a necked, but not fractured tensile specimen, is sectioned longitudinally, it is apparent that crack initiation started along the centerline of the specimen on a plane macroscopically normal to the applied load, initially growing outward in a radial direction (Figure 29). Before necking initiates in the specimen, the stress is the same at any location along the gage length. Failure could then initiate at any point anywhere in the specimen. Once necking initiates in the specimen, the stress distribution is no longer constant along the length or across the cross section. Figure 30 shows that both the axial stress and the hydrostatic stress are highest (more constraint) near the centerline of the necked region, which then becomes the predicted and observed macroscale crack-initiation site. After some growth in the transverse plane, the crack turns and runs on a plane of maximum shear stress. This is because geometric constraint is highest along the centerline of the specimen but is reduced at positions closer to the surface of the specimen. Progressive crack growth leads to the familiar cup-and-cone fracture associated with fracture of ductile cylindrical specimens (Figure 2).

The macroscopic appearance of the fracture surface is characterized by a central fibrous zone, a region containing ridge marks and a shear zone (Figure 31). Ridge marks point back to the crack-initiation site and are an important feature for determining these sites. As pointed out earlier, the coarseness and definition of the ridge marks depend on the general strength level of the material, test temperature, and loading rate. Ridge marks are usually visible without magnification, but sometimes somewhat higher magnification (e.g., to 5 diameters) is required. Therefore, low-magnification examination is desirable if the crack-initiation site is not visible with the naked eye.
Figure 29. Longitudinal section through a necked tensile specimen. Fracture has initiated in the center of the specimen normal to the applied load. After some growth on this transverse plane, the crack turns and runs on a plane of high shear stress as shown. Source: Reference 18.

Figure 30. Stress components in a necked tensile specimen. Adapted from Reference 19.
The size of the cup (fibrous zone) and the reduction in area measured on the broken specimen depend in part on the volume fraction of second-phase particles/inclusions (Figure 32). A large volume fraction of second phases results in a small reduction in area and a large fibrous zone. Alternatively, if the volume fraction is low, the reduction in area at fracture can approach very large values; more than 95% as in annealed, commercially

![Figure 31. The ductile fracture of a tensile specimen showing three possible zones. (a) Centerline fibrous. (b) Radial zone (which may or may not be present). (c) Shear zone. Source: Reference 20, p 217.](image)

![Figure 32. The change in reduction of area with volume fraction second phases (inclusions). Source: Reference 21, p 22.](image)
pure aluminum (1100-O). In such a case, the fibrous region of the fracture can become diminishingly small. Such a fracture process is sometimes described as *ductile rupture*, sometimes as *plastic attenuation*. High anisotropy within the matrix may cause the fracture surface to be noncircular, sometimes resulting in a knife-edge fracture (i.e., highly elliptical). If the material is a single-phase alloy with very few second-phase particles, fracture cannot initiate by *microvoid* coalescence. In that case, fracture can be initiated within a “shear band” or, if local heating occurs during deformation, within an “adiabatic shear band.” *Shear bands* and fracture initiation within shear bands are discussed below.

When a tensile specimen containing elongated inclusions is sectioned longitudinally after fracture, it is observed that the inclusions in the shear zone of the fracture have been rotated and now lie parallel to the fracture surface (Figure 33). The width of the band containing the rotated inclusions is quite narrow, and it described as a *shear band*. Shear-band formation—localization of shear strain to a narrow band—is a natural consequence of plastic deformation. Whether shear bands develop and are visible depends on component geometry (that is, the state of stress), material properties, and the total strain at fracture. They are more visible when strain to fracture is increased, as in *compression* loading relative to tensile loading. (See the discussion in Lesson 2 for compression loading.)

Doing work on a specimen by plastic deformation results in a temperature rise in the specimen (about 95% of the energy required to produce the plastic deformation is dissipated as heat). If the heat is not conducted away from the site of deformation, local heating results. This in turn lowers the yield strength and further concentrates the strain. In this case, the shear bands are described as *adiabatic shear bands*. In extreme cases of large section sizes, poor thermal conductivity, and high strain rates,
phase changes can occur within the band. One example is the formation of thin bands of untempered martensite within shear bands in explosively loaded steels.

Although the external shape of the cross section at fracture and the shape of the cupped region are usually both circular, this is not always the case. As noted previously, strong preferred orientation of the grains in the specimen can cause the shape to be noncircular. If the necked region is not circular, it should be expected that the mechanical properties (especially ductility and toughness) vary with orientation in the material.

**Orange Peel**

When a ductile material having a smooth surface is pulled to fracture, there is usually some roughening of the external surface of the specimen adjacent to the fracture that increases as the grain size increases. However, if the grain size of the material is very large, the complete surface of the specimen becomes very rough, and, in fact, individual grains can be seen on the surface of the specimen (Figure 34). This appearance is described as orange peel. There is still some ductility at fracture, but the cup-and-cone appearance characteristic of a small grain size may disappear. The presence of orange peel very likely indicates improper heat treatment.

**Sheet and Plate Specimens**

When prismatic specimens having a large width-to-thickness ratio are loaded in tension, the fracture-surface topography may be different from that shown above for cylindrical specimens. Square specimens show the same features as cylindrical specimens (Figure 35). There is necking in both the width and thickness directions (rather than radial in the cylindrical specimen), and there is a fibrous zone and shear zone. A ridged pattern may or may not be present. For moderate width-to-thickness specimens, necking can still be observed on both the width and thickness faces of the specimen, but final fracture occurs on a plane containing the width direction (Figure 36). There may be only one shear plane in this final fracture.

As the width-to-thickness ratio is increased still further, the orientation of the fracture surface changes to one in which a trough evident on the

![Figure 34. Tensile fracture of a large-grained cartridge brass. The specimen has elongated, but there is no cup-cone fracture. The surface is rough and reveals individual grains (6.35–12.7 mm, or 0.25–0.5 in.). The common name for this surface roughness is orange peel.](image-url)
Microscopic Appearance of the Fracture Surface

The specimen now contains two types of inhomogeneous shear: one described as a “diffuse neck” and the second described as a “local neck” in which shear localization can occur. Bird (Reference 24) has reported that strain localization initiates at the intersection of the diffuse neck and the non-necked region of the material. These local necks then grow inward toward the centerline of the specimen. There are then five possible locations for fracture initiation. Cracking may initiate at the surface of the specimen face of the specimen is inclined to the width direction (Figure 37).

**Figure 35.** Fracture surface of a prismatic tensile specimen showing a central fibrous zone, radial marks, and a shear zone. Source: Reference 23.

**Figure 36.** (a) View of a prismatic steel tensile specimen showing the fracture surface and necking. Shear lips are visible on the fracture surface and there is a small fibrous zone in the center of the specimen (not visible in fractograph). (b) Fracture surface of a 6061-T6 aluminum tensile specimen. Width/thickness ratio = 2/1. Necking and shear lips are visible as is a central fibrous zone. There is no radial zone. Courtesy of Patrick Werner, University of Tennessee.
Figure 37. Ductile fracture within a local neck of a tensile specimen (width/thickness ratio = 13.5). A diffuse neck is also visible. Fracture initiation is within the shear band at the edge of the specimen. (a) OFHC copper. (b) 400 Monel.

within the local neck (four sites) or at the intersection of two shear bands, where the strain is twice as large as it is in a single shear band. Puttick (Reference 25) has shown this crack initiation in aluminum single crystals loaded in tension. Others have reported crack initiation at the intersection of shear bands in compression. Figure 37 shows crack initiation at the edge of the specimen within the shear band.

Figure 38 shows a plate of 7075-T6 aluminum (width-to-thickness ratio = 23.5) that was pulled in tension to fracture. This material has little strain-hardening capacity and very little or negative strain-rate-hardening capacity. All four planes of maximum shear stress are visible on the fracture surface. Ridge marks indicate that at least one crack initiated at the (relatively rough) saw-cut edge of the specimen. Other ridge marks progress from the specimen interior, indicating fracture at the intersection of the shear bands. There is no central fibrous zone. Figure 39 shows an essentially single shear plane of fracture in a cylindrical specimen of the same material. There is a fibrous zone in the center of the specimen, but

Figure 38. Tensile fracture of a 7075-T6 plate (width-to-thickness ratio = 23.5). Fracture has occurred on all four planes of maximum shear stress. There is no fibrous zone. Fracture has initiated in multiple locations: at the (rough machined) edge of the material and at the intersection of the shear bands.
Figure 39. Tensile fracture of a 7075-T6 aluminum alloy. Fracture is predominantly on a single shear plane. There is an irregularly shaped fibrous zone in the center of the specimen.

the shear zone is planar, not conical as in Figure 2. There is no visible diffuse necking, and the axial strain at fracture is about 5%.

Crack initiation occurs within a region of local high stress that is within a region of high macroscale stress. The actual initiation site may be affected by the presence of microscale imperfections. In such a case, fracture initiation may occur at more than one location. For example, in a three-point-loaded beam, the location of macroscale maximum stress is on the tension side of the beam at half the span length (Lesson 2). Local imperfections near this location of maximum nominal stress may result in more than one initiation site near the location of maximum nominal stress. When threaded fasteners fracture, cracking is often observed in adjacent thread roots. In torsional loading, any plane transverse to the axis is a plane of maximum shear stress, so that crack initiation can occur on more than one plane. If the parallel planes of crack initiation are close enough together, they will coalesce during propagation with cracking on a ligament between the two planes, leading to the presence of a ratchet mark (Figures 40 and 41). Ratchet marks are common in cyclic loading, but also occur in monotonic loading.

As noted previously, metallic crystalline materials deform plastically through the grains by slip and twinning and fracture in a brittle way by cleavage. Additionally, depending on the temperature and the microstructure, they may also deform within, or immediately adjacent to, the grain boundaries.

In an idealized, single-phase alloy containing no inclusions, ductile fracture could occur by continued slip deformation on multiple planes to provide 100% reduction in area (Figure 42). Brittle fracture could occur by cleavage in all grains. Cleavage in this idealized case occurs on a single macroscale plane, but the fracture plane changes orientation on the
Ductile and Brittle Fracture

Figure 40. Schematic showing all possible ratchet marks that may be created in torsion loading so that cracking on multiple planes is likely. The maximum shear stress has the same magnitude as the maximum normal stress. See Lesson 2 for more details. If cracking occurs on two closely spaced parallel planes, the two cracks tend to coalesce as they grow toward the interior.

Figure 41. A ratchet mark forming in bending loading of a circular section. (a) Schematic view of a ratchet mark for a component loaded in bending. (b) Ratchet mark in a specimen loaded in bending. Source for (b): Reference 26.

microscale as the crack propagates across grain boundaries. Different fracture-surface morphologies are observed depending on the orientation relationship between two grains. Often, when a propagating cleavage crack crosses a grain boundary, there is usually nucleation on multiple planes in the new grain. These cracks subsequently coalesce as the crack propagates, creating a characteristic feature known as a river pattern (Figure 43). Coalescence of the multiple cracks “down river” indicates the crack-propagation direction. Microscale river patterns, like macroscale radial and chevron patterns, point back to the crack-initiation site.

Commercial polycrystalline alloys contain second phases and inclusions of varying shape and deformability as well as lamellar structures. These microstructural constituents provide additional mechanisms of crack initiation and propagation that are not present in single-phase alloys. Inclusions
typically do not have any, or very low, bond strength to the matrix, while there may be some bond strength between second phases and the matrix. Additionally, second phases in the microstructure are typically brittle and do not deform by slip. Inclusions may be either ductile or brittle depending on their inherent ductility and the loading conditions. Examples include ductile MnS in fabricated steel, brittle oxides, and nitrides in steel, and iron- and silicon-rich phases in aluminum alloys. A large hydrostatic stress may cause inclusions and or second phases to deform plastically rather than in a brittle manner.

If there is no bond between the matrix and the inclusion (typical), plastic deformation results in debonding and void formation at the inclusion-matrix interface (Figure 44), which creates crack nuclei. If there is bond strength as for second phases, these phases may cleave due to the normal stress transmitted through the matrix, may be cracked due to a shear stress, or may be cracked due to prior compressive strain (Figure 45). Additionally, second phases and inclusions that have a high aspect ratio typically crack at midlength, again creating crack nuclei. Examples include elongated MnS particles in steel, Fe₃C in pearlite, and grain-boundary cementite films in steel.

Within the fibrous zone, the plane of crack propagation may be either a plane of maximum normal stress or one of maximum shear stress depending on the volume fraction and spacing of the inclusions. If the inclusion spacing is small, cracking occurs by void nucleation, either at the inclusion-matrix interface or at half-length in elongated particles followed by void coalescence on the plane of maximum normal stress (pure ductile tearing, Figure 46). Alternatively, for wider particle spacing, strain localization occurs on shear (slip) planes (Figure 47). Void growth in three directions is enhanced by the presence of large triaxial tensile stresses, and it promotes ductile tearing. If the inclusion spacing is large, the voids increase in size by microscale necking in the intervoid ligaments by slip on planes of high shear stress. If the inclusion falls out of
the fracture surface, slip plane extrusions are sometimes visible on the wall of the matrix, and these extrusions should not be confused with fatigue striations. Whether the volume fraction of inclusions is large or small, this ductile fracture process of void nucleation, growth and coalescence is known as microvoid coalescence (MVC). This process is the most common mechanism for ductile fracture. If the volume fraction of second-phase particles is sufficiently low, crack initiation cannot occur by this mechanism. The required low volume fraction of inclusions is typically not obtained in commercial and especially polyphase materials. The fracture appearance at the microscale due to MVC is often described as being dimpled, hence the term dimple rupture (Figure 48).
Figure 44. Debonding at a MnS inclusion in heat treated SAE 4140. Courtesy of Michael West, University of Tennessee.

Figure 45. Cracked cementite particles in a low-carbon steel due to prior cold work. (a) Nital etch.

After a macroscale crack has been created in the specimen, constraint is lost, and deformation continues in a ductile manner by *plane-stress loading* in the remaining ligament. Figure 33 shows the microstructural appearance of the shear zone. Note that the elongated MnS particles, which were originally parallel to the axis of the specimen, have rotated and lie parallel to the fracture surface in the shear band. Also note that this region of intense shear strain is quite narrow. The presence of inclusions or second phases in the shear band can facilitate crack propagation by debonding at inclusion-matrix interfaces.

A dual dimple size is sometimes observed (Figure 48). Quenched-and-tempered steel microstructures typically contain large, weakly bonded
Figure 46. Pure ductile tearing on the plane of maximum normal stress. Source: Reference 29, p 280.

Figure 47. Schematic of ductile fracture by void coalescence. Source: Reference 29, p 270.

Inclusions in conjunction with smaller carbide particles. Age-hardening aluminum alloys may contain large second phases (e.g., CuAl₂) in conjunction with smaller particles rich in iron and silicon. For steels, debonding occurs first at the inclusion-matrix interface and later at a higher stress by debonding at second-phase particles. Qualitatively, larger particles or inclusions create larger dimples, so if fracture initiates at both inclusions and second-phase particles, a bimodal distribution of dimple sizes is observed.
Figure 48. Dimple rupture. Note the variation in dimple size associated with the variation in MnS size in the dimples. Source: Reference 2.

Recent work on an aluminum-lithium alloy suggests different behavior in this alloy system (Reference 30). Careful in situ fracturing in the SEM permitted a detailed study of the fracture process as microcracking occurred in front of the advancing crack tip. The alloy contained several types of second-phase particles, some iron- and silicon-rich that are present in all aluminum alloys, and other phases containing copper, magnesium, zirconium, and titanium. It was concluded that the iron- or silicon-rich phases that were very small played no part in the fracture process. The other phases, which were not cracked prior to loading the specimen, in some cases cracked during deformation (larger particles) and, in other cases, did not (smaller particles). However, the important point is that there was no debonding and void growth observed at the interface between the second phase and the matrix. Debonding at the inclusion-matrix interface did not initiate the fracture process. The primary crack moved through the matrix nominally normal to the applied load. If a sufficiently large second-phase particle in front of and on a closely spaced parallel plane as the advancing primary fractured, the primary crack was then deflected slightly and the crack within the particle became a part of the primary crack. The implication is that debonding may be the most common mechanism to create crack nuclei, but it may not be the only possible process.

In high-strength, limited-ductility alloys, ladle additions for grain-size refinement may result in the presence of nondeformable, faceted inclusions (e.g., nitrides). These inclusions can initiate cracking in the adjacent matrix due to stress concentration at the corners of the inclusions.

When quenched-and-tempered steels are tempered, the fracture surface is often composed of thin ridges containing dimples interspersed between regions containing cleavage facets. The macroscale fracture is brittle. Individual grains may or may not be delineated. River patterns are common and indicate multiple crack initiation in front of the macroscale
Ductile and Brittle Fracture

Dimple Morphology and Type of Loading

This fracture appearance has historically been described as quasi-cleavage. However, there is nothing “quasi” about the cleavage that occurred. Current preference is to describe the fracture process as “cleavage” (Figure 49).

High-magnification examination of the fracture surface reveals dimple shape. Dimple shape can eliminate some possible loading conditions and also indicate the direction of crack propagation. Figure 50 shows void formation for axial, bending, and shear loading. For axial loading, the dimples formed around the second phases are circular. For shear and bend loading, they are elongated and open on one end (parabolic shape). There is often little elongation of the dimples in combined axial and bending loading. Each half of the fracture surface should be examined if the dimples are not round in cross section to distinguish bending from shear loading.

Shear loading can occur for two orientations of the applied load (Figure 50), described as in-plane shear or edge sliding mode (mode II deformation) and out-of-plane shear, anti-plane shear, or screw sliding mode (mode III deformation, not shown in Figure 50). Mode III deformation is the deformation mode associated with torsional loading. For Mode III deformation, the dimples are rotated with respect to the direction of crack propagation. In some ways, the use of the term “mode” to describe the type of loading is inconvenient as term is also used to describe the microscale mechanism (slip versus cleavage and/or transgranular versus intergranular fracture) responsible for fracture. The difficulty can be minimized if the term “loading mode” is used to describe loading conditions and use of the term “mode” restricted to microscopic scale phenomena.

There are other microscale fractographic features in addition to dimples, cleavage, and river patterns discussed above. Some of these features are associated with cyclic loading (for example, fatigue striations). Care must
Figure 50. Influence of loading conditions on dimple shape in ductile fracture. (a) Equiaxed dimples in tension loading. (b) Open-ended dimples in in-plane shear loading (Mode II). Dimples point in opposite direction on opposite sides of the fracture surface. (c) Open-ended dimples in opening mode (mode I). If dimples are elongated, they point in the same direction on each side of the fracture surface. Source: Reference 28.

be taken to not confuse similarly appearing fractographic features caused by different mechanisms. An untrained observer might confuse the lamellar morphology of pearlite, if it is revealed on a steel fracture surface, with fatigue striations. Another somewhat similar-appearing structure is the presence of slip bands extruded on a smooth surface such as the interface between an inclusion and the matrix or at porosity. Striations and striationlike features are discussed in Lesson 4.

Some care must be taken also to not confuse a cracked surface deposit (mud cracks) with the metal fracture surface. If an electron-dispersive analysis by x-ray (EDAX) attachment is available on the SEM, chemical analysis will quickly reveal the presence of a nonmetallic surface deposit and dried body fluids (Figure 51).

It may be possible to clarify features on the fracture surface by etching. This procedure will delineate a lamellar structure (such as pearlite) or needles of a second phase and distinguish it from structure fatigue striations. Once the fracture surface has been etched, the material is no longer in its as-fractured condition. As a result, some caution must be observed before this is done, and if litigation is involved, judicially approved permission of the proposed procedures is probably required.
Additionally, there is sometimes a lack of contrast between two phases on the fracture surface; so if both dimples and cleavage are observed, it may not be possible to associate cleavage with the second-phase particles and dimples with the matrix. Again, etching the fracture surface can help identification. Alternatively, metallographic sections may be taken through the fracture surface.

A common crack-initiation site in cast materials is at locations of shrinkage porosity (Figure 52). Shrinkage porosity is easy to recognize because of the characteristic appearance of the revealed *dendrites* on the walls of the porosity.

![Figure 51. Mud cracks on the fracture surface of a mild steel weldment tested in synthetic seawater. Source: Reference 32.](image)

![Figure 52. Crack initiation at shrinkage porosity in aluminum casting alloy A356 specimen loaded in bending. Crack initiation at the top of the figure as indicated by a faint ridge pattern in the dendritic region. Source: Reference 33.](image)
The group of features used to identify the location of fracture initiation, crack-propagation direction, and loading conditions are:

- Ridge patterns (including radial lines) and chevrons
- Shear lips
- Ratchet marks
- Surface roughness and light reflectivity
- Crack-arrest lines (more accurately, a change to dimpled rupture from cleavage or quasi-cleavage)
- Rubbing and or smeared metal

All of these markings may not be present simultaneously depending on section size, loading conditions, loading rate, test temperature, and relative motion of the fracture surfaces during crack propagation. If ridge patterns or chevrons are present, they point back to the crack-initiation site. In their absence, cracks propagate parallel to shear lips on a multiplane fracture, and, if ratchet marks are present, the crack-initiation site(s) is at the location of the ratchet mark(s). Surface roughness increases as the crack propagates, and the roughest region on the surface is the last to fail. The rougher the fracture surface, the more likely the microscale fracture mechanism is ductile, even though the macroscale surface is “flat.”

Brittle fracture in a coarse-grained material causes light reflection from cleavage facets within individual grains, and the fracture surface appears bright and shiny. Microscale ductile fractures have a matte, silky dull, or dark gray appearance and are not highly reflective.

Consider a tensile specimen subjected to a rising load and containing a cracklike defect on a plane perpendicular to the load (Figure 13). Crack initiation will occur at some critical load at or adjacent to the imperfection. However, there may be insufficient stored elastic strain energy to drive the brittle crack completely through the cross section, so the crack arrests at some distance into the material. Then the load begins to rise again causing some (microscale) ductile crack propagation, but the increased elastic strain energy again causes the crack-propagation mechanism to change to microscale cleavage. This can occur a number of times as the crack propagates across the specimen. Each time the crack arrests and the microscale fracture mode changes to ductile fracture, a distinctive small band is left on the specimen (Figure 53). These bands are described as crack-arrest lines. Note that they have curvature, but the lines point in the direction of crack propagation, rather than away from the initiation site as do radial marks and chevrons. Note too, that the lines do not close along the centerline in the two thickest specimens in the figure for which there is more constraint. Similar features are often observed on Charpy impact specimens. Crack-arrest lines may also be visible on intermittently loaded fatigue fracture surfaces (Lesson 4), where they are identified as beach marks caused by local oxidation or a change in loading spectrum.

Macroscale visible smeared metal can sometimes be examined to identify the final direction of separation. Rub marks at the macroscale
can polish a surface resulting in high reflectivity. They are commonly associated with fatigue cracking and in which case the fracture surfaces are in intimate contact. At the microscale, rubbing may result in fine, random-orientation scratches especially if vibration is present, and it may often obliterate a crack-initiation site. Alternatively, final separation may result in unidirectional rubbing and gouging of the fracture surface, providing some information about loading conditions. Ductile torsional fracture often results in a rubbed “swirl pattern” visible on the fracture surface (Figure 54).

**Figure 53.** (a) Crack-arrest lines formed on the fracture surface (dark curved lines that look like parabolas pointing “up” in the figure) in an O1 tool steel. Shear lips (dark) are also visible on the sides of the specimen. Note that the first arrest line occurs at a further distance from the initiating notch as the specimen increases in thickness. Also note that at intermediate (and larger) thickness, the arrest lines are not closed along the centerline of the specimen, indicating more constraint in this region. Thin dark bands and small shear lips on side of specimen show MVC fracture. Fracture in the higher reflectivity regions is predominantly by cleavage. Source: Reference 31. (b) Load-extension curve for the material shown in (a).

**Figure 54.** Fracture surface of a torsion-loaded steering assembly component that failed in a ductile manner on a plane of maximum shear stress. Note the swirling rubbed pattern on the fracture surface.
Manufactured components may contain manufacturing imperfections that can be either geometric or metallurgical. Microscale cracking will always initiate where the local stress exceeds the local strength of the material.

The distribution of macroscale stresses across a cross section is discussed in Lesson 2. Axial tension until the onset of necking produces a uniform local stress across the cross section, while in bending and torsion the nominal stress is maximum at the surface. Therefore, assuming for the moment that the strength is constant across the cross section, crack-initiation sites in axial loading are equally likely at any place on the cross section, but are at the surface in bending and torsion. However, manufacturing operations often create both macroscale and microscale stress concentrators or, in some instances, cause microscale cracking. Manufacturing imperfections typically occur at the centerline and at the surface (discounting surface-treated components).

Note that crack initiation cannot occur at the centerline in bending or torsional loading. Therefore, the component shown in Figure 55 must have been loaded axially. The radial pattern diverges from what appears to be a near-centerline imperfection. There is essentially no stress at this location in bending or torsion.

Surface treating (induction hardening, carburizing, nitriding, ion implantation) is a common practice to increase both the load-bearing capacity and the wear resistance of parts, especially for cyclic loading conditions. Lesson 4 contains additional details. The issue is raised here to show how a varying microstructure and a varying nominal stress can interact to cause subsurface crack initiation in a location where the nominal stress is not maximum.

Assume that a component has been surface treated to produce a high-strength and wear-resistant surface and that the component is to be loaded in bending or torsion. With reference to Figure 56, the applied nominal bending or torsional stress varies linearly from the surface as shown. Two

**Figure 55.** Crack initiation in an axially loaded AISI 4340 tensile specimen (ultimate tensile strength = 1945 MPa, or 282 ksi). Radial marks point to a small defect near the center of the specimen. Source: Reference 34.
Different material **strength gradients** are shown, one for a “thick” case and one for a “thin” case. Fracture initiation is subsurface for the “thin” case. Subsurface fracture can also initiate if the hardness change from case to **core** is too steep and the case is not very deep. This illustrates the fact that specifications for the surface treatment of parts should be written with significant forethought. It is inadequate to just specify a surface hardness. A second subsurface hardness is necessary to preclude subsurface failure.

**Microscale Crack-Initiation Sites**

In addition to microvoid formation at inclusions and second phases and cracking in second phases or inclusions, a number of other crack-initiation sites have been observed in commercial alloys. However, not all of these initiation sites are to be considered “imperfections” or defects. Cracking always initiates where the local stress exceeds the local strength. The following is a list of potential crack-initiation sites (with some locations or mechanisms listed more than once).

**Transgranular:**

- Intersection of mechanical twins
- Tempered martensite embrittlement—fracture in ferrite between the carbides located in the martensite plate boundaries
- Fracture in cementite plates within a pearlite colony normal to the cementite plates
Crack-Initiation in Surface-Treated Components

- Fracture at pearlite colony boundaries
- Fracture at bainite colony boundaries
- Fracture within ferrite in upper bainite

**Grain-boundary related:**

- Grain-boundary triple points—may be ductile by shear, ductile normal separation, brittle normal separation
- Cleavage crack initiation at a grain boundary
- Void (R) and wedge (W) crack nucleation and propagation at elevated temperature
- Intersection of a mechanical twin and a grain boundary
- Temper embrittlement at prior austenite grain boundaries in steels
- Adjacent to a grain boundary due to solute denudation (austenitic stainless steels, precipitation-hardenable aluminum alloys, and others)
- At a grain boundary due to partial liquation (*incipient melting*)

**Second-phase or inclusion related:**

- Debonding at inclusions and second phases (the dominant ductile fracture mechanism in commercial-purity metallic materials)
- Matrix crack initiation and propagation caused by stress concentration at a faceted second phase
- Matrix crack propagation from a cracked second phase (cracked second phase may have been created by prior forming operations)

**Fabrication related:**

- Matrix propagation from a cracked second phase
- Cracking due to cold work
- Centerline cracking from prior cold work (e.g., a forging burst)
- Shrinkage porosity not closed by working
- At the intersection of a shear band and the surface of the material
- At the intersection of shear bands
- At laps and seams created in forming (forging, wire drawing, etc.)
- At machining marks
- At locations where poor grain flow exits at the surface of a forging

**Welding imperfections:**

- Incomplete penetration
- Incorrect weld metal profile (reentrant angle)
- In weld metal due to *hot tearing*
In the heat-affected zone of a steel due to the presence of untempered martensite in large prior austenitic grains

**Casting imperfections:**

- Shrinkage porosity
- At concentrations of casting sludge
- At internal oxide surfaces due to casting splash

**Environmental:**

- At surface corrosion pits
- Due to hydrogen absorption in grain boundaries (IG)
- Due to aqueous corrosion (TG and IG)
- Due to liquid metal embrittlement along grain boundaries (IG)
- Due to oxygen diffusion and the formation of oxide particles in grain boundaries (IG)

At elevated temperatures where *creep* (time-dependent deformation at constant load) occurs, experimental work indicates that significant viscous flow occurs in or adjacent to the grain boundary in addition to transgranular deformation. As the temperature is increased from room temperature, ductility initially increases, but then it goes through a minimum before again increasing (Figure 57).

Metallographic examination shows that before the maximum in fracture strain there is typically some grain growth in the material and that the minimum in Figure 57 is associated with intergranular (IG) fracture rather than transgranular (TG) fracture in large-grained material. Location of the minimum depends on the penultimate grain size and loading rate. Above the minimum, dynamic recrystallization may occur, in which case fracture is then transgranular in recrystallized material.

Two types of intergranular fracture are recognized: wedge (W) cracking and round (R) cracking (void formation) as shown in Figures 58 and 59. Wedge cracking (Figure 59a) occurs at lower temperatures and higher stresses than does R cracking (Figure 59b). R cracking results from condensation of vacancies on grain boundaries, which lowers the grain-boundary area, increasing the stress on the boundary. Normal separation along the grain boundary may occur by a ductile process, resulting in the formation of small dimples on the boundary (sometimes described as grain-boundary *decohesion*) (Figure 60), but this does not result in significant macroscale plastic deformation. In the temperature range in which R cracks form, there may also be considerable porosity formed in grain interiors. The result is that considerable porosity may be visible in metallographic specimens at and adjacent to the fracture surface (Figure 61). This type of fracture is often termed *cavitation* fracture.
Figure 57. Variation in ductility with temperature. Fracture below the minimum is transgranular. Fracture in the minimum is intergranular. Fracture above the minimum is transgranular in recrystallized material. Source: Reference 21, p 26.

Figure 58. (a) “W” and (b) “R” cracking. Source: Reference 28.

The fundamental causes for intergranular as opposed to transgranular cracking are fairly limited in number, some of which are associated with improper heat treatment and may be alloy specific. Because they are limited, the presence of IG fracture can often be used to help identify the cause for failure. Additional causes for IG fracture are discussed later in this lesson.
Figure 59. Cracking at elevated temperature. (a) Wedge (“W”) cracking in nickel at a grain boundary triple point. Source: Reference 35, p 101. (b) “R” cracking. Tensile axis is vertical. Note higher incidence of voids on grain boundaries perpendicular to stress. Source: Reference 36, p 266a.

Figure 60. Intergranular fracture by microvoid coalescence. Source: Reference 28.

Exposure to elevated temperature implies changes in microstructure with time in service. Plain-carbon steels may graphitize at elevated temperature. Cementite present in the microstructure transforms to graphite, resulting in much lower strength and ductility with long-time service. In an oxidizing atmosphere, decarburization occurs at the surface of the material. If decarburization is extensive, oxide formation also results. Not only is there an oxide coating on the surface, but oxides may be present in the grain boundaries. The latter can result in intergranular crack propaga-
Imperfections Associated with Heat Treatment


tion. Oxide on the surface of a plain-carbon steel does not form an adherent protective barrier so that new material is continually exposed to the atmosphere. In situ graphitization (decomposition of cementite to graphite and ferrite) and scaling resistance are both improved by the addition of chromium to steels. In a carburizing atmosphere, hypereutectoid cementite may be formed in the grain boundaries at elevated temperature when the solubility given by the $A_{cm}$ is exceeded, and/or during cooling to temperatures below the $A_{cm}$ ($A_{cm}$ is the temperature at which cementite completes solution in austenite). Again, intergranular fracture may result. Also, the high-carbon steel has little inherent ductility.

Nonferrous materials are also subject to several deleterious changes in the microstructure when subjected to elevated-temperature service. A few specific examples include:

- Excessive grain growth at elevated temperature
- Partial solution of inclusions at elevated temperature that reprecipitate during cooling, often in the grain boundaries
- Partial liquation of eutectic and divorced-eutectic constituents located in the grain boundaries
- Creation of coarse, grain-boundary precipitates and a denuded zone adjacent to the grain boundary
- Materials containing oxides such as tough-pitch copper that become unstable in reducing conditions (the oxides are reduced to form pockets of steam that then create large hydrostatic stresses; copper oxides are not present in phosphorus-deoxidized copper, so this material is better suited for service in a reducing environment than tough pitch copper)

A material may lose strength and toughness due to improper thermal processing, including both time at a temperature and cooling through a temperature range, or may lose strength and toughness in service at temperature. There are a large number of embrittling phenomena that can occur, some of which are alloy specific. Some examples are given to illustrate the
importance of understanding the role of the microstructure in controlling behavior. A more extensive discussion of embrittlement in steels due to heat treatment can be found in References 17, 38, and 39.

Some austenitic stainless steels after normal commercial processing do not have thermodynamically stable microstructures. These alloys (e.g., 302, 304, 316) contain some low residual carbon in solution that exceeds the equilibrium solubility unless the carbon has been deliberately removed (321 and 347 grades). If these alloys are heated into the temperature range of 675 to 900 °C (1250 to 1650 °F) or slowly cooled through that temperature range, chrome-rich carbides precipitate at the grain boundaries, resulting in embrittlement. The material is then said to be sensitized. “L” grades of stainless (containing a specified maximum carbon content) are available but still become embrittled, although the required time is increased and the degree of embrittlement decreases. Carbide precipitation is accompanied by a chromium-denuded zone adjacent to the grain boundary. When loaded to failure at room temperature, the fracture path changes from transgranular in unsensitized material to intergranular in the denuded zone, accompanied by a reduction in fracture strain, toughness, and corrosion resistance. The denuded zone adjacent to the grain boundary is anodic relative to the grain interior. Standardized procedures are available to test for this condition. Chemical tests designed to test for sensitivity are described in ASTM A 708 (Reference 40). Similar embrittlement can occur in other fcc alloys containing some chromium and either nitrogen or carbon. If carbides or nitrides precipitate due to elevated-temperature service, the denuded zone is not always apparent with metallographic examination, and, in fact, it may not be present if time at temperature was long enough to homogenize the microstructure. Fracture can still be intergranular, but corrosion resistance is improved (Reference 41).

Ferritic and duplex stainless steels are also sensitive to intergranular corrosion attack, but the conditions causing embrittlement are different than for the austenitic grades. Sensitization occurs after air cooling or quenching from 925 to 1100 °C (1700 to 2010 °F). Slow cooling from these temperatures does not result in embrittlement. Again, the mechanism responsible for embrittlement is the precipitation of chrome-rich carbides and for these materials, also nitrides. Lower solubility of carbon and nitrogen in the bcc lattice results in precipitation even with water quenching from high temperatures. The higher diffusion coefficients of nitrogen and carbon permit removal of composition gradients in the matrix by reheating the material in the temperature range 425 to 925 °C (800 to 1700 °F) (and slow cooling permits homogenization during cooling). Required times to remove embrittlement after quenching are of the order of minutes at 925 °C (1700 °F) and greater at lower temperatures.

A type of embrittlement similar to that in austenitic stainless occurs in age-hardening aluminum alloy systems and potentially in all age-hardening systems. Precipitation of a second phase rich in the solute added for age hardening occurs in the grain boundaries; for example, CuAl2 in 2xxx aluminum alloys. The solution anneal prior to aging is supposed to take this
phase into solid solution. If cooling from the solution-annealing temperature is not sufficiently rapid and especially if there is some residual second phase remaining in the grain boundary, solute adjacent to the grain boundary diffuses to the preexisting phase in the boundary, creating a denuded zone.

Figure 62 shows denudation in an age-hardening Al-Zn-Mg alloy. Aging response, fracture toughness, and corrosion resistance are degraded. The situation can be made worse by selection of an improper aging temperature. As for the stainless steels, phase-composition differences create a corrosion couple. However, in the case of aluminum alloys, the couple may be between the grain interior and the precipitated second phases rather than the grain interior and the denuded grain-boundary area.

Partial liquation (also called incipient melting and eutectic melting) can occur during the solution anneal if the temperature and time are not carefully selected. If the divorced eutectic does not go into solid solution before the material reaches the eutectic temperature, the eutectic reaction reverses, causing a pool of liquid of larger volume than the previous solid in the grain boundaries. When this material resolidifies, local shrinkage occurs, leaving porosity in the grain boundaries (Figure 63). Fortunately, this situation is easy to recognize by metallographic examination.

Incipient melting is also an imperfection that occurs due to overheating during hot forging in both ferrous and nonferrous alloys.

Several types of embrittlement can occur in steel depending on composition and heat treatment. They are identified by terms associated with heat treating practice as well as by the temperature ranges in which they occur. Detailed discussions are available in References 17, 38, and 39. In some instances, embrittlement results in IG fracture. Correct thermal processing of these same steels does not ordinarily result in IG fracture.
As indicated in the introduction to this lesson, there are a relatively small number of fundamentally different reasons for brittle fracture to occur in a nominally ductile material. If fractographic examination indicates IG fracture, the analyst should consider the possibility of an adverse operating environment, or an improper heat treatment, especially if the component geometry indicates that geometric constraint should be small.

Embrittlement due to heat treatment can occur in most plain-carbon and alloy steels, from lowest carbon content through high-carbon grades. It can occur in pearlitic, tempered martensitic, and bainitic microstructures.

Low-carbon galvanized sheeting is typically about 3.2 mm (1/8 in.) thick. This is an inadequate thickness to develop plane-strain loading conditions at room temperature. If brittle TG fracture or IG fracture is observed, the cause for failure is not due to geometric constraint associated with gouges, nicks, and so forth unless the material was in service below the DBTT temperature. In one failure of a structure fabricated from this material, brittle fracture was believed to be associated with a high nitrogen content (Reference 43).

<table>
<thead>
<tr>
<th>Type of Embrittlement</th>
<th>Fracture Mode</th>
</tr>
</thead>
<tbody>
<tr>
<td>Strain-age embrittlement</td>
<td>TG</td>
</tr>
<tr>
<td>Quench-age embrittlement</td>
<td>TG</td>
</tr>
<tr>
<td>Blue brittleness</td>
<td>TG</td>
</tr>
<tr>
<td>Strain-relief embrittlement</td>
<td>IG</td>
</tr>
<tr>
<td>Aluminum nitride embrittlement</td>
<td>IG</td>
</tr>
<tr>
<td>Overheating during a solution anneal or forging</td>
<td>IG</td>
</tr>
<tr>
<td>Temper embrittlement (TE)</td>
<td>IG and/or TG</td>
</tr>
<tr>
<td>Tempered martensite embrittlement (TME)</td>
<td>TG</td>
</tr>
<tr>
<td>475 °C embrittlement</td>
<td>TG</td>
</tr>
<tr>
<td>Sigma-phase embrittlement</td>
<td>TG</td>
</tr>
<tr>
<td>Graphitization</td>
<td>TG and IG</td>
</tr>
<tr>
<td>Carburization</td>
<td>IG</td>
</tr>
<tr>
<td>Intermetallic compound embrittlement(a)</td>
<td>IG</td>
</tr>
<tr>
<td>Excessive decarburization (grain-boundary oxides)</td>
<td>IG</td>
</tr>
</tbody>
</table>

(a) May be a form of liquid metal embrittlement and discussed in that section.

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Figure 63. Eutectic melting due to overheating in 7075-T6 (a) and 354-T4 (b) aluminum. Note the porosity on the grain boundaries and the IG fracture.
Low-carbon steels are subject to strain-age embrittlement as well as quench-age embrittlement and blue brittleness. Medium-carbon alloy steels that are used in the quenched-and-tempered condition are subject to both temper embrittlement (also called one-step temper embrittlement) and tempered martensite embrittlement (two-step temper embrittlement). Alloy steels are subject to still other types of embrittlement, such as 475 °C embrittlement in steels containing more than 13% Cr. Tool steels may be subject to “500–700 °C embrittlement.” In many, but not all of these embrittlement types, embrittlement is associated with the location and morphology of carbides precipitated in the microstructure. Strain-age embrittlement (SAE) is associated with both nitrogen and carbon, but nitrogen is assumed to play a more important role. Embrittlement may also be caused by precipitation or partial solution and subsequent reprecipitation of inclusions during annealing. Examples include MnS in steels austenitized at a high temperature and thermal embrittlement of maraging steels due to precipitation of TiC and/or Ti(C,N) as a result of being cooled too slowly after annealing. Embrittlement may also be caused by aluminum nitride in aluminum-killed steels if the aluminum content is too high.

Plain-carbon steels containing ferrite and pearlite (or divorced pearlite) are not normally considered to be age-hardening systems because the dominant strengthening mechanisms are assumed to be associated with the volume fraction of pearlite, grain size, and cold work. However, there is an aging effect due to precipitation of TiC and/or Ti(C,N) carbide and/or nitride precipitation that leads to embrittlement. Multiple terms are used to describe the embrittlement depending on the mechanical and thermal processing used.

Strain-age embrittlement occurs by aging of cold-worked, low-carbon steels, especially rimming grades and capped steels, at room temperature, or slightly higher (as in a hot warehouse during the summer). Two examples where SAE has been identified include punched rivet holes in sheet steel and in boilers (Reference 44). Cold working of these steels is often done to suppress the yield point, and, therefore, these materials often have a small amount of prior cold work. However, when they are heated to temperatures at or a little above room temperature and held for some period of time, the yield point returns due to nitride and possibly some carbide precipitation. The result is a loss of ductility concurrent with an increase in yield strength. In one example (Reference 44), the hardness changed from about 70 HRB in the cold-worked material to about 84 HRB in the embrittled material after 200 days at 40 °C (105 °F). The time necessary to cause embrittlement decreases as the temperature is increased, and only a few minutes are required at temperatures near 205 °C (400 °F). Because the embrittlement is associated with the return of sharp yield behavior and a loss in ductility, this type of embrittlement can be identified by tensile testing of a specimen removed from the component. An alternate and more common method, if the section thickness permits, is to anneal material that has been cold worked about 5% at 260 °C (500 °F). Then Charpy V-notch specimens are prepared and tested to determine if there has been a shift in the DBTT. An increase of the DBTT indicates embrittlement as a result of SAE. Strain-age-embrittled materials fail by cleavage.
A similar type of embrittlement, blue brittleness, occurs in plain-carbon steels and some alloy steels that have been heated between 230 and 370 °C (446 and 698 °F). Again, the embrittled material displays an increase in strength and a decrease in ductility and toughness. The term arises from the color of the oxide scale produced on steels heated into this temperature range. Embrittlement is again associated with carbide precipitation, which can be accelerated by prior cold work. Because yield strength increases with strain rate, blue brittleness can be pushed to higher temperatures with increased loading rates. Fracture is transgranular.

Quench-age embrittlement occurs in low-carbon steels that are cooled rapidly from a temperature just below the lower critical temperature ($A_{cl}$) and then aged at temperatures near room temperature. There is once again an increase in strength and hardness with a concurrent loss in ductility and toughness. As with strain-age embrittlement, quench-age embrittlement is a diffusion-controlled process, so that shorter times are required as the temperature is increased. Additionally, if the temperature at the start of rapid cooling is decreased further below the $A_{cl}$, there is less supersaturation of carbon and a concurrent decrease in the degree of embrittlement. Quenching from temperatures below 560 °C (1040 °F) does not result in embrittlement, and carbon contents greater than 0.12% are not susceptible. Like strain-age embrittlement, quench-age embrittlement occurs by transgranular fracture. Presumably, the fracture will change from MVC to mixed MVC and cleavage and then to essentially total cleavage as the degree of embrittlement increases. Examples and numeric data are available in Reference 44.

Two types of embrittlement should be recognized in welded assemblies. One type, noticed most in low-carbon steels, is associated with precipitation of carbides in the heat-affected zone of the weld. A second type is associated with higher carbon and solute content and is associated with the formation of higher-carbon martensite in large-grained prior austenite in the heat-affected zone.

For lower-carbon-content materials, embrittlement may occur during a postweld stress-relief anneal (stress-relief embrittlement). In this case, some limited precipitation occurs both in the grains and at the grain boundaries during initial cooling. The terms heat treat cracking or reheat cracking are also used to describe this behavior.

Embrittlement due to martensite formation can occur in both plain-carbon and alloy steels in the heat-affected zone, depending on the carbon and solute contents. The heat-affected zone is material heated above the upper critical temperature and is located adjacent to the molten weld metal. As a result, there may be a large amount of austenite grain growth, which increases hardenability. In the lower-carbon contents, martensite formed during cooling can self temper and embrittlement is minimized. However, above about 0.3% C, the combination of fresh martensite and a large prior austenite grain size results in low toughness. Depending on constraint present at the weld, quench cracking may result, and fracture may be intergranular.
The contemporary HSLA steels of which one group has a carbon content of approximately 0.12% and a second group at 0.18 to 0.20% C were developed specifically to provide weldable-grade steels having yield strengths greater than 345 MPa (50 ksi). The low carbon content results in a steel free of embrittlement.

Quench-and-temper operations for heat treatable steels are common operations designed to impart maximum toughness sometimes in conjunction with a high yield strength. Yet, if times and temperatures for the tempering operation are not selected properly and if certain trace alloying elements are present, the resulting fracture toughness may be low. Unfortunately, although it is common to use hardness measurements as a quality-control check of the heat treatment, hardness changes with tempering typically do not reveal the embrittlement.

Embrittlement during tempering operations has been recognized for many years, but debate concerning the fundamental causes of embrittlement has changed as new research tools have become available. One advancement occurred with the development of electron microscopy, which made it possible to examine carbide morphology and location, and a second advancement occurred with the development of Auger spectroscopy, which made it possible to obtain chemical analysis from near-surface material on a fresh fracture surface.

Embrittlement today is felt to be due to three different mechanisms: (1) an improper carbide morphology and distribution resulting from the tempering of martensite, (2) the decomposition of retained austenite during tempering, and (3) the presence of some specific impurity elements and the associated segregation of solute elements during tempering. All three factors may be involved in tempered-martensite embrittlement, but it is primarily the third factor that is involved in temper embrittlement. Associated with these mechanisms in a practical heat treating operation is the possibility of decarburization and oxide formation in the grain boundaries at the surface of the material.

**Tempered martensite embrittlement** (TME) occurs in alloy steels in the tempering temperature range 200 to 370 °C (400 to 700 °F) or during slow cooling through that temperature range. It is therefore associated with tempering operations that produce relatively high strength and hardness. One factor involved is the decomposition of martensite to form a carbide and/or the decomposition of thin bands of retained austenite to form either bainite or tempered martensite. A second factor involved and identified by Auger spectroscopy is the segregation of minor impurities to austenite grain boundaries. Elements that have been identified include tin, antimony, arsenic, and phosphorus. Tempered martensite embrittlement is still considered to be incompletely understood today, in part because more than one mechanism may be involved.

Careful examination of fracture surfaces indicates that fracture can be transgranular, intergranular, or mixed. Some workers report that tempering in the lower part of the embrittling range leads to transgranular fracture, while tempering in the higher part of the embrittling range leads to
intergranular fracture. Transgranular fracture is associated with carbide morphology and location, while intergranular fracture is associated with trace-element segregation.

In the absence of the complicating factors discussed below, TME is associated with transgranular fracture in ferrite or low-carbon martensite along prior martensite plate or bainite plate boundaries. During the tempering operation, martensite initially decomposes to a slightly tetragonal matrix phase and a transition carbide \( \varepsilon = \text{Fe}_{2.4}\text{C} \). The carbide is distributed over the prior martensite plate as well as at the plate boundaries. At a slightly higher tempering temperature, the \( \varepsilon \) carbide and tetragonal ferrite disappear and the equilibrium carbide (\( \text{Fe}_3\text{C} \), cementite) and ferrite appear. A similar morphology can be produced during transformation of retained austenite to bainite, and the same morphology may also occur in microstructures isothermally transformed to bainite. So-called upper bainite is a mixture of ferrite and a rodlike carbide that forms principally at high-angle bainite plate boundaries. If the temperature of the bainite formation is in the lower range of bainite formation, lower bainite is formed and the carbide is again metastable, but it is precipitated more uniformly over the bainite plates. Attempts to improve the toughness of the upper bainitic microstructure are only partially successful. Tempering at high temperatures does improve the upper-shelf toughness, but the DBTT is shifted upward (Reference 50). Fracture occurs within the ferrite in the tempered martensitic or bainitic microstructure and also at the bainite packet boundaries.

An additional factor may be involved in some steels. Retained austenite can be present as thin sheets between the martensite plates after quenching. This retained austenite has a high carbon content and, when tempered, forms thin carbide plates in the prior thin bands of retained austenite (Reference 38).

**Temper embrittlement (TE)** occurs when martensitic microstructures of alloyed steel are tempered in the range 375 to 575 °C (700 to 1070 °F) or slow cooled through this temperature range after tempering at higher temperatures. The embrittlement is diffusion controlled and therefore time dependent. It is associated with the concentration of certain trace alloying elements, notably arsenic, antimony, tin, and especially phosphorus. Because it results from tempering at higher temperatures and for longer times, it is associated with lower strength and hardness. It is a reversible process, in that toughness can be restored by reheating the steel to above the embrittlement range (but below the \( A_{c1} \)) followed by cooling rapidly through the embrittling range.

It is known that segregation of the embrittling trace-level elements is not sufficient to cause embrittlement. The presence of some other alloying element such as chromium and manganese is required. Molybdenum, tungsten, and zirconium greatly reduce embrittlement, and nickel, titanium, and vanadium reduce it slightly (Reference 28). Manganese is routinely added to steel to improve hardenability and to prevent hot-short behavior. Recent alloy development work has suggested that temper embrittlement can be minimized if the manganese content is reduced from above 0.5% to about 0.3%. This then prevents or minimizes embrittlement associated with slow
cooling of heavy sections. It has been known for many years that molybdenum is beneficial in reducing TE, but its mitigating effect is related to how much is in solid solution versus how much is tied up as carbides. Molybdenum tied up as carbides does not suppress TE.

It is again emphasized that, although it is common to use hardness measurements to determine acceptable heat treating, hardness measurements do not detect TE or TME (or most other types of embrittlement). Additionally, standard tensile tests using smooth cross sections of TE embrittled materials do not indicate the embrittlement, except perhaps for a decreased reduction in area for extreme cases. A standard procedure used to detect TE is to measure the DBTT using Charpy impact specimens. Embrittled specimens show an increase of as much as 200 to 300 °F in the 10 ft · lbf transition temperature. Examination of the fracture surface at high magnification in the SEM will show whether the fracture is intergranular or transgranular.

Toughness problems of heat treated steels are sometimes associated with a large, prior-austenitic grain size, and intergranular quench cracking is facilitated by a large austenite grain size. Additionally, the yield strength of the steel is reduced because of its square-root dependence on grain size. A large austenite grain size may be created because of high annealing temperatures used in an attempt to homogenize the steel or overheating during hot forming or forging. High homogenization temperatures may also cause other difficulties. It is common to add manganese to steel to tie up the sulfur as MnS and therefore to prevent hot-short behavior. However, if the temperature is too high during annealing or hot working (e.g., temperatures of 1040 to 1095 °C, or 1900 to 2000 °F), MnS partially dissolves and subsequently reprecipitates on austenite grain boundaries during cooling. The result is low fracture toughness and intergranular fracture. Reprecipitation of the MnS can, if severe, be identified by its morphology (spherical rather than elongated) and its intergranular location.

The 10 ft · lbf transition temperature of hot-rolled mild steel measured in a Charpy test can be shifted more than 100 °F due to changes in prior grain size. Hence, it is desirable from a toughness standpoint to keep the prior-austenitic grain size as small as possible. The traditional means of preventing austenite grain growth has been to use a grain-size refiner. Aluminum has been a common ladle addition for this purpose. However, too much aluminum can result in excess aluminum nitrides or oxides in the grain boundaries and IG fracture.

Service environments include both high and low temperatures as well as the possible presence of a chemically reactive atmosphere. The change in toughness associated with materials having a bcc lattice (and a few materials having a hcp lattice) at low temperature is discussed earlier in this lesson; this section considers high-temperature behavior.

Plain-carbon steels show little change in properties with an increase in service temperature until approximately 400 °C (750 °F). Above this temperature, several changes in behavior begin to occur: (1) there is a loss in elastic modulus, (2) decarburization of steels results in scale formation
and possible oxides in the grain boundaries, (3) fine pearlite begins to spheroidize, (4) cementite in pearlite may decompose to graphite, and (5) time-dependent deformation (creep) begins to contribute to the total strain as does strain resulting from thermal expansion.

Plain-carbon steels heated for long-time service at temperatures above 425 °C (800 °F) graphitize. That is, cementite in the pearlite decomposes to graphite with a corresponding decrease in elevated-temperature strength and ductility. Consequently, it is necessary to stabilize the carbide in some manner if the steel is to be exposed to long-time, elevated-temperature service. Such is the case in the chemical process industries (piping, reaction vessels, heat exchangers). Graphitization is enhanced in aluminum-killed steels containing more than about 0.025% Al. However, silicon-deoxidized steels can also be prone to graphitization. The standard procedure to minimize graphitization is to stabilize the carbides by the use of molybdenum and/or chromium additions. Chromium additions provide an additional benefit. Plain-carbon steels have poor scaling resistance, and when chromium is added to the steel the surface oxide becomes adherent, providing protection to the underlying metal.

### Hydrogen Degradation

The following material is excerpted in part from References 17 and 45.

Hydrogen pickup in steels can lead to severe embrittlement, but other alloys may also be susceptible. There are multiple sources of hydrogen pickup that include:

- Dissolved hydrogen precipitation upon solidification
- Acid cleaning (pickling) baths
- Electroplating operations
- Water or water vapor, or other hydrogen-containing liquids or including methane and hydrogen sulfide
- Pure hydrogen gas

The presence of hydrogen can result in:

- Hydrogen embrittlement
- Hydrogen stress cracking
- Hydrogen environmental embrittlement
- Partial loss of tensile ductility
- Blistering
- Shatter cracks, flakes and fisheyes
- Metal hydride formation

Hydrogen stress cracking is the brittle fracture of a material after hydrogen absorption. Hydrogen diffuses into the material and collects at regions of high stress concentration. Because some time is required for diffusion, there is typically a delay time before fracture occurs under load,
leading to the (poor) term static fatigue fracture. Fracture occurs at a
stress below the nominal yield strength, and susceptibility to this type of
fracture increases as the strength level of the steel increases. Some persons
consider a hardness of 30 HRC or a yield strength in excess of 1100 MPa
(160 ksi) to be the threshold values for the occurrence of hydrogen embrit-
tlement. Lower-strength materials may suffer a loss in ductility, but frac-
ture in a ductile manner. Sensitivity to fracture increases with a decrease
in strain rate and an increase in hydrogen content of the atmosphere. The
temperature dependence of hydrogen embrittlement is low at low temper-
atures, increases to a maximum at intermediate temperatures, and then
decreases again at still higher temperatures.

Hydrogen pickup and subsequent embrittlement associated with electrochemical corrosion is considered to be a form of stress-corrosion cracking, and they are discussed below.

Some of the following discussion is taken in part from Reference 45, and
the reader is referred to that text for more detail. Stress-corrosion cracking
(SCC) is the result of a synergistic interaction of the presence of a materi-
al under load (or residual stress) in a corrosive environment. Features of
SCC include:

- The required stress may be very low (considerably less than the
  yield strength).
- The necessary concentration of the corroding species may also be
  very low.
- The amount of corrosion products visible on the surface of the part
does not correlate with sensitivity to SCC or time to fracture; that is,
  the part may show little visible indication of an impending failure.
- Stress-corrosion cracking generally occurs in systems that in the
  unstressed condition show little corrosion attack.
- At the macroscale, SCC fracture is brittle and can occur in alloys
  that show considerable ductility in the absence of the corroding
  environment (e.g., austenitic stainless steel).
- At the microscale, SCC is often, but not always, intergranular.

For those metal systems that rely on a passive film for corrosion protec-
tion, SCC occurs when the environment causes film rupture and the film is
unable to repair itself. The passive film is quite thin (less than 100 Å) and
typically not macroscale (and in some instances, microscale) visible. The
formation of a protective film and the conditions for its breakdown are
measured electrochemically via polarization curves (Figure 64). Figure 65
illustrates the change in polarization curves for sensitized stainless steel.
Because passive film rupture is a necessary prerequisite to SCC, contem-
porary laboratory procedures to examine SCC consider it to be a two-step
process of crack initiation and crack propagation. Prior to the develop-
ment of fracture-mechanics-based concepts, laboratory procedures to study
SCC used statically loaded tensile and bending specimens exposed
to the environment. The time to fracture was measured as a function of the
Figure 64. Potential ranges of stress-corrosion cracking by hydrogen embrittlement, cracking of an unstable passive film, and cracking initiated by pitting. Source: Reference 45.

Figure 65. Potential curve for an austenitic stainless steel showing changes in the active corrosion region with degree of sensitization. Source: Reference 45.

**applied stress.** Today, there are well-established procedures based on linear elastic fracture mechanics to measure crack propagation and the critical value of the stress-intensity factor for propagation in a specific environment ($K_{ISC}$). Figure 66 shows typical data and defines three regions of crack propagation. Two types of behavior are observed: (1) stage I behavior, in which SCC crack growth occurs at diminishingly small values of $K$ and rises linearly to a plateau and (2) stage I behavior is absent, and the curve rises almost instantaneously to stage II behavior, which then defines $K_{ISC}$. Aluminum alloys tend to show stage I behavior while stage I is absent in stainless steels (Reference 45).
Prior to the development of fracture-mechanics-based concepts, simpler laboratory procedures using U-bend and other geometry specimens were popular and still remain popular today for qualitative work. The development of standard tests to measure crack initiation in a corrosive environment is currently an active research area.

Small changes in a variable (say, alloy composition) can initiate cracking because of the onset of general film dissolution or pitting. A change in dissolved oxygen is also a major factor because the dissolved-oxygen concentration can shift the corrosion potential to values where a passive film is not stable.

In addition to anodic polarization curves discussed previously, other testing procedures have been developed to determine sensitivity to stress corrosion. Metallographic examination is useful for determining grain-boundary attack. Because a tensile stress is a necessary prerequisite to SCC, several test specimen configurations are available to create both bending and axial loading as noted previously.

Stress-corrosion cracking can be both transgranular and intergranular. For some alloys, the cracking mode may change from IG fracture to TG fracture depending on the corroding medium. Intergranular fracture is usually associated with microstructures in which a grain-boundary precipitate is present and in which there is a solute-denuded zone adjacent to the grain boundary. The grain-boundary precipitate may be quite small and difficult to identify. Therefore, it is possible for a given alloy to show either, or both, IG and TG fracture. Transgranular fracture in stress-corrosion systems can create fractographic features in fcc alloys very similar in appearance to cleavage fracture observed in bcc alloys (see Figures 67 to 69). Etch pitting of the fracture surface has been used in an attempt
Figure 67. Stress-corrosion cracking IG fracture plus TG fracture grooving attack in an 18Cr-10Ni alloy. 5 N H<sub>2</sub>SO<sub>4</sub> + 0.1 N NaCl. Source: Reference 46, p 133.

Figure 68. Stress-corrosion cracking cleavagelike fracture in a 16Cr-16Ni alloy. Source: Reference 46, p 144.

to identify the fracture planes and therefore to develop possible dislocation mechanisms for crack nucleation. Transgranular crack propagation in some instances occurs by the formation of deep crack tunnels that leave distinctive grooves on the fracture surface. Tearing then occurs subsequently between the tunnels, resulting in grooved regions bounded by dimples.

Liquid-metal embrittlement (LME) is a phenomenon in which ductility and fracture stress are reduced by surface contact with a liquid metal. The fracture stress may be well below the yield strength of nonembrittled material, and the time to fracture decreases as the stress level increases.
Figure 69. TG fracture in a 18Cr-10Ni steel in a MgCl₂ environment. The fan-shaped aspect of the fracture is a very commonly observed structure. Source: Reference 46.

The necessary amount of embrittling metal can be very small, micrograms in some cases. Fracture is often intergranular and with little branching, but, in other cases, may be caused by transgranular cleavage and in a few isolated instances may be caused by MVC.

Stress concentrators can have a deleterious effect on failure due to LME. As a result, it is common to study LME using fracture mechanics concepts and to measure the critical stress-intensity factor for LME failure. This critical value can be considerably lower than the critical value for plane-strain fracture toughness in an inert environment, and also lower than that for SCC conditions.

Identification of LME failures and observation of the fracture surface requires metallographic examination and/or examination of the fracture surface at a magnification high enough to identify microscale fracture processes. The liquid film causing LME can be tenaciously bonded to the fracture surface and must be removed to identify LME.

Two historically well-known cases of LME involve embrittlement of brasses by mercury (season cracking) (1914) and the embrittlement of railroad axles by copper in friction bearings (1914).

Liquid-metal embrittlement is still not completely understood. Only certain alloy/molten-metal combinations result in embrittlement, and, although certain conditions are generally accepted as being necessary for embrittlement, these same conditions may not be sufficient. Liquid-metal embrittlement is not a diffusion-controlled process, nor is it a corrosion process. It is considered to be a special case of embrittlement when the environment is not inert. Embrittlement apparently has a weak dependence on time of exposure to the embrittling liquid. It is common for the embrittling species to have low solubility in the component, but again, exceptions exist. (One exception is the embrittlement of galvanized steels
by molten zinc in which case zinc penetrates the grain boundaries during hot-dip galvanizing.)

Variables of importance include:

- Composition of the base alloy and the liquid metal
- Temperature
- Strain rate
- Grain size
- Thermal/ mechanical history

Some combinations resulting in embrittlement can be found in Reference 47. Temperature effects are sometimes unpredictable, but embrittlement is usually worst just above the freezing point of the embrittling liquid, although there are several cases in which embrittlement occurs below the freezing point of the liquid (solid-metal embrittlement, or SME). An increase in temperature decreases embrittlement so that LME couples show a ductile-brittle transition with temperature, changing from fracture or cleavage to dimpled rupture at higher temperatures. Grain-size effects are similar to those for fracture in a benign environment; the fracture stress varies inversely with the square root of the grain diameter and an increase in grain size increases the DBTT. Strain-rate effects are also similar to those in a benign environment: an increase in strain rate increases the DBTT.

Materials subjected to a source of neutrons, as in nuclear reactors, may become embrittled. The most obvious change in mechanical properties is an increase in the DBTT as measured by a Charpy test (Figure 70) or an increase in the nil-ductility temperature as measured by a drop-weight test. Variables affecting behavior include:

- Composition and microstructure
- Neutron fluence (dosage)
- Temperature

Radiation (Neutron) Embrittlement

![Figure 70. The shift of the Charpy impact curve to higher temperatures due to irradiation. Source: Reference 48.](image)
Tempered-martensitic microstructures are less susceptible to embrittlement than tempered bainite (especially upper bainite) and pearlitic microstructures. Phosphorus (above 0.015%) and copper (above 05%) are detrimental to behavior.

Below the DBTT temperature, embrittled steels show both transgranular cleavage and IG fracture, while the unembrittled material shows only cleavage. Above the DBTT, embrittled materials show a change to smaller and shallower dimples.

Austenitic stainless steels have also been reported to show degraded behavior due to irradiation-assisted SCC when exposed to radiation in a hot-water environment (Reference 49).

Embrittlement of a nominally ductile member can occur as the result of (1) low-temperature service (bcc materials), (2) high-temperature service at the minimum in ductility with temperature, (3) improper heat treating, (4) several oxidizing and reducing manufacturing and service environment conditions. In many, but not all of these cases, fracture is intergranular rather than transgranular. Since the number of causes for IG fracture is somewhat limited and is often alloy specific, identification of IG fracture in a failed component is helpful in deciding root cause for failure. Identification may be made by either examination of the fracture surface or a metallographic section adjacent to the fracture surface. A metallographic section taken perpendicular to the fracture surface is a good way to examine the role of the microstructure in controlling the fracture process. Unfortunately, this test is destructive and therefore may require approval if litigation is involved.

This lesson covers several topics that have been reported extensively in the literature. A few selected references are cited below for the major areas covered in this lesson.

**General**


**Fracture Mechanics**

Fracture mechanics and the general issue of brittle fracture induced by nonmetallurgical variables

Embrittlement in Steels


Environmentally Assisted Fracture


Failure Analysis and Fractography


References


### Metric Conversion Factors

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<td>°C</td>
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#### Multiple and submultiple units

| 10¹² | tera | 10⁻¹ | deci |
| 10⁹  | giga | 10⁻² | centi |
| 10⁶  | mega | 10⁻³ | milli |
| 10³  | kilo | 10⁻⁶ | micro |
| 10²  | hecto| 10⁻⁹ | nano |
| 10   | deka | 10⁻¹²| pico |

#### Abbreviations

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