

T73S02 Session 18: Toughness Testing

Last Update: 5/11/15

Relates to Knowledge & Skills Areas 3.1 to 3.7

Toughness testing; Why a K test is not possible for tough materials; Basics of J testing; Estimation formulae $J \approx \mu U / A$; Why η is ~ 1 in tension and ~ 2 in bending; Standard geometries (CTS, Charpy); Validity and size requirements; Valid tear length; Fatigue sharpening; Constraint; Side-grooving; multi-specimen method; Unloading compliance; DCPD, ACDP; Mechanisms of brittle fracture; mechanism of ductile fracture; Embrittling effects in service or in fabrication; Typical magnitudes of toughness.

Qu.: How could the fracture toughness of a brittle material be measured?

The point here is that K_{Ic} for a brittle material can be measured by the straightforward method of determining the load at which a cracked specimen undergoes brittle fracture. This will be a well defined load for a brittle failure. Moreover, the value of the LFM SIF can be evaluated at that load, if necessary by FEA, and that provides a direct measure of K_{Ic} .

Qu.: Why would this not work for a ductile material?

For a ductile material, neither of these conditions applies. That is, fast fracture, whilst it might occur eventually, does not indicate the salient load when tearing starts (initiation toughness). How do we know the load when tearing starts? Moreover, the LFM SIF evaluated at any relevant load (initiation or failure) is not the indicative quantity, because the crack tip fields are not LFM.

To have any hope at all that LFM conditions would apply, the reference stress would have to be less than the yield strength. Suppose the yield strength were 150 MPa. If the specimen is loaded predominantly in bending and has a crack about half-way across the section, the uncracked stress would have to be less than $150/4 = 38$ MPa. If the toughness were, say, $140 \text{ MPa}\sqrt{\text{m}}$, and assuming a compliance factor of about 1.5, the crack length would have to be long enough to give $1.5 \times 38\sqrt{\pi a} = 140$, i.e., $a \sim 2\text{m}$, or a specimen of size $\sim 4\text{m}$. Clearly this is not practical. And even this would not really be big enough to produce elastic conditions, since the reference stress is approaching the yield strength.

Qu.: What historical measures of toughness were used for ductile materials?

Historical toughness measures were based on the energy absorbing capacity of some standard specimen. The most important is Charpy energy, since Charpy energies are still used in some design codes. The standard Charpy specimen has a 10mm x 10mm square section and is 55mm long. It contains a 2mm deep 45° V-notch with a 0.25mm root radius. A Charpy test involves a hammer on the end of a pendulum. From a recorded height, the hammer swings and breaks the specimen. The height to which the hammer swings after breaking the specimen is recorded. The difference of the two heights provides a measure of the energy required to break the specimen. The larger this required energy, the tougher the material.

The drawback of Charpy energy as a measure of toughness is that it provides no basis for making assessments of plant components. I suppose that, if one were assessing an impact load – and the strain rate happened to be comparable with that of a Charpy test – then the energy per unit area could be used as the basis of the assessment. But how do you apply Charpy energy to the assessment of static loads? In practice the only

way is to translate the Charpy energy to a fracture toughness in terms of K (understood as \sqrt{EJ}). But such correlations are very imperfect and vary with the material or yield strength. Nevertheless, Charpy energies are still used, e.g., in irradiation embrittlement studies (both Magnox and Sizewell B), in low temperature requirements (PD5500), and for the avoidance of brittle fracture generally (Eurocodes, e.g., EN13445).

Qu.: How does a J-based toughness test work?

Essentially what is needed to carry out a tearing, or J, test is two things,

- We need to be able to evaluate J for the specimen at any load; and,
- We need to know the tear length – or at least we need to know when tearing has just started.

Qu.: How is J found?

Unlike LEFM K , the value of J does not depend only upon the load, but also upon the displacement (strains). Hence, the value of J at a given load, and for a given specimen geometry and crack size, will vary from material to material according to the degree of plastic straining.

In principle the determination of J is not problematic. Provided that we know the stress-strain curve for the material, we can carry out a finite element analysis of the specimen which will give us J at any applied load. However, this is not a practical approach. It would mean that tensile tests would first be required, followed by FE analysis – before the fracture toughness test could be done. This would make the process very lengthy and more costly in material.

A method which was used in the early days was to appeal to the energy release rate interpretation of J. Hence, two tests were carried out with slightly different crack lengths. The area under the load-displacement curve gives a measure of energy, U . The difference in this energy between the two specimens, at the same displacement, provides the “energy release”, δU , due to the change in crack length. Dividing this by the difference in the crack areas, δA , provides an estimate of J. However, this method involves small differences in large quantities, and hence both δU and δA are subject to large experimental errors. Moreover, it requires twice as much material and twice the effort to carry out.

The method that is generally used these days is to appeal to a correlation between J and U, the energy measured (roughly) as the area under the load-displacement curve. Thus, J is found at any point during the test from,

$$J = \eta \frac{U}{A} \quad (1)$$

where the “eta factor”, η , is a dimensionless number typically roughly 1 or 2. The area, A , is the area of the uncracked ligament ahead of the crack.

A refinement of this empirical relationship which is more commonly used is to treat the elastic part of the energy, U_e , separately,

$$J = \eta_e \frac{U_e}{A} + \eta_p \frac{(U - U_e)}{A} \quad (2)$$

The first term, the elastic part, may alternatively be replaced by an expression in terms of the elastic SIF,

$$J = \frac{(1-\nu^2)K^2}{E} + \eta_p \frac{(U - U_e)}{A} \quad (3)$$

BS7448 gives $\eta_p = 2$ for single-edge notch bend specimens whilst for a CTS

$$\eta_p = 2 + 0.522(1 - a_0/W).$$

BS7448 calls for the use of the original ligament area in the J-estimation formula, as opposed to the reducing area as the crack grows. I'm not sure why, although this assumption does give a more conservative (i.e., smaller) toughness.

Qu.: How is the eta factor found?

The value of empirical relationships of the form $J = \eta \frac{U}{A}$ is that the value of η is insensitive to the material (though it depends upon the specimen geometry and crack length). This is because the bulk of both the load and the displacement (strain) dependence of J is captured by the energy, U . In practice, η is determined as a function of crack length, for a given specimen type, by finite element analyses. Testing standards specify the values of eta factors to be used for standard geometries.

As a rule of thumb, $\eta \approx 1$ for geometries in which the stress is membrane dominated, whereas $\eta \approx 2$ for geometries which are bending stress dominated. However, more accurate values are required in practice (as given above for the CTS).

Qu.: Why is $\eta \approx 1$ for membrane dominated and $\eta \approx 2$ for bending?

The crudest level of analysis is to assume perfect plasticity and that the elastic part of the energy is negligible compared with the total energy. In a tension (membrane) geometry, the energy is then given approximately by $U = F_y D$, where $F_y = \sigma_y A$ is the yield load. The ligament area can be written $A = t(w - a)$, where w is the uncracked specimen width, a is the crack length, and t the specimen thickness. Hence,

$$U = Dt\sigma_y(w - a). \text{ But, at a fixed applied displacement, } J = -\frac{1}{t} \frac{\partial U}{\partial a} = D\sigma_y. \text{ But}$$

$$\frac{U}{A} = D\sigma_y, \text{ and hence } J = \frac{U}{A} \text{ and thus } \eta = 1. \text{ QED.}$$

The bending case is argued in a similar manner. The energy is then given

approximately by $U = M_y \theta$, where $M_y = \frac{\sigma_y t}{4} (w - a)^2$ is the yield moment, and θ is

the resulting specimen rotation. This yield moment is for a rectangular section. For other shapes the denominator of 4 would vary. However, for any bend specimen the quadratic dependence of the yield moment on the ligament would continue to hold,

i.e., $M_y \propto (w - a)^2$. So, more generally we write $M_y = \frac{\sigma_y t}{N} (w - a)^2$ for some number,

N. At a fixed applied displacement (rotation), $J = -\frac{1}{t} \frac{\partial U}{\partial a} = \frac{\theta \sigma_y}{N} \cdot 2(w - a)$. Hence,

$$\frac{J}{U} = \frac{2(w - a)}{t(w - a)^2} = \frac{2}{A}, \text{ or, } J = 2 \frac{U}{A}, \text{ so that } \eta = 2. \text{ QED.}$$

Work hardening and the elastic contribution to the energy will mean that the η factors differ from these perfectly plastic values.

Qu.: How is the tear length measured?

Multi-Specimen Method

One of the early techniques was simply to use a number of specimens, and take each one to a different load. The art was to judge what load might just cause tearing to start and to take one specimen to around this load, and the others to higher loads. The specimens would then be broken open by dunking in liquid nitrogen and hitting them with a hammer. It is then easy to distinguish on the fracture surface between the tearing which has occurred under test (especially if this was at high temperature) and the brittle fracture at cryogenic temperature. [For ferritic materials, the brittle fracture will be cleavage, i.e., transgranular, and clearly distinct from the ductile tearing under test. Even without a change of fracture mechanism, if the test was at high temperature, discolouration of the torn surface will provide adequate discrimination]. The tear length, Δa , can then be measured directly under an optical microscope. Hence, for each specimen we have a value of J and a value of Δa corresponding to the end of the test on that specimen. Plotting the points of J against Δa , one for each specimen, gives us a “ J -resistance curve”. By extrapolation or interpolation, the value of J at some conventional level of tearing (e.g., 0.2mm) provides our initiation toughness measure.

This method suffers from two major drawbacks. Firstly it is very hungry on material. It takes several specimens to get a single measure of toughness. More seriously, the logic of the process requires all the specimens to start from exactly the same crack length. In practice this is impossible to contrive due to the requirement to fatigue sharpen (see below).

Single Specimen Methods – DCPD/ACPD

It is far better to obtain a measurement of the crack length during the progress of the test. One means of doing this is to use electrical potential drop. Electrodes are attached to the specimen on either side of the crack plane. The deeper the crack, the smaller the cross-section through which current must flow and hence the greater the electrical resistance. Hence, the potential difference needed to maintain a fixed current will increase as the crack grows, thus providing a measure of the crack length. It is a non-linear measure, however, and needs careful calibration. [Incidentally, any computer program which can treat steady state thermal conduction problems can be used for DCPD – the problems are mathematically identical: temperature = potential, heat flow = current].

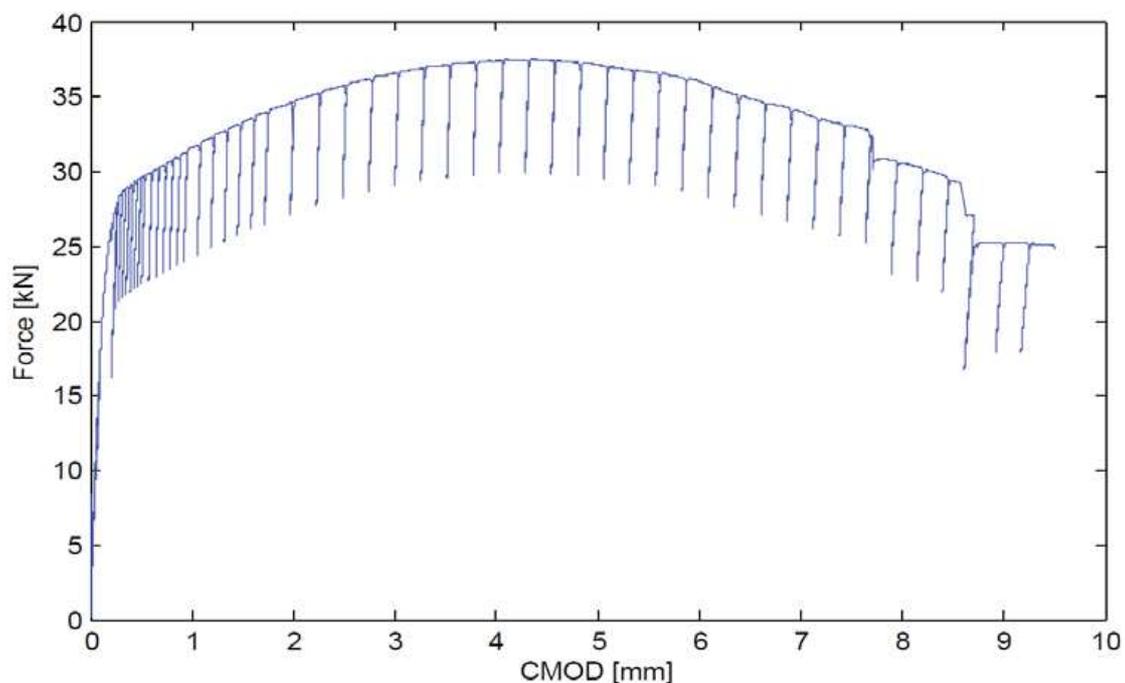
Potential drop techniques can be used with either DC or AC current (DCPD and ACPD).

Whilst DCPD/ACPD are the methods of choice in creep crack growth tests, these days the preferred method for toughness tests is “unloading compliance”. The reason for PD techniques being less favoured in toughness tests is due to the increasing plastic strains as the crack grows. The plastic strains can also cause potential changes, which confuses the calibration.

Single Specimen Methods – Unloading Compliance

The unloading compliance method uses the fact that the elastic stiffness of the specimen varies with crack length. Moreover, when a specimen is unloaded, it unloads “down the elastic line”. So the elastic compliance (reciprocal of stiffness) can be measured at any time during the test by reducing the load temporarily. It is obvious when this technique has been used. The raw data from a toughness test is the load-displacement trace. The unloading periods are clear to see as elastic unloading lines at intervals along the curve. Automated facilities are available to the experimentalist to obtain the compliance directly from the logging/control devices used (i.e., you don’t have to measure it yourself off the traces – though you can do so as a check).

Example of Unloading Compliance:
Load versus CMOD graph for a SENB-specimen test
(from Minnebruggen et al, Sustainable Construction and Design, 2011)

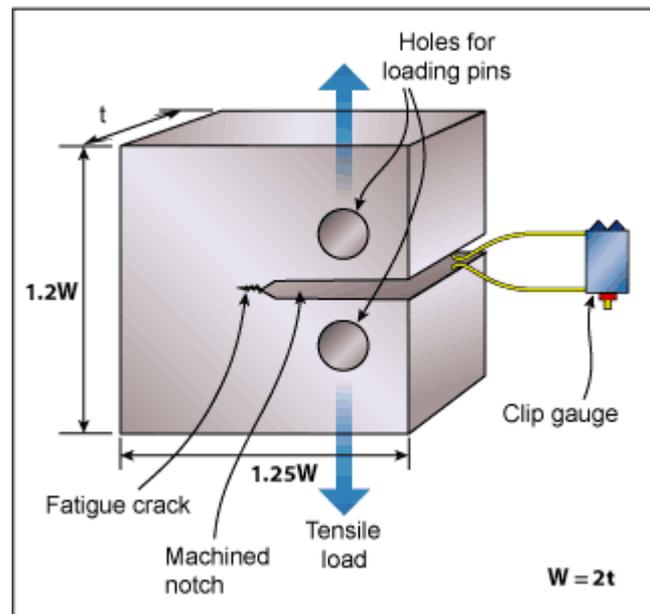


Qu.: What are the standard specimen geometries?

The most common, and preferred, geometry is the Compact Tension Specimen (CTS). It is important to realise that the name is a serious misnomer. Whilst there is a membrane component to the stressing, the stress is dominated by bending. Hence the eta factor is close to 2 (in fact ~ 2.25 for typical crack lengths). The standard “2-inch” CTS is actually rather larger than 2” square, and 1” thick. (It is 2” from the centre of the loading holes to the back face). BS7448 allows a 2” CTS to be between 0.5” and 2.5” thick.

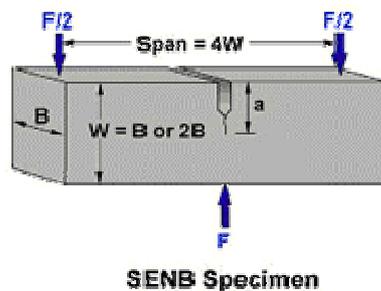
Where there is insufficient material available, a CTS may be made with scaled-down dimensions, but retaining the same shape (aspect ratio). Similarly, to investigate constraint effects or to ensure validity at longer tear lengths, a CTS with scaled-up dimensions might be used. However, the larger the specimen, the bigger capacity testing machine is required.

Example: Compact Tension Specimen (no side grooves)



It is also quite common to use Charpy sized specimens, but with the standard V-notch either replaced by a “crack” or with a crack grown from it. This can be tested in 3-point bending or 4-point bending. BS7448 sanctions a range of 3-point bend geometries, of which the Charpy geometry is a special case. Like the CTS, the recommended geometry has thickness (B) equal to *half* the in-plane dimension (W) – and hence is not of Charpy form.

Example Three-Point Bend Specimen (may be roughly of Charpy form)



Possible tension geometries are edge-notched tension specimens or centre cracked plates. However, these tend to be used only when investigating constraint effects, and are not the preferred choice for valid (constrained) toughnesses. The same is true for circumferentially cracked tubes or bars.

Qu.: How is a sharp crack introduced?

To ensure the toughness measurement is valid the crack should be sharp. In truth, for materials of reasonably high toughness you might get much the same result for a blunt tip, so long as the root radius is sufficiently small. An EDM notch might suffice. However, there is no guaranteeing that this is the case – unless comparisons are done with the toughness derived from sharp cracks. And for less tough materials, a sharp crack is essential. Even the best machining methods (EDM, or a diamond slitting wheel) will produce a rounded notch root. To obtain a sharp crack the usual procedure is to subject the specimen to fatigue load cycling prior to the main test. Care must be taken not to use too high a load during fatiguing. Any significant plastic straining of

the ligament would undermine the toughness test - and excessive fatiguing will break the specimen. Growing a fatigue crack of just a few mm is sufficient.

The disadvantage of fatiguing is that there is less control over the length of crack at the start of the toughness test, because the fatigue growth is not exactly predictable.

Qu.: What is “side-grooving”?

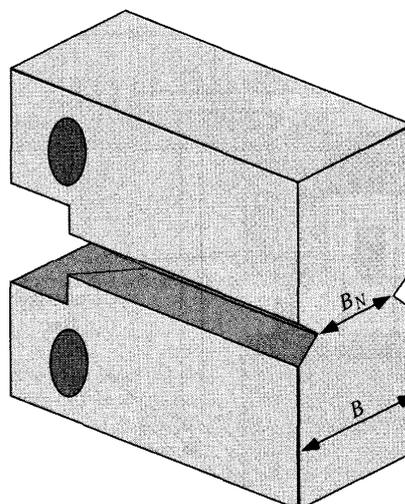
These days, CT specimens are most often side-grooved as standard. In fact side grooving is obligatory under BS7448 for use with the “single specimen technique”. This involves machining a straight V-notch to a depth equal to 10% of the thickness in both faces of the specimen along the crack line, thus reducing the ligament area by 20%. A wide range of notch angle is acceptable (30° to 90°) and the root radius should be $0.4 \pm 0.2\text{mm}$ (in BS7448 anyway).

The use of side-grooving relates to constraint. Material is more highly constrained (which means, remember, that stresses are more triaxial) near a notch. Hence, by introducing an additional notch feature across the whole specimen width, and on both faces, the average constraint on the ligament is increased. Since valid toughness measurements require high constraint, this improves the toughness measurements (i.e., it is likely to make the measured toughness *smaller*). But, in addition to this, there is also a related practical matter.

Without a side-groove, the crack will tend to tear by unequal amounts at different depths. Generally, the tear length will be greatest mid-thickness, and less at the surfaces. The initially straight crack front thus takes on a thumb-nail shape after tearing. The reason for this lies in the differing constraint at different depths into the specimen. The free surfaces are necessarily in plane stress, because this condition is imposed by the out-of-plane stress being zero at the free surface. However, constraint will increase at depths within the specimen as the out-of-plane stress component builds. This is confirmed by the greater tear lengths (remember that plain strain promotes fracture, or tearing, due to the higher hydrostatic stress component).

However, with side-grooves it is found that the tearing is much more uniform along the crack front. Essentially we have got rid of the plane stress bits near the surface. Whether the rest of the specimen is *fully* constrained (plane strain) is a moot question – but it is certainly *more* constrained.

Side-Grooved CTS



Qu.: How are side-grooved specimens analysed?

This is quite simple. It is good enough to re-interpret the ligament area A as the net area between the side-grooves, and then to use the same empirical J-estimation

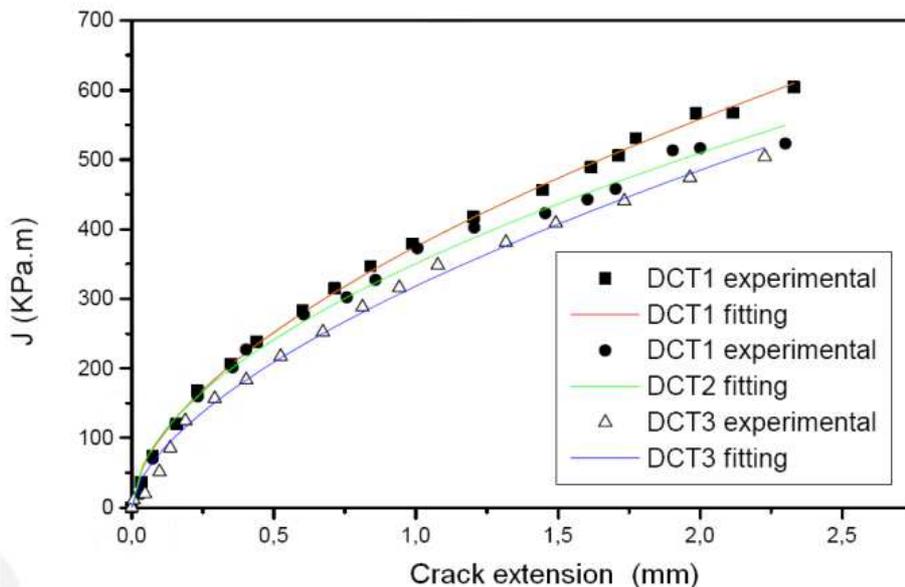
procedure as before, $J = \eta \frac{U}{A_{net}}$ where $A_{net} = B_N(W - a)$ - see graphic. In practice

there may be refinements to this – for which please consult the experimentalists and/or the testing standard in question.

Qu.: What is the definition of “initiation” toughness?

In ductile materials the crack will tend to blunt due to plastic straining prior to tearing. Thus, the discrimination of very small amounts of tearing can be confused by the blunting. For this reason the initiation toughness is not usually defined with respect to zero tear length (although, in principle, this could be done by extrapolation). Instead a small but non-zero tear length is used as a standard, most often 0.2mm. An early method was to fit the data on a plot of J versus Δa , and then read off the “initiation” toughness as the J value at a tear length of 0.2mm.

Example J-R Results



However, procedures these days (e.g., BS7448 or the ESIS standard) prefer to define initiation as the point of intersection between the J-resistance curve obtained from fitting the test data and a “blunting line”. This will generally lead to a larger toughness, since the implied tear length is greater (for example, see the toughness recommendations for 316ss parent material in R66). For example, the ‘blunting line’ is defined in BS7448 as having a slope on the J v Δa plot of 3.75 times the UTS at the test temperature.

Qu.: What is the “tearing modulus”?

The testing procedures described above provide a curve of J versus tear length. This is the J_R resistance curve. The tearing modulus is the slope of the J resistance curve after the point of initiation, $\frac{dJ}{da}$. If a linear approximation is adequate, the toughness after

tearing can be estimated as $J_{\Delta a} = J_{0.2} + \frac{dJ}{da}(\Delta a - 0.2mm)$.

The upward slope of the J resistance curve is a result of the stability of tearing. The material/specimen can sustain loads greater than the initiation load without breaking, but with increasing amounts of tearing.

Assessment methodologies, and indeed the J_R resistance curve itself, give the impression that the stability of ductile tearing is due to the material becoming tougher after tearing. The scientific nature, or otherwise, of this point of view makes for an interesting discussion. This issue, the physical nature of stable tearing, and why it is stable, is discussed in a future session.

Qu.: What is meant by the “validity” of a fracture toughness measurement?

The theoretical meaning of “validity” has been discussed in a previous session. It relates to the crack tip fields being of HRR form (if power law hardening were assumed). There are many toughness testing standards but they share common features. The validity limit is usually expressed as a minimum size requirement for the test specimen (and this is usually a function of geometry, i.e. constraint). This size requirement is often that both the thickness and the remaining ligament should exceed some stated multiple of J_i / σ_y . For bending dominated specimens (e.g. CTS) this multiple might typically be x20. For tension geometries it will be much larger (perhaps x200) due to the lower constraint. Consequently tension geometries would have to be very large to produce valid (fully constrained) toughnesses, and hence are little used.

The validity limits on the tearing resistance curve are usually in two parts: an upper limit on J_R and an upper limit on Δa . The upper limit on Δa may simply be a geometrical limit, e.g. 10% of the initial ligament. The upper limit on J_R is typically $B\sigma_f / 20$, where B is the specimen thickness and σ_f is the flow stress (average of 0.2% proof and UTS).

Qu.: What is a typical upper limit on the valid tear length?

For typical pressure vessel steels and standard toughness specimen sizes (2” CTS), the valid limit on tear length is typically in the range 0.6mm to 1.4mm. Note however that this will generally be the tear length when J reaches the maximum valid magnitude, e.g., $B\sigma_f / 20$, rather than the explicit limit on tearing. (10% of the ligament will generally be greater for a 2” CTS).

However, if pressure tests are carried out on large vessels containing cracks, it is found that the best means of estimating the maximum sustainable pressure is to continue the tearing assessment to the point of instability, which might be many mm of tearing.

Qu.: What are the metallurgical mechanisms of fracture?

Ductile Fracture

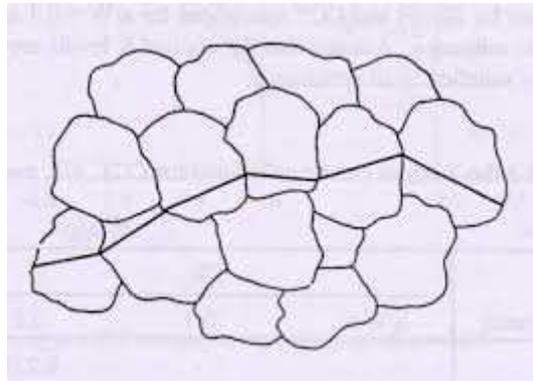
For tough steels in the ductile regime, the mechanism of tearing is “micro-void growth and coalescence”. This produces a dimpled fracture surface appearance.

Brittle Fracture

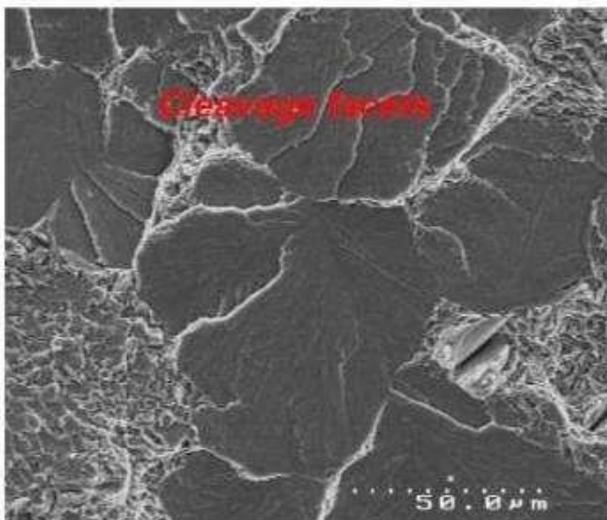
There are broadly two classes of brittle fracture mechanism: transgranular (also called intragranular), and intergranular.

Transgranular (intragranular) refers to the fracture path passing through the crystal matrix of the grains on crystallographic planes. The resulting fracture surface has an obviously ‘crystalline’ appearance, because the visible facets are perfectly flat across individual grains. This mechanism is also called “cleavage” because the individual grains are cleaved in two.

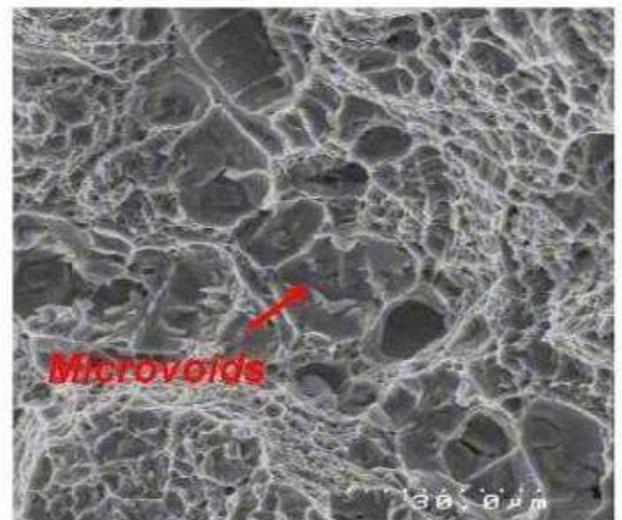
Cleavage Fracture Path



Intergranular fracture refers to a fracture path along the grain boundaries. The resulting fracture surface is also obviously ‘crystalline’ in appearance, since the individual grains are being displayed.



Cleavage Facets in Brittle Fracture



Microvoid Coalescence in Ductile Fracture

Microscopic View of Fractures

For ferritic steels at low temperatures the fracture mechanism is usually cleavage (transgranular).

Precipitation of elements such as sulphur or phosphorus onto the grain boundaries, perhaps due to exposure to irradiation, can result in brittle intergranular fracture in ferritic steels. (This discovery led to problems over the final decade or so of the life of Magnox stations with steel reactor pressure vessels).

Austenitic steels do not exhibit cleavage behaviour, and hence have superior toughness at low temperatures. (The reason relates to their fcc structure, as contrasted with the bcc structure of ferritics, see below).

Also, austenitic steels are far less sensitive to irradiation embrittlement than ferritic steels, and tend to retain good toughness even after irradiation.

Qu.: Why are austenitic steels tougher?

Why are ferritic (BCC) materials subject to brittle mechanisms, but austenitic (FCC) materials generally are not? I do not have a full explanation. However part of the reason may be that, in ferritics, the brittle fracture starts as dislocation pile-ups at carbide precipitates within the grains. These brittle carbides fracture first, and the crack then extends into the rest of the grain. This mechanism either does not occur in austenitics, or, if carbides fracture, the micro-crack does not propagate into the rest of the grain. The lower likelihood of a carbide initiated fracture in austenitics may simply be because of their lower carbon content. (300-series stainless steels have 0.03% to 0.08% carbon, compared with up to ~0.4% carbon in CMn steels, and up to ~0.2% carbon in low alloy ferritics). But there is something missing from this explanation.

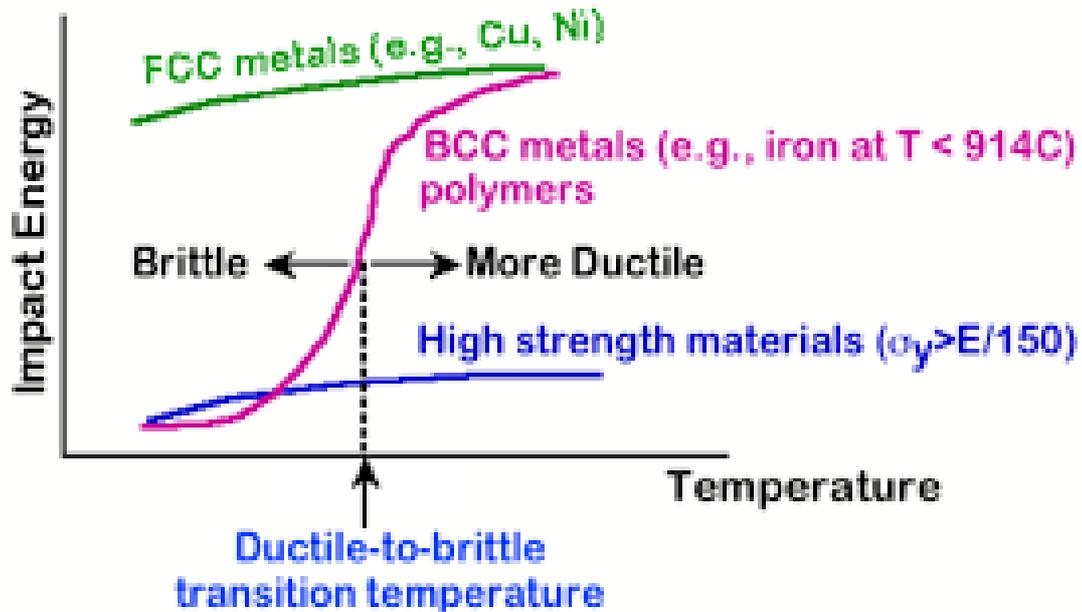
Another difference between FCC (austenitic) and BCC (ferritic) is the larger number of active slip planes in the former – which is why austenitics tend to be particularly ductile.

A further explanation for the difference between BCC and FCC structures may lie in the fact that, at low temperatures, the fracture of ferritic steels appears to occur primarily at twins.

However, austenitic steels *can* exhibit low toughness behaviour. They are subject to thermal ageing at temperatures above about 500°C which substantially reduces their fracture toughness and can put them in the ‘brittle’ regime. There are probably two mechanisms which contribute to this. One is the precipitation, or coarsening, of brittle phases, e.g., carbides. The other is the gradual conversion of residual delta ferrite into sigma phase. This is an inter-metallic phase (neither austenitic FCC nor ferritic BCC) whose crystallographic structure does not have ready slip planes, i.e., it’s brittle. It is sigma phase which has the potential to reduce impact energies of austenitics to horribly low levels. But this will happen only if, (i) there was enough delta ferrite originally, (ii) the delta ferrite is distributed adversely (so as to potentially form continuous cracks), and, (iii) the time-at-temperature is sufficient to drive the thermodynamic conversion (it probably needs something above 500°C – maybe quite a bit above, I’m not sure).

Qu.: How does the fracture toughness of ferritic steel vary with temperature?

In order of increasing temperature: lower shelf / transition region / upper shelf. For CMn steels the transition region can be, extremely roughly speaking, around room temperature. This might mean quite brittle behaviour at room temperature, or even slightly above. It might mean that the toughness of the steel varies markedly between $\sim 20^{\circ}\text{C}$ and (say) $\sim 60^{\circ}\text{C}$. For this reason, the proof testing of ferritic pressure vessels is sometimes stipulated to be carried out 'warm', perhaps above $\sim 60^{\circ}\text{C}$.



The transition region spans a range of temperatures, so the definition of a “transition temperature” is rather arbitrary. Some definitions are:-

- (a) FATT = fracture appearance transition temperature. This is defined as the point at which a microscopic examination of a fracture surface appears to be $\sim 50\%$ cleavage and $\sim 50\%$ ductile. So it could be rather subjective;
- (b) An arbitrary Charpy energy level, usually 40J;
- (c) ASME uses NDTT = nil ductility transition temperature. This is obviously a lower temperature than the FATT, but tends to be more sharply defined.

The lower shelf toughness of structural ferritic steels are typically $20 - 40 \text{ MPa}\sqrt{\text{m}}$, whereas upper shelf toughnesses are generally above $100 \text{ MPa}\sqrt{\text{m}}$, perhaps typically $140 \text{ MPa}\sqrt{\text{m}}$. The toughness of austenitic steels is comparable with, or greater than, the upper shelf toughness of ferritic steels.

Qu.: How does fracture toughness change due to neutron irradiation?

The broad effect of irradiation is to increase strength whilst decreasing ductility, and hence decreasing toughness. This is broadly true for both ferritic and austenitic steels, but austenitics are substantially less sensitive than ferritics. Neutron irradiation can cause very large shifts in the toughness transition temperature of ferritics, e.g. by the order of hundreds of $^{\circ}\text{C}$ in susceptible weldments at doses experienced by Magnox reactor pressure vessels (RPVs), i.e., hundreds of 10^{-5} dpa. Consequently, a change in

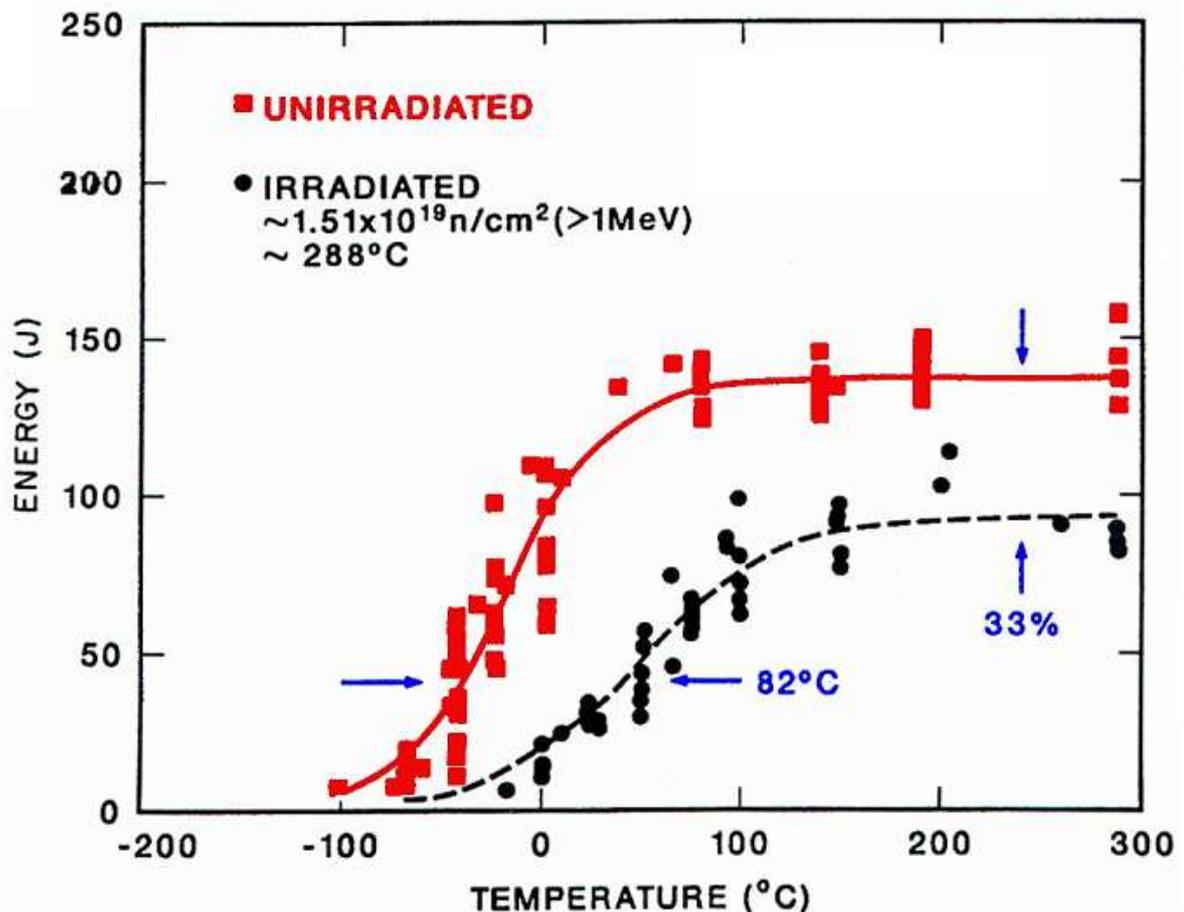
material behaviour from fully ductile to fully embrittled is perfectly feasible in strong neutron fluxes.

Dose-damage relations are a huge field. Detailed data fits are provided in R66 Section 15 for materials of relevance to AGRs. Note in particular that parent and weld may respond very differently due to differences in chemical composition, to which the irradiation embrittlement is very sensitive.

For example, in carbon-manganese steels, embrittlement is far worse in steels with relatively high copper content. For this reason Magnox RPVs were challenged by the embrittlement of the submerged arc welds used in their fabrication, since these had high copper content. Embrittlement of CMn steels is also exacerbated by high levels of P or S, since these species can precipitate onto the grain boundaries under the influence of irradiation and lead to intergranular fracture. Note that the mechanism of brittle fracture can be different in irradiated CMn steels compared with unirradiated steels below their transition temperature (i.e., intergranular fracture rather than intragranular cleavage).

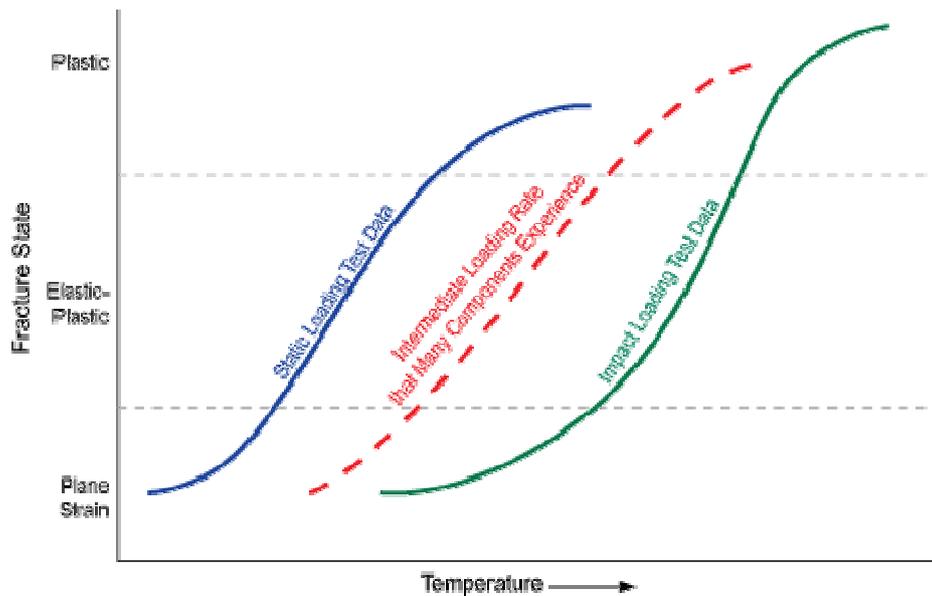
Generally, ferritic steels operating close to, or within, the transition region will be most sensitive to irradiation. The upper shelf toughness of ferritic steels is relatively insensitive to irradiation, although the tearing resistance can be substantially reduced.

Example of Effect of Irradiation on Ferritic Toughness Transition Temperature
(NB: Not necessarily representative of the effect on the upper shelf, which, at doses of relevant to, say, RPVs, usually increases)



Qu.: How does strain rate affect the fracture toughness?

For ferritic materials exhibiting a toughness transition, the toughness transition temperature is increased by dynamic loading. This means that the transition toughness is *reduced* by dynamic loading. In contrast, the upper shelf toughness is increased by dynamic loading. The lower shelf toughness is unchanged. See R66 Section 13.7 for a quantitative relation between strain rate and transition temperature. Note that the strain rate must be measured near the crack tip (see R6 Section I.5.3).



Qu.: How does exposure to high temperatures affect fracture toughness?

Exposure to high temperatures for a sustained period can lead to degradation of toughness (thermal ageing). R66 Section 13 has a limited discussion of this, including explicit guidance for austenitic steels in Section 13.4.4. Note, however, that there is more recent advice for some of the 300-series stainless steels, for which the service degradation of toughness can be substantial (see the R66 User Queries database). Weld material is more severely affected than parent.

The mechanisms behind thermal embrittlement of austenitics were discussed above. Welds tend to be more severely affected because they have a greater initial delta ferrite content (which results from rapid weld cooling) and also because their columnar grains, and columnar delta ferrite ‘fingers’ provide a opportunity for continuous cracks to form across the section.

Qu.: How does a chemical environment affect the fracture toughness?

The presence of certain chemical species can seriously degrade toughness, the most famous being hydrogen. Of course many compounds or ions can cause corrosion of various types, especially in aqueous environments. But we are concerned here with fracture toughness rather than more general chemical attack.