Testing Methods for Determination of Fracture Toughness of Metals

H. T. Corten
University of Illinois, Urbana, Illinois.

Introduction. "Low stress fracture" occurs when a structural or a machine member fractures in service at a level of nominal stress (gross section stress) below or near the yield strength. Such fractures initiate at a "flaw" or "crack" in the member. While the nominal stress may be below or at the yield strength, the stresses and strains very close to the crack front reach high values, values sufficient to cause onset of rapid crack extension.

"Fracture toughness" is a measure of the resistance of a member to onset of rapid crack extension. "Subcritical crack extension" accompanying repeated loading, environmental attack, or monotonically increasing load, frequently precedes onset of rapid crack extension. As the flaw or crack extends slowly, the intensity of the stress and strain field at the crack tip is increased. When the crack tip stress and strain field reaches a critical value, onset of rapid crack extension occurs. The purpose of a fracture toughness test is to measure a parameter such as load, deformation, energy, fracture appearance, K_{Ic}, K_c, COD, J_{Ic}, etc., preferably in a small specimen that is related to onset of rapid crack extension in the full scale member.

Observations of fracture surfaces have led to descriptive terminology such as "brittle appearance" which may mean separation on a plane normal to the nominal tensile stress (flat fracture) based on a macroscopic observation or it may mean "cleavage" separation if the observation is made with a microscope. Further, a member is said to exhibit "brittle behavior" if the load-deformation curve exhibits no measurable plastic deformation prior to fracture. While it is true that for ferritic-pearlitic steel structures "low stress fractures" usually are "brittle" in terms of both appearance and behavior, this terminology frequently is inappropriate for high strength (martensitic) steels and non-ferrous metals.

Similarly, "ductile fracture appearance" may mean separation on a plane at 45° to the nominal tensile stress (slant fracture) or fibrous appearance on the macroscopic scale and "dimpled rupture" on the microscopic level. "Ductile behavior" means that plastic deformation is observable in a load-deformation curve prior to fracture.

Each of these terms is useful and related to "low stress fracture," however, they are not synonymous with "low stress fracture." In this paper, the primary interest in "fracture toughness" is in relation to "low stress fracture."

Relationship to Conventional Mechanical Properties. "Fracture toughness" is related qualitatively to the conventional mechanical properties, specifically strength, ductility, and area under the stress-strain curve. The features that distinguish "fracture toughness" from these conventional properties arise from the fact that the flaw or crack causes a severe stress and strain concentration at the crack front, reduces the "gage length" to a minimum, and (because of the high stress and/or strain gradients) introduces a multiaxial state of tensile stress and strain in the region close to the crack border. The "mean normal stress" is increased substantially more than the "stress deviation" due to the crack (1). This phenomena is referred to as transverse "constraint." Constraint is observed to substantially reduce the ductility as compared to the ductility in the conventional tensile test (2,3,4). Both because of the difficulty of analyzing the stresses around flaws and cracks, especially in the plastic zone, and because of the lack of knowledge of the reduction of ductility under these conditions, most engineers have preferred to measure "fracture toughness" using specimens that to various degrees simulate the flawed or cracked member.

Relationship to Members that Have Fractured in Service. When "low stress fracture" has occurred in service or proof test, frequently, it is found that factors in addition to those mentioned above are present (5,6). Among these are:

(1) residual stresses which may add to the load induced stresses,

(2) material embrittlement caused by any one of several mechanisms (temper embrittlement, hydrogen, reversed straining or strain aging, irradiation, etc.) associated with fabrication, heat treatment, or service, and

(3) environmental influences such as low temperature, high strain rates, (particularly for ferritic-pearlitic steel structures) and aggressive environments.

Most of the above mentioned features have been recognized and studied for a number of years (5,6). Fracture toughness tests are needed specifically to evaluate and control the material embrittlement and environmental influences. While these factors influence "fracture toughness" and may be variables in a study, most of them can be separated from the question of methods of measurement of fracture toughness.

Fracture Toughness Measurement Parameters. While many parameters have been proposed as measures of fracture toughness, two types of parameters can be identified. The first type can be characterized as empirical and involves a quantity which is both related to fracture toughness and measurable or observable in a particular test. Examples in this category include: total energy to fracture a member (Charpy, Izod, dynamic tear test, etc.) fracture appearance (percent fibrous or percent "crystalline"), lateral expansion (or contraction), and the "transition temperature" corresponding to particular values of energy, fracture appearance, or lateral expansion.

In large preflawed wide plate tensile specimens, gross and net section stress as well as total energy for onset of rapid crack extension have been measured (5). From results of this kind obtained over a range of temperatures, a "transition temperature" may be defined.

A related series of tests allow determination of a temperature corresponding to "crack arrest" under various load conditions (5). The first tests of this kind, the Robertson and Esso crack arrest tests, measured a temperature for crack arrest starting with a particular nominal stress. A smaller limited-deformation bend test, the "nil-ductility transition" test (NDT), allows determination of the highest temperature at which a small (arc strike size) crack started in a brittle weld bead is propagated through the base metal of the NDT specimen (5).

The second type of fracture toughness parameter involves identifying a parameter based on a stress, strain, or energy analysis of the crack tip region of both a structure and a specimen. When a single parameter uniquely characterizes the crack front stress or strain field in both the structure and the specimen, a critical value of this parameter, defined by the onset of rapid crack extension, is ideal for characterizing fracture toughness. This approach, known as fracture mechanics, has been most extensively developed for the small scale yielding (linear elastic) situation (7, 8, 1). A critical value of K_I, the stress intensity factor, (or equivalently, $\mathcal{G}_I$, the crack extension force), which characterizes the crack tip stress field, serves as a measure of fracture toughness. When both small scale yielding and maximum transverse constraint are present, a single parameter, K_{Ic}, is sufficient to measure fracture toughness.* Small scale yielding and maximum transverse constraint are specified in terms of the allowable size of the "formal plastic zone" as compared with the in-plane dimensions of the specimen including the crack length, and the specimen thickness respectively (10, 1, 9). These relationships will be defined quantitatively in a later section.

---

*Other requirements for K_{Ic} include a properly fatigue cracked specimen and a small but detectable amount of crack extension precedes the determination of the critical value, K_{Ic} (9).

In a higher toughness regime, large scale yielding and limit load behavior frequently are observed, still in the presence of maximum transverse constraint. Fracture toughness parameters include: crack opening displacement (COD) (11) or stretch designated δ (conceived of as occuring close to the initial crack front), and the J-integral (1, 12). Preliminary work indicates that specimen size requirements are necessary to insure maximum transverse constraint for these specimens. From a blunted crack slip line field, a region of "intense plastic strain" directly ahead of the blunted crack but buried deep in the plastic zone has been identified as a significant dimensional parameter (1). Even deep within the plastic zone, the large crack front plastic strain gradients generate transverse constraint within this region. The size of this region of intense plastic strain is approximately 2δ. Preliminary results suggest that specimen thicknesses from 25δ to 50δ may be sufficient to maintain maximum transverse constraint even for fully plastic yielding (13). In addition, preliminary data indicate that the minimum fracture toughness can be measured only when the in-plane specimen dimensions are adjusted so that the dominate plastic deformation also is in-plane.

The J-integral represents a generalization of $\mathcal{G}$, the crack extension force. J was developed for non-linear elastic material behavior (equivalent to deformation plasticity theory). In the instances where it has been investigated, the critical value, J_{Ic}, determined from a small specimen was substantially the same as the value of $\mathcal{G}_{Ic}$ obtained using a specimen sufficiently large to satisfy the K_{Ic} size requirements (14).

Returning to the realm where fracture of a specimen occurs before the nominal stress reaches the yield point, the parameter K has been extensively used as a measure of fracture toughness when the transverse constraint is relaxed. Considering thin sheet material where the thickness is small compared with the plastic zone size and the planar dimensions of a structure or specimen (including the crack), the transverse constraint approaches a small value.

With loss of transverse constraint, the plastic zone size is increased. Further, starting with a Mode I crack, small amounts of crack extension lead to cracking on a 45° slant. It is observed that even though the crack has extended an increment, Δa, and is longer, the crack arrests, presumably because the new 45° slant crack configuration is tougher than the original Mode I crack configuration. Subcritical crack extension is observed to varying degrees during rising load tests, prior to onset of rapid crack extension, depending upon the material. For materials where the amount of subcritical crack extension during a rising load test is considerable, usually it is found that geometrically similar specimens of different sizes (same thickness) give different values of K at onset of rapid crack extension. This dependence of the critical value of K on specimen dimensions and particularly crack size is disturbing, however, it is not difficult to understand. Fracture toughness depends upon the plastic deformation mode (in-plane vs. 45° transverse slip) in front of the crack tip. During subcritical crack extension, transverse constraint is reduced and the deformation mode changes from in-plane to 45° transverse slip. Presumably the fracture toughness is increased by this change in deformation mode at least as much as the K factor is increased due to the longer crack. This discussion leads directly to consideration of R curves as a means of describing fracture toughness.

Resistance Curves and the Onset of Rapid Crack Extension. The concept of "resistance curves" or R Curves is not new, however, quantitative investigation was undertaken only recently (1, 15, 16). A fracture toughness parameter, such as $\mathcal{G}$, K, or energy, is used as a measure of resistance to crack extension as a function of the crack length. A series of three R curves are shown schematically in Fig. 1. These three curves might represent three different materials of rather different toughnesses, or they could represent the same material tested in three widely different thicknesses (constraint). In this later case, curve 1 (or 1') is appropriate for a thick specimen with maximum crack tip transverse constraint, and conversely, curve 3 (or 3') is appropriate for thin sheet material.

Consider a large plate specimen containing a small central crack of length, $2a_o$. For simplicity R will be measured in terms of the $\mathcal{G}$ value corresponding to a particular crack length. Since $\mathcal{G}$ is a linear function of a, short straight lines have been drawn through the origin cutting the R curve at different points. These short lines represent the change in $\mathcal{G}$ for small increases in crack length (at constant load). When for a small increment of crack extension, Δa, beyond a_o, these $\mathcal{G}$ lines are well below the R curve, subcritical crack extension is very limited. As the load is increased the $\mathcal{G}$ lines interesect the R curve at higher values of R and finally a $\mathcal{G}$ line is just tangent to the R curve. Now any small increment at crack extension, Δa, leads to a value of $\mathcal{G}$ that is larger than the R corresponding to the new crack length $(a + \Delta a)$. Thus the tangency point of the $\mathcal{G}$ line to the R curve marks the onset of rapid crack extension and the value, $\mathcal{G}_c$.

Now consider a second specimen of the same material and same thickness but twice as large in all in-plane dimensions. The initial half-crack length, $2a_o$, is shown on the abscissa as the starting point for a second set of R curves. This second set of R curves was drawn in Fig. 1 by simply displacing the first set (a_o) to the right along the abscissa to $2a_o$. Experimental evidence suggests that the R curves developed from specimens geometrically similar in in-plane dimensions (same thickness) will approximately superimpose following an appropriate horizontal shift. When this second specimen is tested, the onset of rapid crack extension is denoted by the point of tangency between the $\mathcal{G}$ line and the R curve. Comparing the critical values of $\mathcal{G}$ from these two experiments, it is evident that the $\mathcal{G}$ values for R curves 1 and 1' will be nearly identical due to the sharp knee in these R curves. Conversely, for R curves 3 and 3', a significant difference may exist between the critical values of for the two specimens of the same material and thickness.

Typical R curves obtained using a double compliance method are shown in Fig. 2 for an aluminum alloy sheet for two different specimen configurations (15). Heyer and McCabe elected to measure R in terms of the K value. The crack line wedge loaded specimen (CLWL) allowed R values to be measured over a wide range of values of Δa, because the K value decreases during crack extension Δa (at constant load). The center cracked tension specimen (CCT) data compares favorably with the CLWL data. Most of the R curves that have been obtained for either different specimen configurations or different specimen sizes (same configuration) agree well, however, more experience with R curves is needed.

A second type of R curve is shown in Fig. 3. Pellini and Judy (17) elected to measure R in terms of the average energy per unit area to completely fracture a dynamic tear specimen tested on the upper shelf. Using the same specimen thickness and crack configuration, the unfractured ligament (depth) of the specimen was varied. The quantity, b, in Fig. 3 is the uncracked ligament. The ordinate values, E/A, are the average for the entire run of the crack through the uncracked ligament, b. The mechanical properties of these steels are listed in Table 1. Steel F-1 exhibited a flat fracture and no increase in average fracture energy, E/A, as the uncracked ligament was increased; steel L-1 exhibited a mixed mode fracture (approximately 50 percent slant) and the average value, E/A increased modestly. Steels H-6 and H-7 displayed 100 percent shear fractures and rapidly rising R curves as b increased. The slope of the E/A-b curves increased as the upper shelf value increased for the standard 1 in. D.T. specimen. An increase in E/A accompanies the development of the shear lip (17). Further, the development of the shear lip is a function of the specimen depth to thickness ratio.

Viewed in the perspective of "unique" R curves for a given material and thickness, the point of onset of rapid crack extension is, in general, a function of member geometry including crack length, and specimen thickness and depth. To amplify on this point, specimens which meet the K_{Ic} size requirements (small scale yielding and maximum transverse constraint) behave according to curve 1 (or 1') in Fig. 1. The onset of rapid crack extension is abrupt and nearly independent of crack length (22). In this case a single parameter, K_{Ic}, is adequate to characterize fracture toughness.

For large scale yielding and fully plastic behavior where maximum transverse constraint is maintained, onset of crack extension remains reasonably abrupt. An R curve between curves 1 and 2 (Fig. 1) appears to describe the behavior (14, 18). Again a single parameter, either J_{Ic} or δ_{Ic}, appears to be adequate to characterize fracture toughness.

For situations where the transverse constraint is reduced, it appears that the R curve is modified in the direction $1 \longrightarrow 2 \longrightarrow 3$ (Fig. 1). One exception to this generality is apparent: temperature and strain rate sensitive metals such as ferritic-pearlitic steels tested below and in the transition temperature range. It appears that the strain rate sensitivity is responsible for the abrupt onset of rapid crack extension, since the fracture toughness decreases significantly in the presence of the increased strain rate sensed by material elements ahead of the moving crack.

For high strength steels (martensitic structure), some strainless steels, aluminum alloys, and ferritic-pearlitic steels (tested on the upper shelf), a reduction of transverse constraint tends to modify the R curve from $1 \longrightarrow 2 \longrightarrow 3$.

Consider steels that exhibit R curves in the range from curve 2 to curve 3 (Fig. 1), or steels L-1 thru H-7 (Fig. 3). In a structure made of large flat plates of one of these steels, the R curve depends upon both the material toughness and the plate thickness. In the structure, the run of the crack will develop the complete R curve up to the point of tangency in Fig. 1. The point of tangency depends upon depth of the cracked member and the loading (tension, bending, combination, etc.). To make a fracture toughness test that reflects the toughness that will be realized in service requires a reasonable modeling of the behavior of the service member in a laboratory specimen. An R curve is required that is obtained from a specimen of the material in the thickness that will be used in service. To be useful, the R curve must be obtained up to the point of tangency for the service configuration. From the data in Fig. 3 for 1 in.

thick plate, it appears that a "crack run" (subcritical crack extension) of 4 to 5 inches may be necessary to develop an adequate R curve in an impact bending specimen. Other "low energy blow" tests, conducted by Pellini and Judy (17), suggest that the R curves are beginning to flatten out for steels L-1 and H-6 at or beyond 3 inches of crack run.

When tests are conducted on specimens where the "crack run," b, is too small to adequately model the structural configuration (same thickness assumed), the value of fracture toughness (regardless of the toughness parameter used) will be lower than the value that will be realized in service. For a particular application and material, tests in this category that are employed for quality control purposes serve the intended purpose adequately. However, when the purpose of the test is to select and judge the adequacy of a higher strength metal for the same service, the use of a specimen that does not adequately model the R curve for the structural configuration can easily lead to questionable conclusions.

In Table 1 and Fig. 3, steels L-1 and M-5 differ only in heat treatment. Assume, for purposes of comparison, that onset of rapid crack extension occurs in a structure corresponding to a crack run, b, (unfractured ligament) of 4 inches for both materials. Similarly, assume that measurements of toughness are made with specimens which have crack runs, b, of 1 inch. From Fig. 3, energies per sq. in. of approximately 800 ft. lb. /in.^{2} and 250 ft. lb. /in.^{2} would be known from laboratory tests for steels M-5 and L-1 respectively. The actual toughness available in the structures would be 2000 ft. lb. /in.^{2} and 600 ft. lb. /in.^{2} for the M-5 steel to the L-1 steel respectively. While the ratio of laboratory measured toughnesses are approximately the same as the ratio of the toughnesses realized in service, it is difficult to specify what quality control measures would be necessary to insure the same degree of fracture free service.

With regard to "crack run," the standard Charpy V (or keyhole) notch specimen with an uncracked ligament smaller than the thickness appears to be an unsuitable specimen configuration for measurement of fracture toughness intended to simulate a large plate structure which will reach onset of rapid crack extension well up on an R curve. The R curves for many tough metals will be in the region between curves 2 and 3 in Fig. 1. The cracked structure likely will develop a 45° thru-thickness plastic deformation pattern accompanied by large shear lips or a completely slant fracture surface. Conversely, the deformation in the Charpy specimen is dominantly in-plane and approaches the plane strain slip-line pattern (24). Further, depending upon the thickness of the cracked plate structure, the degree of transverse constraint in the Charpy specimen is not likely to be the same as in the structure at the onset of rapid crack extension.

These differences between the behavior of a structure (45° thru-the-thickness deformation) and the behavior of a small specimen (in-plane deformation) may well be important if thru-the-thickness (short transverse) anisotropy, caused by variations in the steel making and rolling practices, plays an important role in determining the toughness of flat plates (1).

Some of the important concepts regarding measurement of fracture toughness emerging from these considerations of R curves are summarized in Fig. 4. With regard to transverse constraint, the extremes, full constraint and low constraint, have been emphasized. The transition region between the extremes appears to be more complex and requires further work and clarification.

Measurement Point. In discussing R curves, the point of tangency between the $\mathcal{G}$ curve and the R curve was noted as point of onset of rapid crack extension. In the region of low transverse constraint (Fig. 4), it is desirable to use a specimen from which an extensive portion of the R curve can be developed. From analysis of cracked structure, a $\mathcal{G}$ curve can be superimposed upon the R curve to estimate the fracture toughness to be realized in the structural element.

Turning attention to the region of full constraint, onset of rapid crack extension for low stress and near yield strength fractures of structures is reasonably abrupt (23). When small specimens are employed to measure fracture toughness of tough metals, their behavior is elastic-plastic and frequently fully plastic. Following onset of rapid crack extension in a small elastic-plastic specimen, the high stiffness of many deformation type testing machines leads to rapid unloading and crack arrest. As the test proceeds, this process may be repeated several times. A question arises as to which of the small irregularities in the slowly rising load-deformation curve is significant with regard to structural fracture (18, 19). Frequently, with an R curve between curves 1 and 2 (Fig. 1), a portion of the crack front will move forward providing a small increment of subcritical crack extension. When the irregularities on the load deformation curve associated with crack extension and arrest are small and comparable with the electrical noise in the recording system, the load-deformation curve, by itself, provides information unsuitable for determining onset of crack extension. This problem requires other methods for detecting onset of rapid crack extension, and attention to an "operational definition" of a measurement point.

When very tough metals are tested using specimens that develop maximum transverse constraint and a fully plastic hinge, the onset of rapid crack extension may be delayed until the maximum load point or beyond. The load-deformation curves (or M-θ curves) for 1 T CT specimens shown in Fig. 5 (19) approach a constant maximum load at a displacement of from two to ten times the elastic displacement. These curves were carefully reproduced from the original records. For larger displacements, the load decreases slightly. For a compact tension specimen or a notched beam specimen, strain hardening will increase the bending moment, however, at a decreasing rate as the deformation increases. Expansion (compression side) and contraction (tension side) change the shape of the notch cross-section and shift of the neutral axis, however, the bending moment should be little affected. The fully plastic bending moment constraint factor for a beam with a crack or sharp notch is decreased

as the notch is blunted. Using Green and Hundy's results (24) and assuming that the crack opening stretch, δ, is equal to twice the root radius, the moment may decrease up to 5 percent. The decrease in constraint factor may be responsible for the long horizontal portion of several of the curves and possibly to initial decrease of load in Fig. 5. In addition, there is the obvious possibility that small increments of crack extension caused the load to remain constant and eventually to decrease.

For a forging steel, it was observed that onset of cracking in beam specimens occurred just after the maximum load point for a rounded curve or just beyond the horizontal portion of a flat top load-deflection curve. For center-cracked tension members (plane strain), onset of crack extension occurred somewhat beyond the point of maximum load (14b). Recent experiments on 1 T-CT structural steel specimens indicated that the onset of crack extension was shifted from a point beyond maximum load for $a/w = 0.5$, to smaller displacements on the rising portion of the load-deflection curve for a/w in the range 0.6 to 0.9 (25). These considerations suggest that it is premature to select the maximum load point as the measurement point.

Dimensional Parameters. Size effects in fracture testing have been observed for many years (20). The introduction of crack length, a, through the stress intensity factor, K (or $\mathcal{G}$) for small scale yielding accounted for one major effect of size, the crack length. This was followed by the rationalization of thickness, B, as it influences transverse constraint (21). In the absence of an elastic-plastic three-dimensional stress analysis, Irwin introduced the concept of dimensional parameters. For the small scale yielding region, Irwin identified the plastic zone size, $2\,r_Y$, as the significant dimension. The formal plastic zone size was defined as:

$$r_Y = \frac{1}{2\pi}\left(\frac{K_I}{\sigma_Y}\right)^2 \qquad (1)$$

where K_I is the stress intensity factor and σ_Y is the appropriate yield strength.

For plane stress conditions, the 0.2 percent offset yield strength, σ_{ys}, has been used. For plane strain conditions, constraint elevates the yield strength to a value averaged thru the thickness of approximately $\sqrt{3}\sigma_{ys}$.

Irwin suggested that K would be a useful fracture toughness parameter when the plastic zone size, r_Y, was small compared with the crack size, a. This requirement allows the use of only the singular term of the linear elastic stress analysis to characterize the crack tip stress field. This concept was employed by ASTM to establish the empirical requirement for crack length for a plane strain K_{Ic} test as (9),

$$a \geq 2.5\,(6\pi)\,r_Y = 2.5\left(\frac{K_{Ic}}{\sigma_{ys}}\right)^2 \qquad (2)$$

In a similar manner, Irwin identified the mid-range of the "thickness fracture mode transition" as occurring when (21),

$$2\,r_Y = B \qquad (3)$$

The variation of (K_c/σ_{ys}) with the ratio $\left[\frac{B}{(K_c/\sigma_{ys})^2}\right]$ is shown in Fig. 6 for a steel and an aluminum alloy (26,21). Depending upon anisotropy, the elevation of fracture toughness, K_c above K_{Ic}, for thin sections ($4B \sim 2\,r_Y$), does not always reach the magnitudes shown in Fig. 6.

To achieve full transverse constraint the thickness, B, must exceed the value given by Eq. 3 by an amount that ASTM empirically established as (9),

$$B \geq 2.5\,(6\pi)\,r_Y = 2.5\left(\frac{K_{Ic}}{\sigma_{ys}}\right)^2 \qquad (4)$$

In Eqs. 2 and 4, r_Y is the formal plastic zone size associated with the average yield strength elevation caused by full constraint.

Krafft (28) and others (28,1,29) recognized the need for an additional dimensional parameter, associated intimately with the crack front fracture process and, therefore, much smaller than the plastic zone. Mechanics analysis (1) suggested that Well's parameter, the crack opening stretch, δ, defined for an elastic-perfectly plastic material as,

$$\delta = \alpha\,\frac{\mathcal{G}}{\sigma_Y} = \alpha\,\frac{K^2}{E\sigma_Y} \approx \alpha\,\frac{J}{\sigma_Y} \qquad (5)$$

as the appropriate parameter. Refined analysis suggests $\alpha \sim 0.5$, however, for the dimensional considerations employed here α will be taken as 1.0. Not only is δ representative of the separation of the crack faces close to the crack tip, it also characterizes the size of a region of "intense strain" directly in the path of the "blunted" crack (1). Because of the large in-plane strains in this region, there is a strong tendency toward thickness contraction. The surrounding material, experiencing smaller plastic strains, constrains the transverse contraction and elevates the mean normal stress (1).

For fully plastic plane strain behavior, slip line fields around cracks indicate very different mean normal stress elevations for different specimen configurations (30). Thus, even for plane strain conditions, as the mean normal stress changes, it is expected that a "mean normal stress parameter," in addition to a fracture toughness parameter, may be needed to adequately characterize the fracture toughness of a component (30,1). Data for a forging steel using thick center cracked tensile members and notched beams did not show evidence of this need for a second parameter: each specimen type fractured in a fully plastic condition at essentially the same value of J_{Ic} (14b). This question requires further clarification.

A qualitative pattern of fracture toughness behavior was suggested in Fig. 4. Using the size of the plastic zone, r_Y, and the specimen dimensions, the region of small scale yielding with full transverse constraint was delineated. An obvious next step is to delineate the other regions in Fig. 4.

A primary division in Fig. 4 is between two crack tip modes of deformation and separation, namely in-plane plastic deformation, Mode I, and 45° out-of-plane plastic deformation, combined Modes I and III. Important specimen dimensions are thickness, B, and unfractured ligament b. For notched beams and tensile specimens strained into the elastic-plastic and fully plastic region, the transition occurs in the range when $B \approx b$. In-plane plastic deformation and Mode I crack extension are promoted by $B > b$. Preliminary data for beam and tensile specimens suggests that $B = 2b$ is sufficient to lead to dominantly in-plane deformation and Mode I fracture behavior (25). Conversely, 45° out-of-plane plastic deformation and combined Modes I and III crack extension are promoted by $B < b$. To develop full slant fracture may require $3B \sim \Delta a$, where Δa is subcritical crack extension. Thus, to develop complete R curves in the fully plastic region generally requires that $B << b$ (15).

For in-plane elastic-plastic and fully plastic behavior, full transverse constraint is specified in terms of B and δ_{Ic}, the size of the crack tip region of intense strain. By analogy with the small scale yielding, Paris (13) estimated that full transverse constraint would be achieved in the elastic-plastic and fully plastic range when,

$$25\,\delta_{Ic} \le B \le 50\,\delta_{Ic} \qquad (6)$$

Work is underway to more fully validate this condition.

The measuring capacities for full transverse constraint of a fatigue cracked Charpy specimen ($a/w = 0.5$, $B/b = 2$) and a 1 T-CT specimen ($a/w = 0.5$, $B/b = 1$), in the small scale yielding and the fully plastic range are shown in Fig. 7. Clearly the measuring capacity for full transverse constraint has been markedly increased for these two specimens when fully plastic behavior is allowed and either δ_{Ic} or J_{Ic} is employed as a fracture criterion. For toughnesses higher than the measuring capacity shown in Fig. 7, these specimens give values of toughness that are somewhat too high. Note that $\alpha = 50$, was employed in Fig. 7; probably this is a conservative value of α. It would appear that one of these specimens would have the capacity to measure a fracture toughness sufficient to insure the safety of most structures even when low yield strength high toughness metals are employed.

Recent developments in Charpy-type specimen testing including (1) fatigue cracked specimens with deeper cracks ($a/w \longrightarrow 0.5$) and (2) instrumentation to record both load and deflection (time) (31,32), promise significant advances in quantitative measurements of Mode I full transverse constraint fracture toughness. It appears that a fatigue precracked Charpy-type specimen, scaled up by factors of 2 and 3, would cover the entire range of toughnesses of low strength high toughness metals. For measurements below and in the transition temperature range, the impact test will remain an essential tool for measuring toughness and "transition temperature" of ferritic-pearlitic and low strength tempered martensitic steels. For strain rate insensitive metals (nonferrous, austenitic stainless steels, precipitation hardening steels, etc.) and the ferritic-pearlitic and martensitic steels above their transition temperature, slow bend and tensile tests appear to be more appropriate.

In the 45° out-of-plane deformation region, Mode I plus III, with low transverse constraint, methods suitable for measuring R curves over the entire range from small scale yielding to fully plastic behavior have been developed to the point that extensive exploration of the R curves for a number of metals in different thicknesses is in order. In the fully plastic range, Heyer and McCabe have observed a limitation on the level of the plateau of the R curve that can be measured with a particular size specimen. Even when $b >> B$, the plateau level increases as b is increased (16). At present, R curve determination is reasonably sophisticated. Another purpose of exploration should be to develop simplified methods for R curve determination once the patterns of behavior are established.

From results in the small scale yielding range (Fig. 6), it is anticipated that low constraint fracture toughness will exhibit a peak in toughness over some range of thicknesses. For small scale yielding, this peak occurs at approximately $4B = 2r_Y$. The location and breadth of the peak should be studied in the elastic-plastic and fully plastic regions.

From a fracture toughness point of view, the thickness range corresponding to this peak toughness is the most efficient range of thickness in which to employ a particular metal. The thickness range of peak toughness may be a consideration in design, including selection of a particular metal as well as the thickness of the section.

Having separated toughness measurements into at least two categories, in-plane full transverse constraint Mode I separation, and 45° out-of-plane low transverse constraint Modes I plus III separation, it is clear that the structural application must guide the selection of the appropriate type of measurement. Regardless of the fracture toughness parameter that is selected (total energy to fracture, load and deflection from instrumented tests, R curves, δ_{Ic} or J_{Ic}), it is a forward step to proportion a specimen to "model" the gross deformation and fracture separation features anticipated in the machine or structural application. Further, as quantitative studies using sophisticated techniques (instrumented tests) and fracture toughness parameters (R curves, δ_{Ic}, J_{Ic}) are extended, a sound foundation is layed for the interpretation of simpler, less expensive quality control tests.

References

1. D. C. Drucker, J. R. Rice, Eng. Fracture Mech., V. 4, No. 4, April, 1970, pp. 577-602.
2. J. D. Lubahn, Fracturing of Metals, ASM, Cleveland, 1948, pp. 90-132.
3. V. Weiss, Mechanical Behavior of Materials, V. 1, Soc. of Mat. Sci., Japan, 1972.
4. D. P. Clausing, Inter. J. of Fracture Mech., V. 6, 1970, p. 71.
5. W. J. Hall, H. Kihara, W. Soete, A. A. Wells, Brittle Fracture of Welded Plate, Prentice-Hall, Inc., N. J., 1967.
6. W. D. Biggs, The Brittle Fracture of Steels, MacDonald and Evans, Ltd., London, 1960.
7. G. R. Irwin, Handbuch der Physik, Springer, V. VI, 1958, p. 551.
8. ASTM STP No. 381, 1965.
9. ASTM Designation: E 399-707, 1971, ASTM Standards.

10. G. R. Irwin, P. C. Paris, Fracture, V. III (Liebowitz, ed.) Academic Press, N. Y., 1971, pp. 1-46.
11. A. A. Wells, Welding Res. Abroad, V. X, No. 1, Jan., 1964, pp. 2-9.
12. J. R. Rice, Fracture, V. 2 (Liebowitz, ed.) Academic Press, N. Y., 1968, pp. 192-311.
13. P. C. Paris, discussion of ref. 14, ASTM STP 514, 1972.
14. (a) J. A. Begley, J. D. Landes, ASTM STP 514, 1972.
    (b) J. D. Landes, J. A. Begley, ASTM STP 514, 1972.
15. (a) R. H. Heyer, D. E. McCabe, Presented at 1969 Nat. Sym. on Fracture Mech.
    (b) R. H. Heyer, D. E. McCabe, Presented at 1970 Nat. Sym. on Fracture Mech. Both papers to be published in Eng. Fracture Mech.
16. R. H. Heyer and D. E. McCabe, Unpublished report, Armco Steel Corp., Sept., 1972.
17. W. S. Pellini and R. W. Judy, Jr., Naval Res. Lab. Report 7187, Oct., 1970.
18. S. T. Rolfe and S. R. Novak, ASTM STP 463, 1970, pp. 124-159.
19. F. J. Witt and T. R. Mager, Nuclear Eng. and Design, Vol. 17, No. 1, 1971, pp. 91-102.
20. T. E. Stanton and R. G. C. Batson, Proc. Inst. Civ. Eng., Vol. 211, 1921, pp. 67-100.
21. G. R. Irwin, J. Bas. Eng., Trans. ASME, V. 82, 1960, pp. 417-423.
22. M. H. Jones and W. F. Brown, ASTM STP 463, 1970, pp. 63-101.
23. S. Grigory, Nuclear Eng. and Design, V. 17, No. 1, 1971, pp. 161-169.
24. A. P. Green and B. B. Hundy, J. Mech. Phy. Solids, V. 4, 1956, pp. 128-144.
25. R. H. Sailors, unpublished work, Univ. of Illinois, 1972.
26. J. E. Srawley and C. D. Beachem, Naval Res. Lab. Report 5771, May, 1962.
27. G. R. Irwin, Eng. Fracture Mech., V. 1, 1968, pp. 241-257.
28. J. M. Krafft, Appl. Materials Res., V. 3, April, 1963.
29. V. Weiss, Fracture, V. 3 (Liebowitz, ed.) Academic Press, N. Y., 1971, pp. 227-264.
30. F. A. McClintock and G. R. Irwin, ASTM STP 381, 1965, pp. 84-113.
31. ASTM STP 466, 1970.
32. A. S. Tetelman and S. L. Server, U. C. L. A. Report Eng. - 7153, July, 1971.

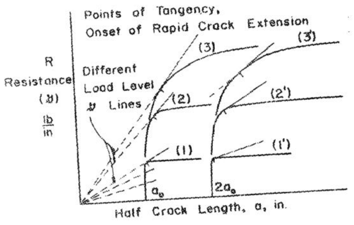


Fig. 1 Schematic R Curves

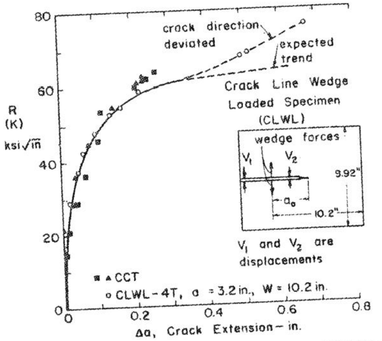


Fig. 2 R Curve for 7075-T6, Aluminum, 0.0625 in. Thick, RW Direction (15)

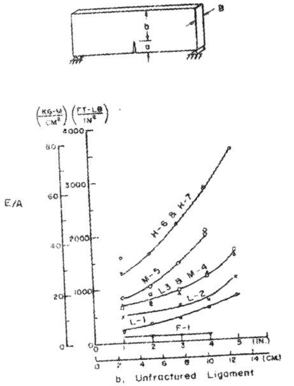
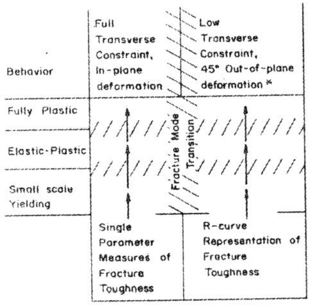


Fig. 3 Summary of E/A as a function of b which define the R-curves for Steels listed in Table I (17)

Steel	σ_{ys} ksi	σ_u ksi	RA %	El. %	Shelf Level DTE ft. lb.
F-1	120	149	30	12	300
L-1	162	185	48	14	1450
L-2	111	117	53	16	2100
L-3	182	201	66	18	2900
M-4	98	108	43	17	2720
M-5	125	141	61	19	4390
H-6	144	159	61	18	6650
H-7	83	98	68	24	6570



*See discussion of apparent exception for strain-rate sensitive metals in text.

Fig. 4 Schematic Summary of Fracture Behavior

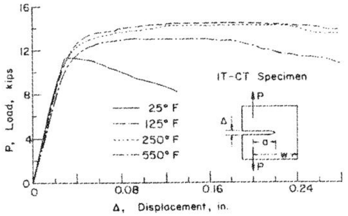


Fig. 5 Load-Deflection Diagram for IT-CT Specimens of A533B Steel at Four Temperatures (19)

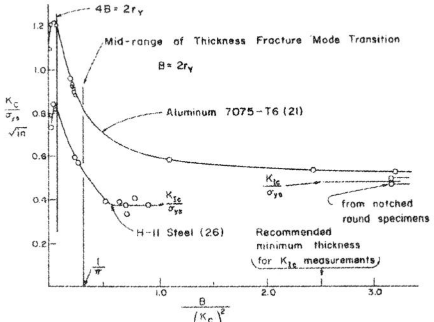


Fig. 6 Thickness Fracture Mode Transition for Small Scale Yielding Behavior

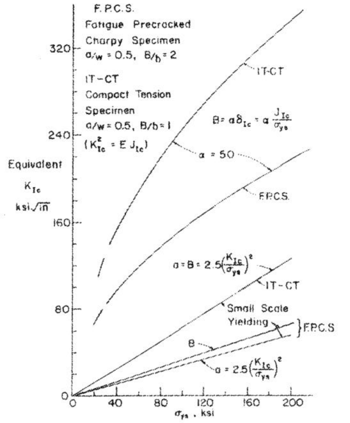


Fig. 7 Estimated Measuring Capacity for Full Transverse Constraint for Fatigue Precracked Charpy and Compact Tension Specimens.