Welding Research

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1971 ADAMS LECTURE

Principles of Fracture-Safe Design—Part II

The integration of metallurgical and mechanical parameters provided by the Ratio Analysis Diagram (RAD) defines conditions requiring the use of plane strain fracture mechanics in design, and conditions for which resistance to ductile fracture (R curves) is the approximate structural design reference

BY WILLIAM S. PELLINI

Editor's Note: Part I of the 1971 Adams Lecture appeared in the previous issue of the Welding Research Supplement. In his opening remarks the author pointed out that a weldment as a composite of three metallurgical entities—weld, heat-affected zone, and base metal—introduces complexities of fracture strength definition that demand the utmost sophistication in the understanding and the treatment of the fracture parameters. In this context the author then undertook to review the development of our knowledge of fracture-safe design from the early 1940s to the present. In doing so, he discussed:

- *Origins of the Problem and Early Studies.*
- *Evolution of Natural Crack Tests.*
- *Crack Arrest Tests.*
- *Fracture Analysis Diagram (FAD).*
- *Shelf Considerations.*
- *Fracture Mechanics Tests.*
- *Section Size Effects on the Transition Temperature Range.*
- *Expanded Version of FAD*

Part II begins below with a discussion of strength transition.

WILLIAM S. PELLINI is Superintendent, Metallurgy Division, Naval Research Laboratory, Washington, D.C.

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The Strength Transition

The strength transition is defined as the general effect of increased strength which results in decreasing the shelf level of high strength steels. The term signifies a shelf transition from high to low values due to increased strength. The strength range over which the shelf transition is evolved is related to the metallurgical quality of the steel, as will be described. There is a parallel to the temperature-induced transition in that both transitions in mechanical characteristics are controlled by metallurgical factors of microscale origin.

Increasing the yield strengths of steels to above 80 ksi (60 kg/mm^2) generally requires the addition of alloy elements (Ni, Cr, Mo, V, etc.) plus the use of quenching and tempering (Q&T) or other special heat treatments. These heat treatments result in fine dispersions of metallurgical phases (such as carbides) which are essential for raising the strength level above that of the as-rolled or normalized C-Mn mild steels. The fine structures also have important effects on the transition temperature range, involving potential displacements to low temperatures for optimum combinations of alloys and heat treatment. Physical metallurgy has been evolved to a high degree of scientific sophistication in procedures of microstructural optimizations for specific levels of yield strength and section size. All modern high strength alloys have been designed on these principles, and their fracture toughness characteristics are definitely not a matter of accident. Table 1 presents a highly simplified summary of the compositional aspects of typical, weldable, high strength steels with notations relating to processing factors to be discussed in the following sections.

Table 1—Weldable High Strength Steels

Type	Yield strength range, ksi (kg/mm²)	Section size limit,[a] in. (cm)	Heat treatment and melting practice[b]	Primary alloy elements, % C	Mn	Ni	Cr	Mo	Co	V	B	P and S impurity level, %
Commercial low alloy	90–125 (63– 88)	2 (5)	Q&T	0.15	1.0	1.0	0.50	0.50	—	0.05	0.005	0.03
			(Air)	0.15	1.5	—	—	0.50	—	—	0.005	0.03
Ni-Cr-Mo	80–100 (56– 70)	6 (152)	Q&T (Air)	0.15	0.30	3.0	1.5	0.50	—	—	—	0.02
	130–140 (91– 98)	6 (152)	Q&T (Air)	0.10	0.80	5.0	0.50	0.50	—	—	—	0.008
High Ni + Co	160–200 (112–141)	6 (152)	Q&T (VAR to VIM+VAR)	0.20	0.30	9.0	0.80	1.0	4.0	0.10	—	0.005
12-5-3 maraging	160–200 (112–141)	6+ (152+)	Q&T (VAR to VIM+VAR)	0.02	—	12.0	5.0	3.0	—	—	Ti & Al 0.10 to 0.30	0.005
18-8-4 maraging	200–240 (141–169)	6+ (152+)	Q&A (VAR to VIM+VAR)	0.02	—	18.0	—	4.0	8.0	—	0.10 to 0.30	0.005

[a] For optimized properties.
[b] Codes: Q&T—quench and temper; Q&A—quench and age; Air—air melting under slags; VAR—vacuum arc remelt; VIM—vacuum induction melting; +—for best properties.

Figure 26 illustrates the temperature displacements of DT energy transition curves for Q&T alloy steels of 110 ksi (77 kg/mm²) yield strength, as compared to a C-Mn steel of 48 ksi (34 kg/mm²) yield strength. This displacement in the transition temperature range is obtained with retention of high levels of shelf fracture toughness. With increasing yield strength to levels of 200 ksi (140 kg/mm²), there occurs a progressive decrease of the shelf level (strength transition), an increase in the transition temperature range, and finally the elimination of the temperature transition features. These effects may be noted in Fig. 26 from the DT test curves relating to steels of increasing yield strength levels. The low-level dashed curve labeled 110 ksi represents a commercial steel which has been optimized with respect to cost rather than maximum fracture toughness properties. Again, the properties are intentional and not accidental.

The general effects of increasing strength level on the temperature and strength transitions are presented schematically in the three-dimensional plot of Fig. 27. The vertical scale references the DT test energy. One of the horizontal axes defines the transition temperature range features, and the other the strength transition features. Two surfaces are indicated in this plot. The outer surface relates to the best (premium) quality steels that have been produced to date for the respective yield strength levels. The inner surface, which lies at generally lower DT test energy values, pertains to commercial products of lowest practical cost. The latter steels are produced with minimum (marginal) alloy content for the section size involved and are melted by conventional practices which result in relatively large amounts of nonmetallic inclusions. Both aspects evolve for reasons dictated by minimization of costs.

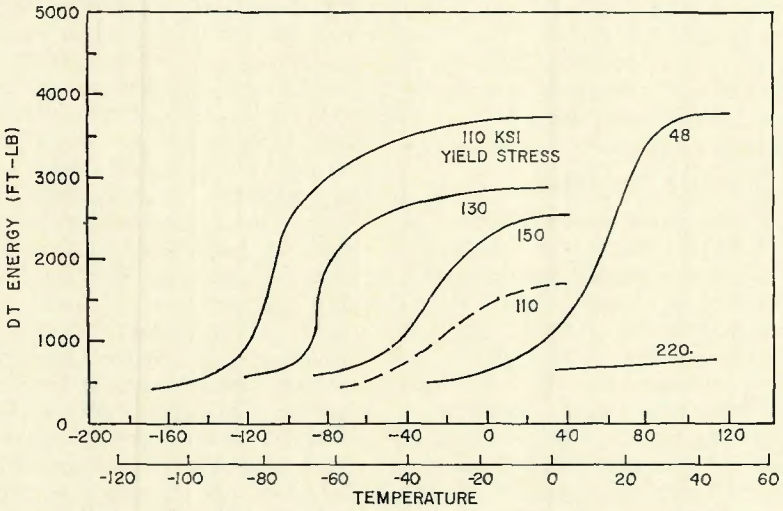


Fig. 26—Illustrating the general trends of increasing yield strength (solid curves) on the 1 in. DT test temperature transition range and on shelf level features. The effects of increasing void site density (see Fig. 17 in Part I) are indicated by comparison of the two curves for the 110 ksi steels. The solid curve relates to low void-site-density metal and the dashed curve to high void-site-density metal

Low-alloy contents restrict the nature of metallurgical transformations to less than optimum microstructural states and, therefore, result in higher transition temperature ranges as compared with the high-alloy grades of higher cost. The relatively low cleanliness of the low-cost commercial steels provides sites for void incubation which decreases the level of shelf fracture toughness. Accordingly, the strength transition, to the brittleness levels of plane strain fracture toughness, is attained at lower yield strengths as compared to premium products of high cleanliness. These combined effects are evident by comparison of the two steels of 110 ksi yield strength level, represented by the solid and dashed curves of Fig. 26. The low position of the inner surface (Fig. 27) for the commercial steels is dictated by these factors. It is important to note that the metallurgical factors, which control the temperature and strength transitions, are different and independent in major degree. The metallurgist must therefore consider a wide range of options in evolving steels which are optimized with respect to temperature and shelf transitions, as well as to strength level, section size, cost, weldability, etc.

Extensive studies of the characteristics of high strength steels were conducted by NRL during the period 1962 to 1968. These studies resulted in the first specific definition of the three-dimensional surface for the best steels, shown in Fig. 28. As such, it represents the limit of technological attainment with respect to both the temperature and strength transitions.

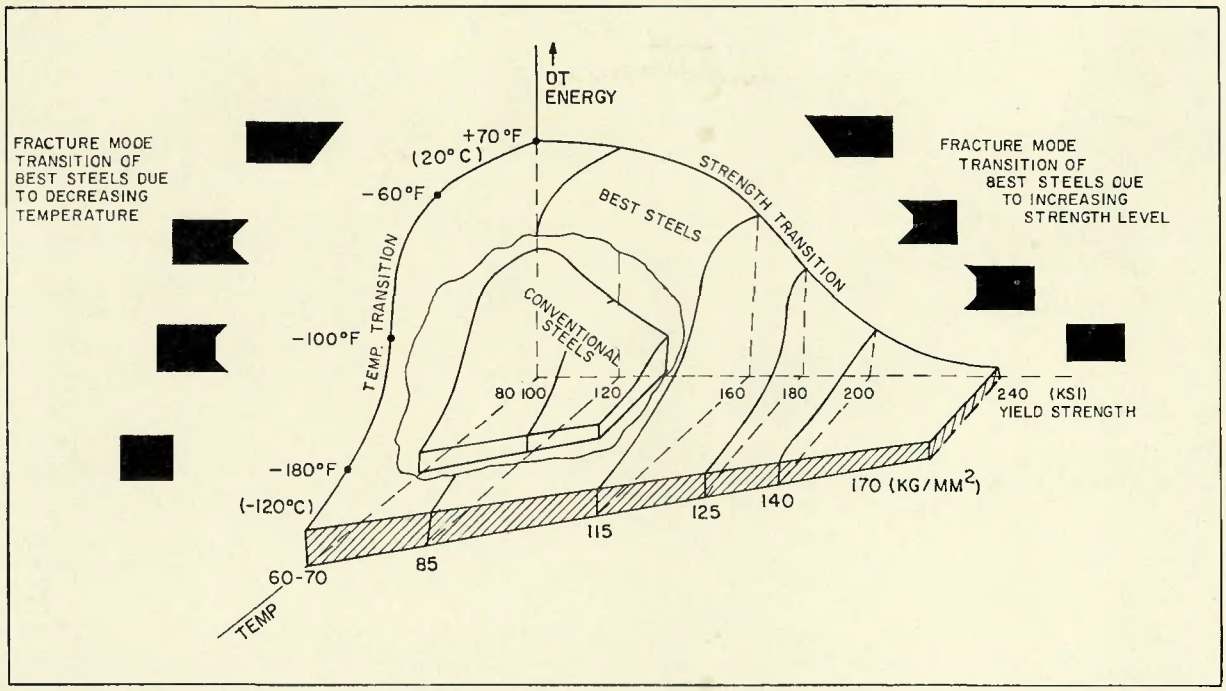


Fig. 27—Schematic illustration of three-dimensional DT test energy surface evolved by combined effects of temperature-and-strength-induced transitions. The nature of changes in fracture appearance for the 1 in. DT specimens are illustrated by the drawings. Transitions from ductile to brittle fracture are developed as a consequence of decreasing temperature or increasing strength level

The shaded region at the bottom of the surface follows the contour of the toe region of the DT test temperature transition range. Accordingly, it indexes the course of the temperature-induced plane strain regime for which fracture mechanics applies. Since the transition features were determined by the use of 1 in. DT test specimens, the plane strain limit relates to section of small size, as indicated by the notation "plane strain—small B." The temperature limits of plane strain for large-section size should be located at approximately 60 to 80° F (33 to 45° C) higher temperatures, if the metallurgical quality is retained for the large-section size.

The nature of the DT test specimen fracture in the plane strain region is flat and devoid of shear lips. The designation of "mixed mode *f*B" signifies that the nature of the fracture, in the region of the three-dimensional surface which is not shaded, involves flat central regions with surface shear lips. The relative fraction of flat and slant fracture is a function of thickness. The shaded region at the top of the surface indicates very high levels of fracture toughness and fractures of ductile (plastic) features.

The effects on fracture initiation and propagation characteristics, which result from following the downward course of the surface (as the result of increasing strength or decreasing temperature), may be visualized from the Cylindrical Bulge Explosion Test series shown in Fig. 29. These tests involved plates of 1 in. thickness featuring a 2 in. long, through-thickness, brittle, electron beam weld. A sharp crack of this dimension is thus introduced into the test plate as the metal approaches near-yield stresses in explosion loading. The capability of the metal for accepting plastic overload in the presence of this sharp crack is the feature of interest. The series represent typical results for tests conducted in the ductile, midtransition, and toe regions of the three-dimensional surface. An exactly similar change from ductile to brittle performance ensues as the result of decreasing temperature or increasing strength level.

These tests were used in early studies of the structural significance of the strength transition and for purposes of validating the inferences of the DT test. It was found that the fracture appearance of the explosion test fracture surfaces were reproduced exactly by the DT test. In other words, the dynamic mechanical severity of the DT test was equal to the extreme (ultimate) conditions of the explosion test.

A cut through the three-dimensional diagram at 70° F (20° C) should indicate the effects of increasing yield strength on the shelf

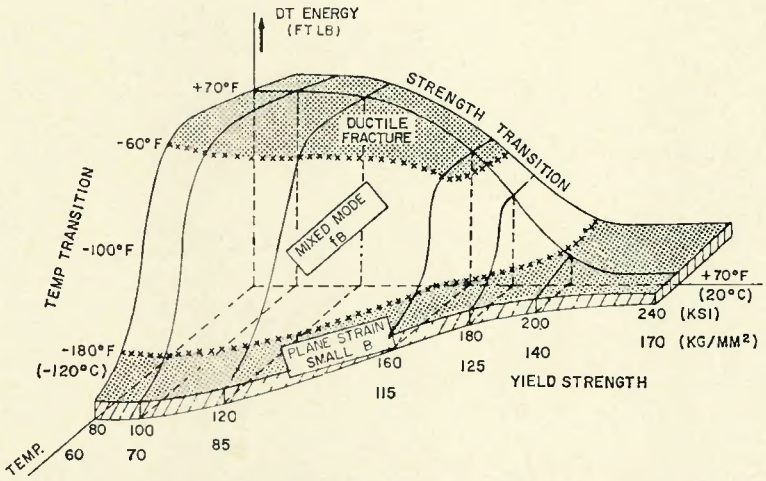


Fig. 28—Three dimensional DT test energy surface for best (premium) steels. For definition of the specific DT test shelf scale values, refer to the curve labeled 1968 Technological Limit, Fig. 30

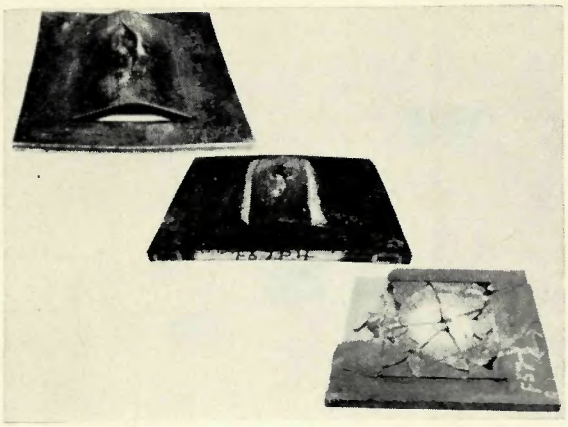


Fig. 29—Cylindrical Bulge Explosion Test series illustrating the structural significance of temperature and strength transitions from high to low positions of the three-dimensional surface of Fig. 28

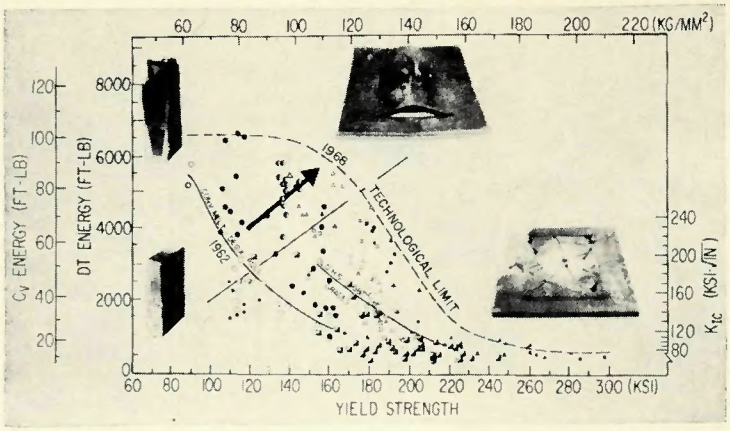


Fig. 30—Summary of DT test shelf energy data resulting from extensive survey of all principal alloy types and production methods. The significance of the DT test energy scale for the 1 in. specimen is indexed to the change in fracture appearance of the DT test (brittle to ductile fracture) and also by the equivalent change denoted by the explosion tests. The C_v shelf energy scale and K_{Ic} scale are indexed by correlation

level fracture characteristics. This follows because all of these steels have attained shelf levels at this reference temperature. Figure 30 presents a summary of shelf level data points that were obtained for different steels as a function of increasing yield strength. These data points represent the *shelf level* fracture energy of the 1 in. DT specimen. The test orientation is uniformly in the weak direction for plates that do not feature 1:1 cross rolling. This procedure avoids placing a premium on obtaining high transverse (strong direction) fracture toughness at the expense of the other direction. Thus, it clearly indicates the basic effects of metallurgical features relating to void site density. For engineering purposes, the fracture test direction should always be taken in the direction of anticipated fracture propagation in the structure. This will result in relocation of the data points in the diagram, with *no change* in the methods of analysis to be described.

There is no implication that tests should always be made in the weak direction—the choice is a matter of engineering decision. From a fracture analysis point of view it is equally wrong to test in the strong or weak directions if the fracture path is deduced to be in other than the test direction. It should be noted that steels of high-cleanliness quality, which are required to maximize fracture toughness at high strength levels, show very little directionality, even in the absence of 1:1 cross rolling.

The C_v shelf level scale has been added by correlation and serves to index the strength transition significance of the C_v shelf value. It is emphasized that the correlation between DT and C_v shelf characteristics does not imply that both tests attain shelf characteristics at the same temperature. In general, the C_v test attains shelf levels at lower temperatures than the DT test. In brief, the C_v test does not represent the true location of the temperature transition range and, thereby, poses major difficulties of interpretations. The function of the K_{Ic} scale is discussed later.

The steels available in the early phases of these studies fell under the two lower curves noted as the Technological Limit of 1962. The outer bound Technological Limit curve, noted as 1968, represents the metallurgical improvement obtained by the steel industry during this period, as highlighted by the bold arrow designation.

As presented in Fig. 30 the Technological Limit Diagram is only useful in presenting a panorama of general trends in shelf level fracture toughness. The real need was to evolve a system for indexing significant mechanical parameters to this panorama which can be translated to structural performance. The course of evolution of the desired indexing system began with recognition that the low energy values of the DT test related to brittle fracture, and the high values relate to highly ductile fracture, as illustrated by the DT test specimens in the figure.

The next observation is provided by the photographs of the explosion test samples which correspond to the high and low positions in the diagram. Based on such observations the slant line drawn across the diagram serves to separate the diagram region of brittle, unstable, elastic fracture from the region of highly ductile fractures for steels in the order of 1–2 in. thickness. Clearly, only the region

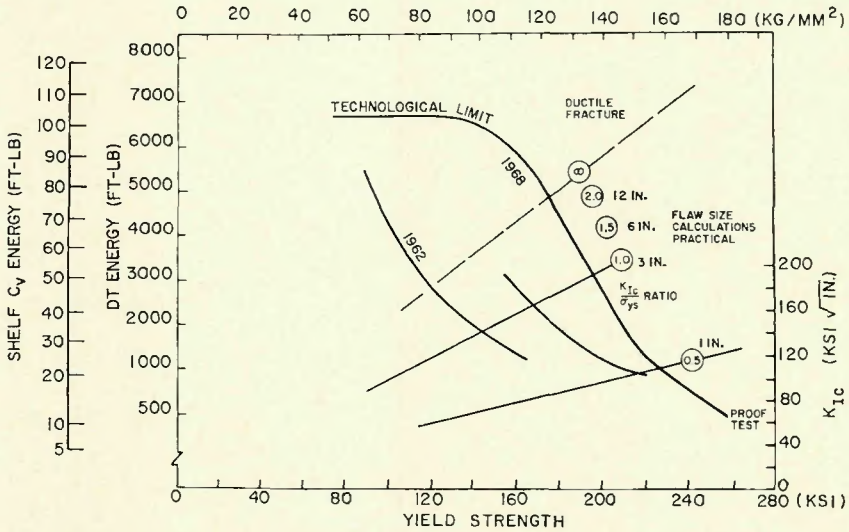


Fig. 31—Ratio Analysis Diagram (RAD). The grid of K_{Ic}/σ_{ys} ratio lines indexes the region of the diagram which relates to a plane strain fracture. The dimensions of critical flaw sizes, which relate to the ratio values, are defined in Fig. 21

which lies considerably below the slant line is potentially indexable to the K_{Ic} plane strain fracture mechanics parameter.

Ratio Analysis Diagram (RAD)

By 1968 the indexing of the plane strain fracture region of the strength transition diagram was achieved and the K_{Ic} scale was added. This accomplishment was of major importance because the diagram could now be zoned to represent regions which related to specific K_{Ic}/σ_{ys} ratios. Figure 31 presents the final version, defined as the Ratio Analysis Diagram (RAD). It should be noted that the diagram is expanded to include low values of K_{Ic}. The 0.5 and 1.0 ratio lines are located accurately in the expanded diagram because valid K_{Ic} data were obtained for this level of fracture toughness.

The dashed line noted as the infinity (∞) ratio represents the best estimate that can be made at this time of the limit to which unstable plane strain fracture toughness can be measured by K_{Ic} tests of large section size. These measurements would relate to K_{Ic}/σ_{ys} ratio 2.0 value or higher, and thus, would require specimens of section size in excess of 10 in. The high metallurgical ductility of steels with C_v and 1 in. DT test shelf energy values in excess of this line is amply demonstrated by highly ductile fracture for section sizes in the order of 3 to 4 in. thickness. There is no basis for expectations of developing plane strain fracture (brittleness) as the result of additional increase of section size for such highly ductile metals. Confirming evidence that the ∞ line location is correct is evolving from large scale K_{Ic} tests.

The 1.5 and 2.0 ratio lines must lie in the narrow gap remaining between the ∞ and the 1.0 ratio lines. There is no attempt to define the location of these lines because the significance of K_{Ic} values for ratios in excess of 1.0 is subject to question. These involve questions of the significance of K_{Ic} instabilities which are followed by rising load, as well as the procedures for plastic zone corrections of K_{Ic} values. Thick-section K_{Ic} tests will be required to better define the area of uncertainty between the 1.0 ratio level and the general location of the ∞ ratio estimate. Accordingly, the K_{Ic} scale has proper meaning to about the general level of the 1.0 ratio line or slightly higher; it should not be used for design purposes at higher levels at the present state of knowledge. The section sizes required for improved definitions of the K_{Ic} scale are indicated by the thickness notations associated with the ratio values. In summary, these relationships indicate that as the valid 1.0 ratio correlation to DT energy is exceeded, there is a rapid increase in K_{Ic}/σ_{ys} ratio values. This is generally similar to the rapid rise in static and dynamic ratios as the DT energy curve rises above the lower toe region for the transition temperature case.

It should be noted that a thickness significantly less than the requirement for the ratio value would result in plane stress fracture due to inadequate constraint. For example, plates of 1 in. thickness are plane strain limited to approximately a 0.63 ratio value. If the intrinsic fracture toughness of the metal significantly exceeds the 0.63 ratio value, it is not possible to enforce a plane strain condition on the 1 in. plate section, i.e., plane strain is unattainable for this section size.

The RAD may be separated into three general regions. The *top region*, above the ∞ ratio line, relates to

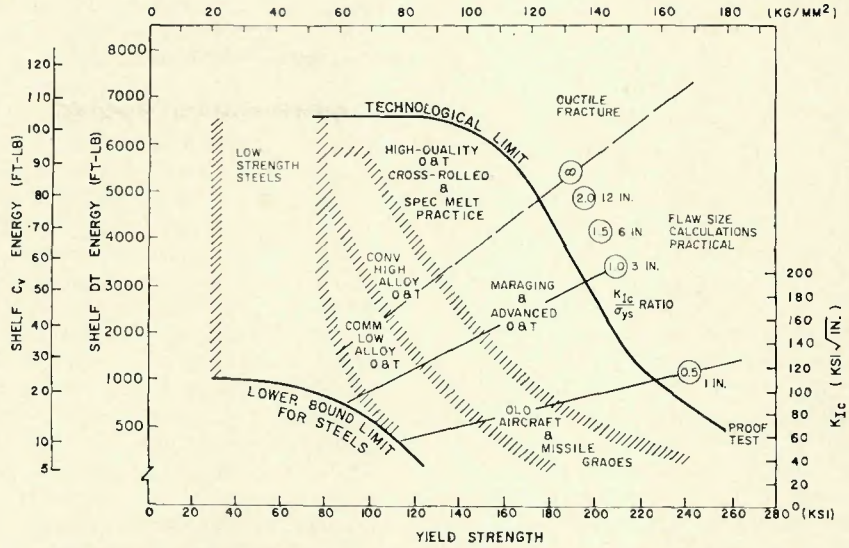


Fig. 32—Metallurgical zoning of the RAD which defines the general effects of melting and processing factors on the strength transition. The three corridors of strength transition relate to metallurgical quality (void site density) which controls microfracture processes and, thereby, the macroscopic fracture toughness of the metal. The locations of generic alloy steel types are indicated by the notations

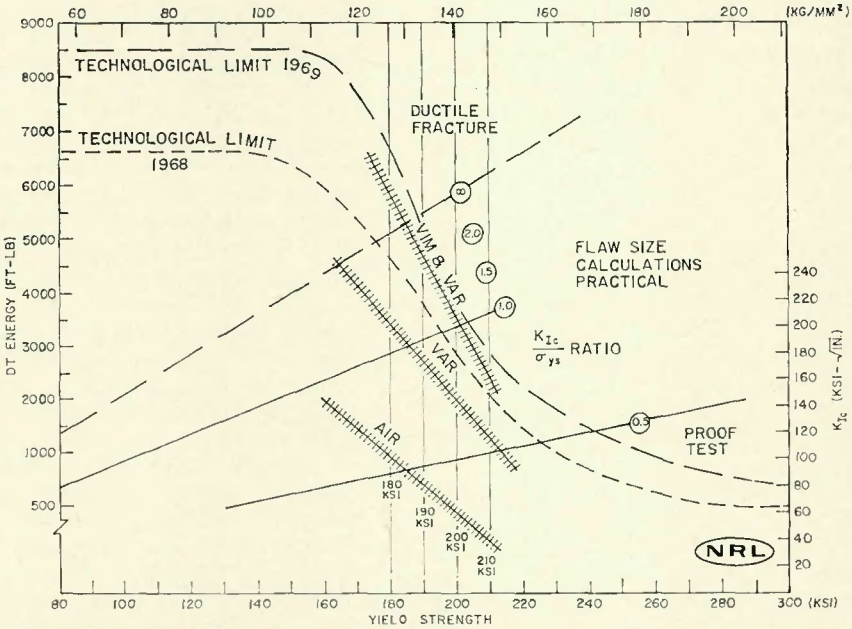


Fig. 33—Metallurgical zoning of the RAD which defines the effects of specific melting practices. These include air melting under slags (Air), vacuum induction melting (VIM) and vacuum arc remelting (VAR); a combination of these is indicated by the VIM-VAR notation. The extension of the Technological Limit curve to the 1969 position results from the use of advanced VIM-VAR melting practices at large production scale

ductile fracture. The *bottom region* below the 0.5 ratio line relates to low levels of plane strain fracture toughness, involving flaw sizes which are too small for reliable inspection. This can be deduced by returning to Fig. 21 and noting the flaw sizes associated with ratios below 0.5. These are generally in the order of tenths of inches for stubby flaws and decrease to hundredths of inches for long thin flaws subjected to high elastic stress levels. While fracture mechanics applies well in principle for this region, it applies poorly in practice due to the flaw detection problem. This region requires proof-test procedures for ascertaining preservice structural integrity.

The width of the remaining "slice" of the RAD, the *central region* for which plane strain fracture mechanics calculations appear feasible, depends on the section size. For very thick sections (to over 10 in.) the applicable region available for possible plane strain calculations is between the 0.5 and ∞ ratio lines. For section sizes in the order of 0.6 to 2.5 in. thickness, this region is bounded by the 0.5 and the 1.0 ratio lines. This range represents a very thin slice through the population of the steels represented by the diagram.

Factors which relate to the metallurgical quality of the steel become readily evident by metallurgical zoning of the RAD. The full span of fracture toughness definitions (plane strain to plastic fracture), provided by the DT test specimen energy scale, and the ratio lines definitions for the regions of plane strain fracture, serve as a fixed mechanical reference grid. The analytical value of this grid is exploited by the superimposition of zone reference systems which partition the diagram in relation to metallurgical features.

A simple, yet highly significant, zoning of metallurgical type is developed in Fig. 32 by tracing the effects of increasing strength level on the shelf transition features of various generic classes of steels. A series of metallurgical quality corridor zones, which are related to the melting and processing practices used to produce the steels, become evident. The lowest corridor zone involves relatively low-alloy commercial Q&T steels produced by conventional low-cost melting practices. The corridor is defined by the strength transition of these steels to the 0.5 ratio level, as the result of heat treatment to yield strength levels in the order of 130 to 150 ksi (92 to 105 kg/mm^2). Optimizations of the alloy content, coupled with improved melting and processing practices, elevate the corridors to higher levels. The strength transition to the 0.5 ratio is shifted accordingly to higher levels of yield strengths.

Recent metallurgical investigations of high strength steels have emphasized processing and metal purity aspects rather than purely physical metallurgy considerations of transformations. New scientific knowledge of void growth mechanisms and of the importance of void-site-density factors clearly indicates that control of metal bridge ductility can only be effective within limits. As these limits are reached by optimization of microstructures, it then is essential to obtain further increases in ductility by suppression of void initiation aspects. Void sites are provided by the presence of nonmetallic particles of microscopic size featuring noncoherent phase boundaries with the metal grains. Compared to coherent metal grain boundaries, the noncoherent phase boundaries separate easily. The extent to which such void sites can be eliminated is related to a number of factors including melting and deoxidation practices as well as the P and S impurity contents.

Figure 33 illustrates the effects of specific melting practices which relate directly to void-site-density. Air melting under conventional slags, followed by normal deoxidation practices, results in steels which have numerous nonmetallic particles, as is easily seen with low-power optical microscopes. The carbon deoxidized, vacuum induction—vacuum arc remelted (VIM-VAR) metal requires high-magnification electron microscope examination in order to locate the few nonmetallic particles that are present. Clearly, a new era in steel making has evolved which involves producing ultrahigh-purity steels at large production scale. Figure 33 also notes a relocation of the Technological Limit curve of the RAD to the 1969 position which has resulted from these metal processing improvements by the steel industry.

The RAD corridor relationships provide a vastly improved definition of metallurgical factors. Systematization of vast amount of data becomes feasible in easily understood form. The required directions for metal processing improvements are predictable from these corridor relationships. Moreover, it becomes apparent that the increased costs of high-quality processing are economically defensible, as well as technologically essential, at high strength levels. Analyses which combine economic, metallurgical, and fracture considerations over the full span of strength levels are of major interest to both users and producers of the steels.

Applicability of Fracture Mechanics to the Strength Transition

The foregoing discussion of the RAD has emphasized that a choice can be made between steels of brittle or ductile shelf characteristics over a wide range of yield strength levels. The integration of metallurgical and mechanical parameters provided by the RAD serves to define conditions which require the direct use of plane strain fracture mechanics in design, as well as those which preclude its use. These definitions are most important to engineering applications of fracture-safe design.

The following major points should be noted in these respects:

1. The combination of metallurgical quality and yield strength features that provide for practical use of plane strain fracture mechanics is relatively restricted. A modest decrease in yield strength, or an economically feasible increase in metallurgical quality, may elevate the metal from plane strain fracture levels and thereby ensure that plane stress (plastic) fracture controls.

2. The very low levels of plane strain fracture toughness below the 0.5 ratio severely restrict the practical engineering use of metals because of inspection limitations. The only practical inspection procedure for this level of plane strain fracture toughness is the proof test method which defines whether or not critical flaws existed by failure of the structure. However, no warning is provided that flaws of near-critical size exist. A slight increase in such flaw size by fatigue or stress corrosion may then occur in service, leading to unexpected proof of attainment of the critical flaw sizes.

3. If the structures involve relatively large plate thickness and feature relatively simple geometric configurations, flaw size-stress calculations may be feasible for metals with plane strain properties in the range of approximately 0.8 to 2.0 ratios. The flaw sizes involved will generally be within limits which permit reasonably reliable nondestructive inspection. The engineering reliability of this approach increases markedly with an increase in the ratio value.

4. For most engineering structures of complex design, fracture safety is best assured by choice of metals featuring plane stress fracture toughness. It is emphasized that the ratio value above which the plane stress (ductile) conditions apply is a function of the section size.

Fracture-safe design which is predi-

cated on the inherent plane stress fracture properties for the metal precludes the application of plane strain fracture mechanics. Thus, the design field should not insist on the characterization of plane stress metals by pseudo K_{Ic} values of the "lower bound" type. Because of the unknowing pressure for developing K_{Ic} values, a surprisingly large number of K_{Ic} tests are conducted for purposes of defining the "lower bound" K_{Ic} values. These values have no meaning except that a measurement of plane strain fracture toughness could not be accomplished, and therefore the K_{Ic} value, if attainable, must lie above the plane strain measurement capacity of the specimen or the section size in question. This practice is introducing confusion in the technical literature, particularly as these values are then used in design or in designating steel quality without reference to being "lower bound."

In conclusion, the use of plane strain fracture mechanics principles for fracture-safe design in relation to the strength transition is required only for a relatively narrow range of steels, in comparison to the total population evolved by the metallurgist. Except for the strength range in excess of 190 ksi (135 kg/mm^2) yield strength, the engineer has the choice of fracture-safe design based on using inherently ductile metals.

Welds and Weldability Factors

The early observation that ship failures never followed weld paths was explained by their generally lower transition temperature range in comparison to the steel plates. The C_v 15 ft-lb transition temperatures and the NDT temperatures of the welds were found to be below the temperatures of ship fractures. In general, the development of weld metals for low strength metals with equal or better transition temperatures than plates or forgings is not a difficult metallurgical task.

While the heat-affected zone of mild steels may be moderately embrittled compared to the base metal, the geometry of the heat-affected zone exerts a strong protective influence. Welds are normally of V or VV geometry and the heat-affected zone is slanted 45 to 60 deg with respect to the usual stress vector directions which parallel the base metal. The slant feature results in difficult propagation of brittle fractures, which tend to have normal incidence (90 deg) to the stress direction. Fractures which run in the heat-affected zone of mild steels are a rarity for the combined reasons of protective geometry and moderate embrittlement.

Retention of fracture toughness characteristics in welds and heat-affected zone which match plate properties becomes metallurgically more difficult with increase in strength level. The general problem is that optimum combinations of strength-fracture toughness characteristics are obtained by careful selection of heat treatment cycles, including austenitizing temperature, cooling rate, tempering temperature, and tempering time. These aspects are exactly controlled in the mill heat treatment of the base metal. In order to attain reasonably equivalent mechanical properties for the weld metal and heat-affected zone, the heating and cooling cycles must be controlled within reasonable limits by the manipulation of heat inputs, melting rates, number and size of beads, etc. For other reasons, which relate to cracking tendencies in cooling of the weld, it may be necessary to adjust the weld metal composition to different alloy contents than for the plate.

With increasing strength level, the weld metal becomes the most difficult aspect of the problem of developing weldable steels. For steels in the yield strength range of 130 to 180 ksi (90 to 125 kg/mm^2) the research effort expended for the weld metal development ordinarily exceeds that for the base material by several fold. Nevertheless, remarkable advances have been made in interrelating these complex factors. An example is provided by the weld metal zoning of the RAD presented in Fig. 34. This presentation highlights the strong effects of metal quality factors as related to the welding procedure. For attainment of high corridor features it is essential to use weld wire of high metallurgical quality (equal to the low-void-site density of high corridor base metals) and then to protect the weld metal pool by the use of inert gas shielding. The gas tungsten-arc method is superior to the gas metal-arc method in these respects. Because of poor protection to atmospheric gases, shielded metal arc welds are limited to the low corridor irrespective of improvements in wire quality. Adjusting weld alloy compositions to control microstructures is not sufficient—the metal quality aspects related to void growth (cleanliness) must be controlled also if high corridor characteristics are required.

The heat-affected zone of high strength steels may pose difficult metallurgical problems depending on the alloy content and metal quality. For the high alloy base materials of high RAD corridor characteristics, the heat-affected zone problem is minimized and excellent properties are obtained with latitude in the control of welding parameters. The high alloy contents promote the development of optimum metallurgical structures over fairly wide ranges of weld heat treatment variables.

For the low alloy commercial steels of low-corridor characteristics, the heat-affected zone problem becomes more difficult to resolve. This results from the marginal alloy contents which are barely adequate for developing desirable microstructures under controlled mill heat treatment conditions. Thus, the heat-affected zone for such metals may be seriously degraded by off-optimum welding procedures. The heat-affected zone

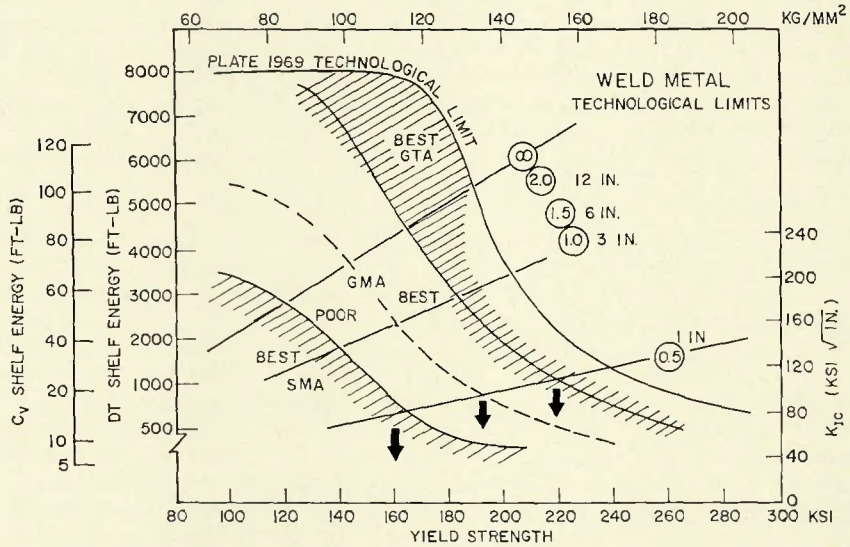


Fig. 34—Zoning of the RAD which defines weld metal factors of first order importance to strength transition features. The strength transition corridors relate to weld metal deposited by shielded metal-arc (SMA), gas metal-arc (GMA), and gas tungsten-arc (GTA). The bold arrows indicate the strength level of transition to the 0.5 ratio value

shelf characteristics for such steels in the range of 90 to 120 ksi yield strength (63 to 85 kg/mm^2) may then drop to very low levels. As a result, preferential fracture in the heat-affected zone becomes the expected fracture mode. Heat-affected zone failures of this type have been noted for pressure vessels. Such fractures are not affected significantly by wide variations in service temperature because they are related to the shelf transition and do not involve transition temperature considerations.

The evolution of practical laboratory test procedures for characterization of heat-affected zone fracture properties has been complicated by the effects of weld geometry. A promising solution to this problem has been attained recently by McGeady. The basis of the solution involves the separation of metallurgical and geometrical factors, i.e., each aspect is investigated separately. The intrinsic fracture toughness characteristics of the heat-affected zone is measured by the use of a weld geometry which provides a zone (on one side) that is exactly normal (90 deg) to the plate surface. The notch of the DT test is located in this zone and the fracture path is then contained entirely in the heat-affected zone. The DT energy reading indicates the effects of welding on heat-affected zone properties by comparison to the plate properties. Welding procedures which cause the least loss of fracture resistance in the heat-affected zone can be identified for a specific steel. Comparisons may then be made between different steels. This procedure provides a practical laboratory test method for the definition of the optimum heat-affected zone quality that can be developed for various steels. Effects of weld geometry can then be examined separately by the Delta Test (see references) or by the Explosion Bulge Test method.

The importance of this development is that the relatively large size weldment tests can then be used in their proper role as "prototype structural tests," i.e., to simulate the behavior of a part of the structure. As should be expected, the optimum heat-affected zone condition defined by the DT test results in the best performance when weldments are subjected to the Delta and Explosion tests. In effect, the large scale tests validate the implications of the DT-heat-affected zone test results. With the evolution of confirming data of this type for a wide variety of steels, the need for the large scale weldability tests should be eliminated in the near future. The best overall performance of the weldment can be guaranteed by separate, prior optimization of plate, weld and heat-affected zone properties using the DT test.

New Issues in Fracture Research

There are broad varieties of metal type–structure type combinations for which the bases of fracture-safe design lie squarely in the conditions for fracture *initiation*. Accordingly, the test specimens must model conditions which relate to initiation of fracture. There are equally broad varieties of metal type-structure type combinations for which the bases of fracture-safe design must be found in the conditions for *extension* of the fracture. For these cases, the specimen modeling must relate to the extension aspects.

The fracture research activity of the past twenty years has been dedicated almost totally to questions of fracture initiation for relatively brittle metals. Since the problems of fracture extension have been largely neglected, very few principles exist for engineering guidance. Attention should now be given to these unresolved problems, because they are rapidly becoming of major engineering consequence.

A large and growing fraction of the structural metals in engineering use may feature low or intermediate levels of resistance to the propagation of plane stress (non-brittle) fracture. These features may also be described as "low shelf" or "low tearing energy"—all terms have the same meaning. Since the low-shelf metals are neither brittle nor highly ductile, the engineer is confronted by an "in-between" state of semi-ductile fracture. The basic question for the semi-ductile state is that of fracture extension by low-energy tearing. Whether or not fracture will result by the extension of a low-energy tear requires reasonably refined consideration of the type of structure and levels of stress imposed. There are two basic types of structures—rigid and complaint. Rigid structures have limited capability for release of elastic strain energy, generally sufficient only for fracture extension involving relatively brittle metals. Compliant structures may feature high total energy, sufficient to cause fracture extension for relatively ductile metals.

The low strength steels generally feature high levels of tear resistance, except for unusual conditions of excessively preferential alignment of inclusions (weak directions of rolled or forged products). The intermediate strength steels may feature low levels of tear resistance and, therefore, present a special case of major engineering significance. The immediate importance of the intermediate-strength steel case evolves from prior experience with low strength steels. This experience has led to incorrect assumptions that restricting service temperatures to above the cleavage-to-ductile fracture transition provides positive assurance of fracture-safe performance. Unfortunately, this is not always the case for these steels, and structural failures have been experienced involving fractures which "do not show cleavage." Thus, the unexpected structural failures have been the cause of alarm and confusion. Actually, such failures should be expected, if the basic factors involved are understood in terms of metal type-structure type relationships.

The new problems result from four interrelated factors:

1. The increasing interest of designers in the use of intermediate strength steels of 70 to 130 ksi yield strength (50 to 90 kg/mm^2).
2. The economic competition to supply these steels with minimum cost increase over that of the conventional low strength steels. This requirement tends to promote the use of production practices which result in low-shelf (low tearing energy) type steels.
3. The relatively high stresses which are applied to these steels. This factor results in very high stress intensities at flaw tips, which induce crack initiation.
4. The extensive utilization of these steels in structures of high load-compliance-features. As the result, the mechanical force for continued tear extension may increase rapidly with enlargement of the tear.

There is no intent to present a case that only high shelf materials should be used. The shelf level requirements are dictated by the type of structure. The important point is that fracture research should be directed to the characterization of fracture extension factors. In general, the problem requires engineering definition of the lowest level of tearing energy that should be used for particular applications. Test methods, which provide a significant measure of the resistance to fracture extension, are required.

The nomenclature of crack extension, as contrasted to tear extension, requires explanation because the terms are often used interchangeably. A mechanically correct definition should recognize crack extension as fracture with retention of the sharp features of the crack tip. Fracture extension involving appreciable blunting of the crack tip is best described as tear extension. Thus, brittle fractures are of crack extension type and the semi-ductile or ductile fractures

are of tear extension type.

The *R* Curve Expression of Fracture Extension Resistance

Fracture mechanics research in the early 1960s was directed to measurement of the fracture extension resistance of metals in terms of the plastic work energy parameter $\mathcal{G}_c$. The fact that the expression is valid only for conditions of nominal elastic loads, and other reasons involving experimental complexities, led to an early end of this effort. The feature of "rising $\mathcal{G}$" in the fracture extension of metals was defined as the resistance curve (*R* curve). The prior fracture research of the 1950s recognized that the resistance to fracture increased with extension from a crack tip for the case of ductile metals. However, the metals of interest at the time were low strength steels of high-shelf (high-ductility) type and quantitative investigations were not deemed necessary.

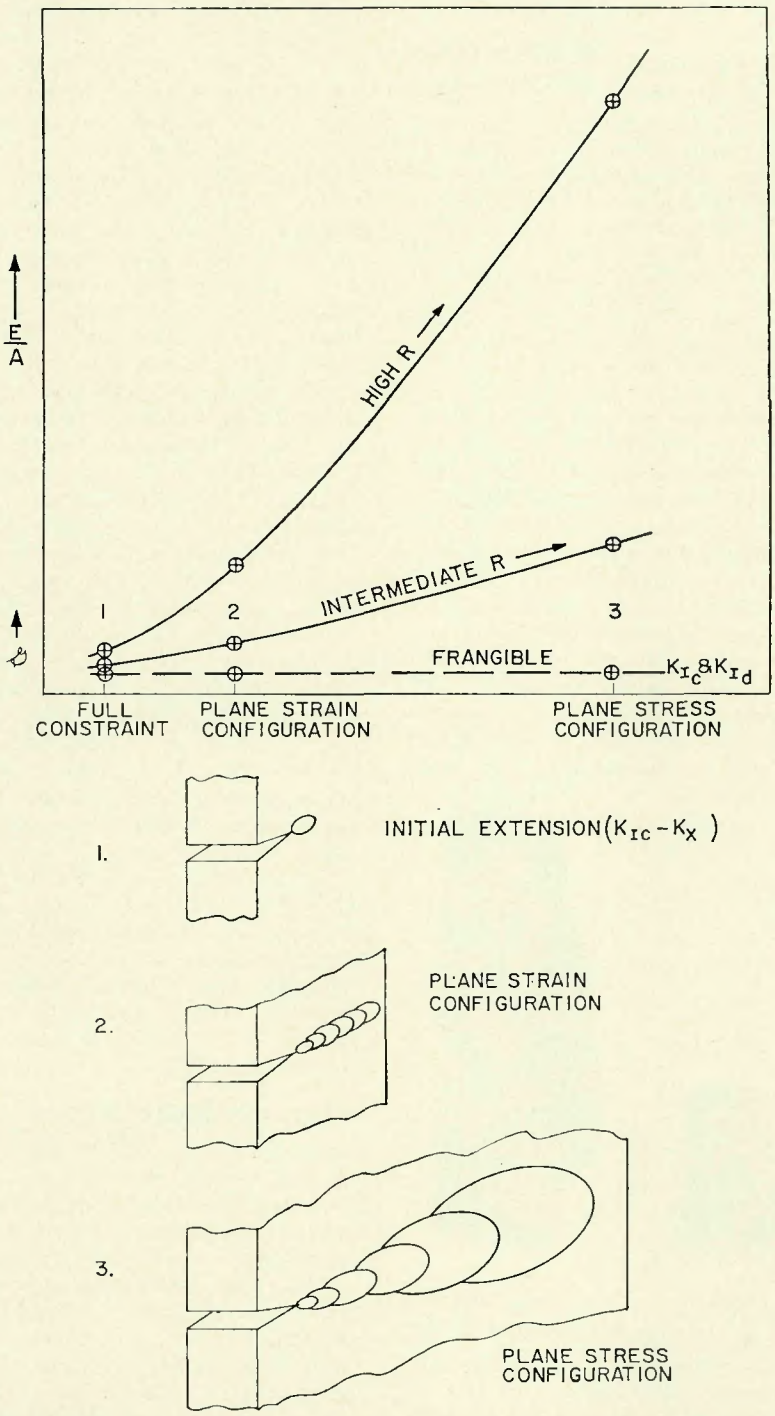


Fig. 35—Relationship of R curve slopes to the fracture extension energy measured by various tests, including (1) fracture mechanics tests; (2) the plane strain configuration DT test; and (3) the standard DT test which represents a plane stress configuration. The plastic zone growth conditions which determine the energy measurements for the three tests are illustrated for the case of highly ductile metal (high R)

Fracture extension resistance characteristics are best considered in terms of separation of metals into two broad classes—defined as frangible (unstable, brittle extension) and nonfrangible (stable, ductile extension). A most important aspect of nonfrangible metals is that the fracture extension resistance increases (to a limiting characteristic value) with movement away from the initial crack tip. For frangible metals, the fracture extension resistance remains at the same level or actually decreases (for rate sensitive steels). Thus, the frangible metals feature a "flat," nonrising, *R* curve. The nonfrangible metals feature rising *R* curves, ranging from a shallow to a very steep rise, depending on the intrinsic metallurgical ductility.

The rate of rise of the *R* curve is related to the degree of crack blunting which is developed in the process of fracture extension, as follows:

Nonrising R curve—crack tip sharpness is retained because the critical plastic zone size is small.

Shallow-rising R curve—crack tip blunting occurs gradually by a process of increasing plastic zone size. The small degree of blunting, developed in the first unit extension, results in forming a larger plastic zone size for the next step, which causes additional blunting and, therefore, an additional growth of the plastic zone size, etc. A stable condition is then achieved, following which there is no further increase in crack blunting or plastic zone size.

Steeply-rising R curve—crack tip blunting occurs rapidly as the result of gross plasticizing of the zone in advance of the tip, starting with the first rupture increment. The effects are then accentuated in further extension; again there is a leveling-out to a stable condition.

Increases in plastic work energy per unit fracture area extension (E/A) will evolve due to the crack blunting and plastic zone size increases. The magnitude of the E/A increases in the course of fracture extension for ductile metals is related to the rate of rise of the *R* curves.

The rate of rise of the *R* curves will be reflected in the nominal engineering stress required for fracture extension. For steeply-rising *R* curves, the nominal stress required for the initial tear extension will exceed yield and then require increase of the plastic load stress for continued extension.

There are no fracture-mechanics type tests which can define *R* curve

factors that involve plastic stresses. In fact, there is no basic theory which would provide for analytical procedures or guidance in evolving such tests. Since fracture extension resistance is a matter of energy absorption, tests of energy measurement type are ideally suited for the purpose. A practical approach to the definition of R curve features has been evolved by adjustment of the configuration of the DT test specimen.

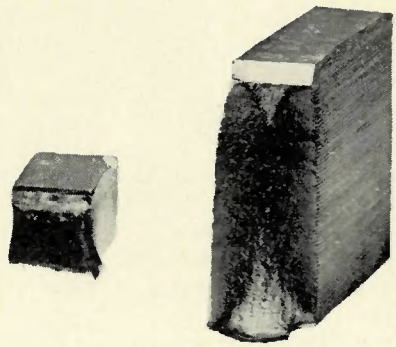
DT Test Definition of *R* Curve Slopes

The unique feature of the standard DT test is that it provides for measurement of either plane strain (brittle) or plane stress (ductile) levels of fracture toughness. Previous discussions have centered on the relationships of the DT energy values to K_{Ic}/σ_{ys} ratios for conditions of unstable (plane strain) fracture. Relationships of the energy values to the R curve slopes, for conditions of stable fracture, have been established recently. The slope of the R curve is determined by the use of the DT test in two configurations—the standard DT test geometry and a "reduced-run" geometry. The R curve slope is defined by the difference in plastic work energy required for fracture in the two configurations.

The significance of these statements may be understood in terms of a general discussion of geometric effects. Figure 35 presents schematic drawings for three types of fracture test configurations. The top schematic drawing (1) illustrates a *fracture mechanics test* which is configured and instrumented to detect the conditions for the initial extension of a crack. Frangible metal is thus characterized by the K_{Ic} or K_{Id} parameters. A ductile metal of steeply-rising R curve features would be "characterized" by a K_x (invalid) measurement.

The schematic drawing (3) at the bottom of Fig. 35 represents the standard DT test geometry, noted as the *plane stress configuration*. The feature of this configuration is that sufficient fracture extension path is provided to *allow* the development of whatever degree of plane stress fracture toughness that is characteristic of the metal. The drawing indicates the fracture extension transition from a small plastic zone at the crack tip (plane strain) to a large plastic enclave (plane stress), which is characteristic of a ductile metal.

The schematic drawing (2) at the center of Fig. 35 illustrates a reduced-run version of the DT test. It is defiined as a *plane strain configuration* because the growth of the plastic enclave is largely prevented due to inadequate fracture extension path. There is no intent to claim that the conditions are fully equivalent to the plane strain constraint acting at the crack tip. The purpose is to obtain an energy absorption measurement, by practical procedures, which is related to a condition which approaches plane strain constraint.



PLANE STRAIN CONFIGURATION — PLANE STRESS CONFIGURATION

Fig. 36—Illustrating that the geometry of the standard Charpy V (C_v) test is of plane strain configuration type. Increasing the fracture extension distance for the C_v test results in a fracture transition from essentially flat to slant fracture features (plane strain to plane stress). The standard C_v test never develops a slant fracture, even for highly ductile metals, because of the configuration features

The graph at the top of Fig. 35 illustrates schematically the rise of the energy per unit fracture area (E/A) measured by the three test methods.

PLANE STRAIN CONFIGURATION — PLANE STRESS CONFIGURATION

Fig. 37—Illustrating the change in fracture appearance features of the DT test for the plane strain and plane stress (standard) configurations. The fractures are representative of the plane strain to plane stress transition for a ductile, high-shelf steel. The difference in E/A values for the two test configurations defines the slope of the R curve

The E/A value for the fracture mechanics test (expressed as $\mathcal{G}$ which is an E/A measurement) relates to the full constraint condition acting at the crack tip. The E/A value for the plane strain DT configuration should show an increase which is related to a decrease in the level of plane strain constraint. The E/A value for the plane stress DT configuration should show an additional increase, which is related to loss of constraint as the plane stress state evolves. The three curves of Fig. 35 represent metals of frangible (flat curve), shallow-rise and steep-rise R curve characteristics. The slopes indicate the Δa (crack extension increment) rate of transition to lower levels of constraint, i.e., rate of transition to plastic fracture. A practical procedure for measurement of the R curve slopes evolved from comparison of the E/A values of the two DT test configurations, i.e., fracture extension (2) compared to (3).

It is interesting to note that the C_v test configuration results in a mechanical constraint condition of near-plane strain type. This aspect has been described in detail by Clausing of the U.S. Steel Laboratory. Figure 36 illustrates the change in fracture appearance resulting from the increase in the fracture path length of the C_v test specimen. It should be noted that C_v test specimens never show full slant fracture, even for the most ductile metals. In effect, the C_v test configuration provides a measurement of fracture extension resistance differences that are related to point 2 of the graph in Fig. 35. The larger differences that are measured at point 3 are not disclosed.

Similar changes in fracture appearance of the DT test in the two configurations are indicated by Fig. 37. The illustration relates to a steel of high shelf, steeply-rising R curve characteristics. This is evident by the development of a full-slant fracture for the standard DT test configuration.

Characterization of *R* Curves

The R curve characteristics of structural steels, aluminum and titanium alloys are under active investigation at the Naval Research Laboratory. At this point, we shall present the results of the first exploratory studies conducted in mid-1970; these vividly illustrate the R curve characteristics of a wide variety of steels. In the studies the DT test specimen configuration was modified by introducing variations of fracture length. The steels selected for R curve studies were 1 in. thick plates, representing a wide range of yield strength and DT test shelf properties. The tests were

conducted at temperatures corresponding to the upper shelf, i.e., above the temperature transition region.

A graphical representation of the range of yield strength and shelf properties is provided by plotting these data on the steel RAD—Fig. 38. The method of RAD plotting should be reviewed at this point. The section thickness (B) determines the maximum plane strain constraint that can be developed at the tip of a through-thickness-crack (or plate-edge crack). With increasing K_{Ic}/σ_{ys} ratio value, there is an increase in plastic zone size, i.e., increasing ratio signifies increasing metal ductility. Since a given section thickness provides a fixed value of mechanical constraint, there is a limit to the plane strain K_{Ic}/σ_{ys} ratio value that can be measured—for 1 in. thickness this limit is 0.63. The absence of ratio lines above 0.63, in Fig. 38, indicates that the higher ratio values, depicted in the conventional RAD, have no meaning for plates of 1 in. thickness—hence, these are not shown. The significance of DT energy values, which lie above the 0.63 ratio line, is that they represent increasing resistance to plastic (plane stress) fracture. The index of increased resistance is the DT shelf energy level.

Seven of the steels of this study involve plane stress fracture properties, as is shown by the location of the data points above the 0.63 ratio line. The reference code for these seven test steels indicates the relative position in the RAD, i.e., low (L), medium (M), and high (H). Steel F-1 is indicated to lie below the 0.63 ratio line and, therefore, should be characterized by the plane strain fracture toughness (K_{Ic}) scale.

The RAD location of steel F-1 also indicates that unstable fracture propagation should result, following the "first event" of initial extension from the crack tip. The R curve for this steel should be "flat" (nonrising). The test data confirmed these expectations —the E/A values for specimens of short and long-run geometrical features remained constant at a low value, 100 ft-lb/in.^{2}—Fig. 39. The fracture appearance was flat, without evidence of shear lips, for all test configurations.

The high-shelf RAD location of steel H-6 signifies a high level of plane stress fracture resistance. The R curve for this steel should be expected to exhibit steep slope, deriving from gross blunting of the crack tip and the formation of a large plastic enclave. The expected large increase in E/A with tear extension length is illustrated in Fig. 40.

The low-shelf RAD location of steel L-1 signifies low resistance to plane stress fracture extension. The R curve should be expected to exhibit low slope, deriving from the slight blunting of the crack tip and the formation of a small plastic enclave. The expected behavior of this steel is confirmed by the data presented in Fig. 41. The shallow R curve is indicative of a much slower build-up of E/A with increasing Δa as compared to the high shelf steel described above. The fracture surfaces shown above the curve demonstrate that this steel does not attain conditions of full-slant fracture, as should be expected from the limited rise of the R curve. The characteristic metallurgical ductility of this steel is not sufficient to cause crack tip blunting. Thus, the fracture extension resistance is only marginally above that of frangible materials. These various features are predictable from the RAD location, barely above the 0.63 ratio line for plane strain.

Figure 42 presents a summary of the R curve characteristics for the eight test steels. The relative level of the R curves, as well as the slopes, may be noted to rise in the general order of the DT shelf energy values plotted in Fig. 38. Because of the characteristic form of the curves, it is not necessary to conduct extensive tests to plot the full R curve. Two points, derived from a plane strain DT type test and the standard DT test, are sufficient to establish the slopes of the R curves.

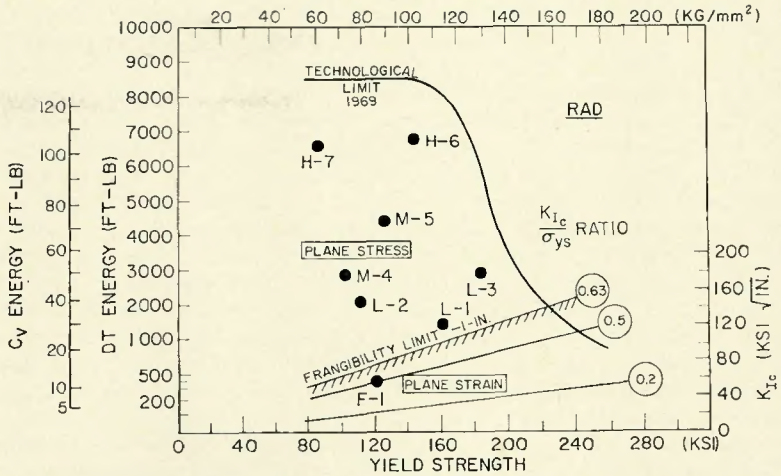


Fig. 38—Location of the test steels in the plane strain and plane stress regions of the Ratio Analysis Diagram (RAD). For steels of 1 in. thickness (B), plane strain fracture is limited to the zone below the 0.63 ratio line. The RAD prediction is based on the fracture mechanics relationships $B \geq 2.5\ (K_{Ic}/\sigma_{ys})^2$ for plane strain fracture

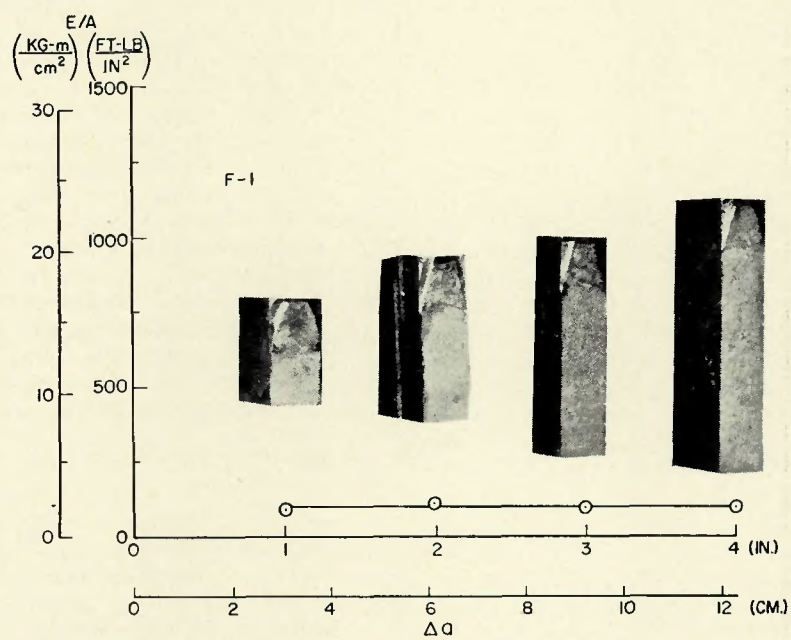


Fig. 39—Fracture appearance and R curve features for the frangible, F-1 steel

A curve fitting analysis of the R curves was made by plotting the data on log-log coordinates, which results in the straight-line relationships shown in Fig. 43. The slopes of the lines vary only between 1.85 and 2.0 for these seven steels. If the slope is taken as approximately 2.0, a relation is inferred where energy varies as $E = R_p(\Delta a)^2[f(B)]$ R_p is a constant defining the position of the curve on the log-log plot, which is different for each steel, Δa is the crack extension, and the exponent 2 is the slope of the log-log plot. The $f(B)$ term represents a function of the thickness (B) of the specimen.

These data relate to a single value of thickness; establishment of generalized relationships, including thickness (B), requires tests involving wide variations of both Δa and DT test specimen thickness. The $f(B)$ relationships are being established in a highly satisfactory manner, and it appears that the $f(B)$ term is approximately $B^{1/2}$. The ability to express the characteristic R curve features by an empirically derived equation, involving only specimen dimensions and a material dependent constant (R_p), is most valuable since it will permit characterization of materials of any thickness and resistance level by a single procedure.

The foregoing analyses do not imply that the resistance to fracture extension increases indefinitely with extension. The analyses apply only for the extension interval which involves the *fracture mode transition* from plane strain constraint at the crack tip to the characteristic plane stress mode of the metal. The R curve rise which derives from this transition provides the "barrier" to fracture extension and the analyses relate directly to this factor. Following completion of the fracture mode transition the R curve should "saturate," i.e., level out to a characteristic fixed level of resistance to continued extension. It is emphasized that the characteristic fracture mode is in fact attained by the standard DT specimen. This fact has been confirmed by extensive comparisons confirmed that the DT test pro-Cylindrical Explosion Tear Test (8 in. fracture run for a 1 in. thick plate) and the fracture mode attained by the DT test specimens. The fracture modes were found to be identical for both test procedures in all cases, as described previously. These comparisons confirmed that the DT test provides a proper definition of the plane stress fracture features of the metal.

Comparison of the H-7 steel with H-6 (Fig. 42) indicates identical R curves and significantly different values of yield strength. These aspects are independent parameters for different steels and for different orientations in the same steel. In this respect, reference is made to the wide variations in DT shelf energy characteristics for the same strength level that are represented by the data of Fig. 30. The recent R curve studies indicate that the DT shelf energy value is directly related to the R curve slope, i.e., steels of different yield strengths will feature the same R curve slope if the DT shelf energy values are the same. However, this observation does not imply that the same R curve slope will be adequate for structural design involving steels of increasing yield strength. In fact, the fracture extension resistance must be increased to offset increased allowable stresses, as described below.

On the RAD, Fig. 31, a constant critical flaw size-stress level condition is denoted by a constant K_{Ic}/σ_{ys} ratio line; increases in yield strength require increases in K_{Ic} to maintain the ratio at a constant value. The slope of the ratio line with respect to the DT energy scale signifies that increased DT energy absorption is required to offset the effects of increased K levels resulting from increases in yield strength. Similarly, there is a requirement for increased fracture extension resistance (R curve slope) to offset the increased level of allowable plastic stresses (acting on a given flaw) which result from increasing yield strength. This requirement must be met by increasing the DT shelf energy value as a function of increasing yield strength.

In simple terms, the prevention of fracture extension for a specified flaw size will require increased R curve slope (increased DT shelf energy) because of the increase of allowable stress levels which result from increased yield strength. The analytical definition of specific design requirements for fracture extension resistance (in terms of R curve slopes or the equivalent DT shelf energy) must be related to flaw size, allowable stress level and compliance characteristics of the structure. These relationships must be evolved by correlations with structural prototype tests for reasons which are explained in the following section.

Engineering Significance of *R* Curve Slopes

Information derived from the characterization of R curve slopes should provide for correlations with *prototype structural tests*. These correlations should result in empirical calculation capabilities for fracture-safe design based on critical plastic stresses or plastic strains for fracture extension. A logical question then arises: Is it possible to evolve such analytical procedures from first principles, i.e., from laboratory test measurement of critical plastic stresses or plastic strains for fracture extension, which are coupled directly to a mathematical analysis? If so, a case would be presented for waiting for the evolution of more sophisticated laboratory tests.

Unfortunately, a generalized analytical approach which would apply to all types of structures is not feasible, except as evolved empirically by structural prototype testing. Generalized analytical approaches, based on first principles, run into complications of structural features which per se cannot be analyzed directly by mechanical principles. For example, a tensile loaded plate structure will not behave in the same fashion as an internally loaded pressure vessel. The presence of flaws in pressure vessels may cause bulging (geometric instability); moreover, the energy available for fracture extension depends on the pressurization medium (gas or liquid), diameter to thickness ratio, etc. For these reasons, it is necessary to analyze the structural response on a case basis. Such studies are being conducted in various laboratories for gas transmission lines and various types of pressure vessels. If a proper sequence of such tests is repeated for different generic structural configurations (flat plate structures, etc.), a generalized analytical framework keyed to simple laboratory tests should emerge.

Definition of the R curve slopes provides information which relates to the resistance developed in first stages of fracture as well as the resistance to indefinitely continued extension. Both of these aspects increase with increase in the R curve slopes. It is not necessary to define these aspects separately for engineering purposes. Correlation to structural prototype test results is served by definition of the R curve slopes. The prevention of an extensive fracture is decided in the first stages of extension, i.e., while the R curve characteristics of the metal come into play. If fracture extension continues through this phase of increasing resistance, the energy barrier to extensive propagation is breached. Thus, arrests should not be expected after a stable level of fracture resistance is attained. Ordinarily, increasing fracture extension force will evolve in the process of enlargement of a tear, to a degree that is determined by the compliance characteristics of the structure. The R curve slope must be adequate to offset the increase in fracture extension force —that is, the increase in the "re-

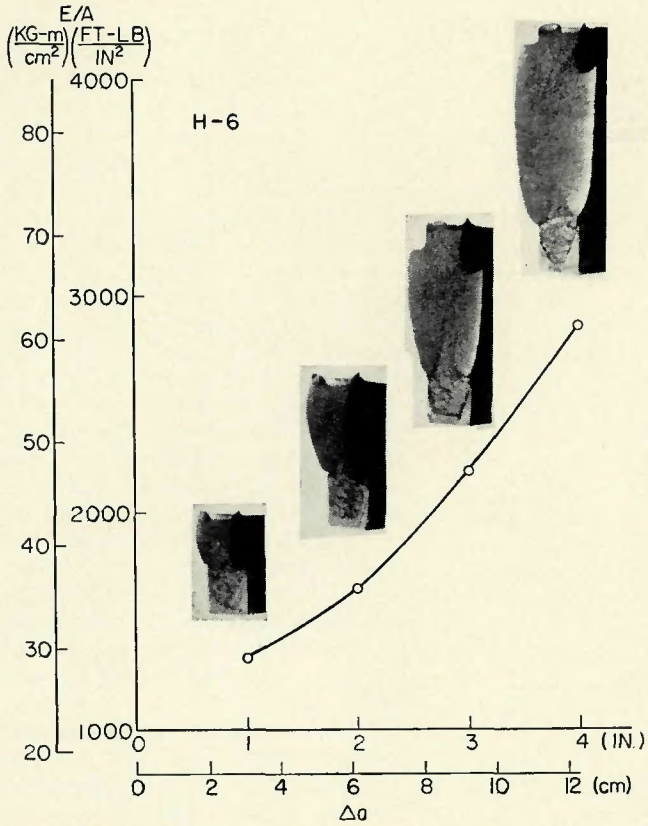


Fig. 40—Fracture appearance transition and R curve features for the high shelf, H-6 steel

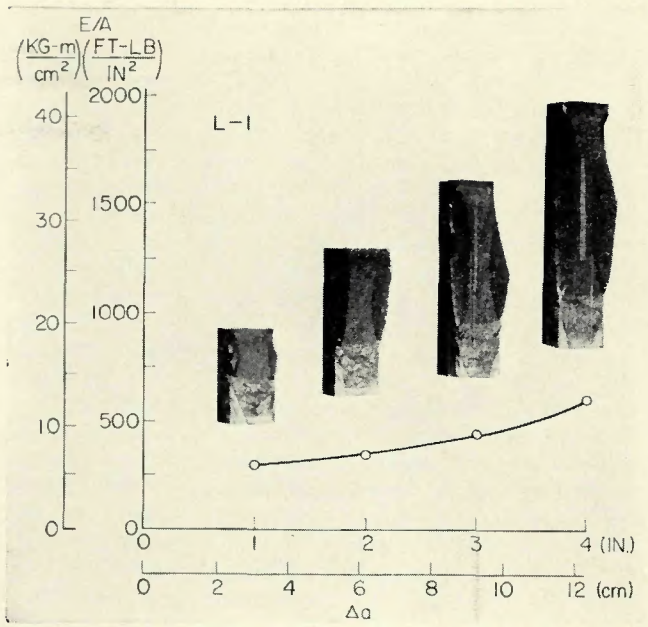


Fig. 41—Fracture appearance transition and R curve features for the low shelf, L-1 steel

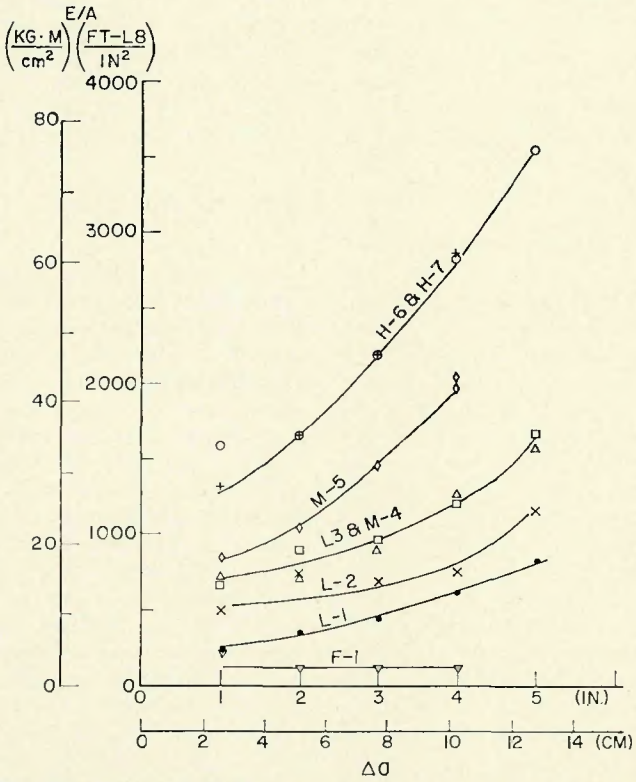


Fig. 42— Summary of E/A-Δa data which define the R curve features of the various test steels. There is an increase in R curve slopes and increased intercept values (first extension increment) with increase in shelf level as defined in the RAD plot of Fig. 38

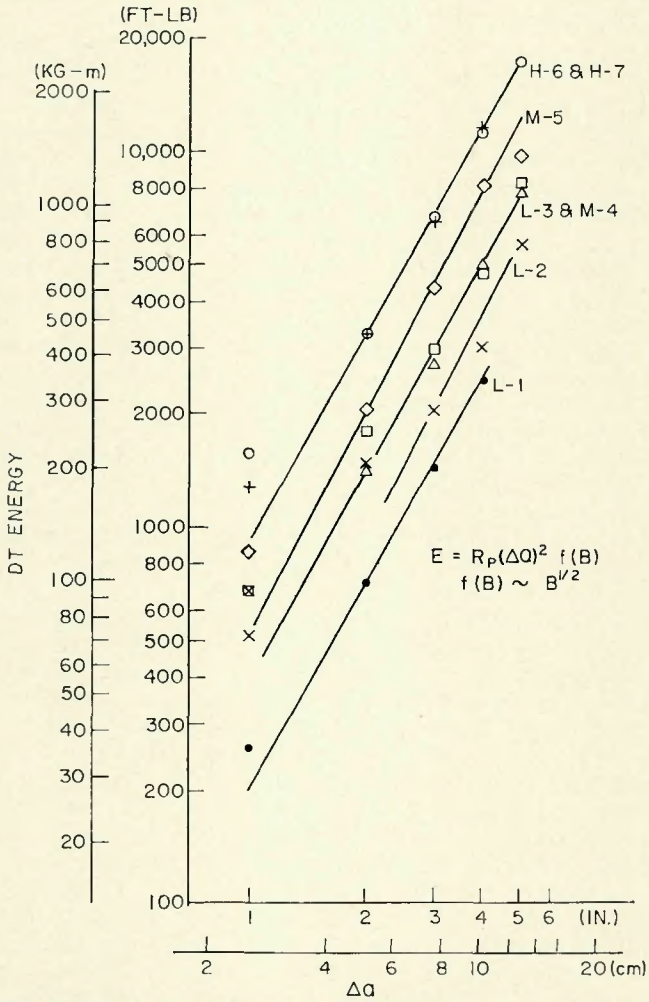


Fig. 43—Log-log plot of Fig. 42 data. The nature of the relationships may be expressed by the formula noted in the graph, where E=energy, R_p=constant, Δa=extension length and $f(B)$ or $B^{1/2}$=expression of the thickness (B) constraint effects as deduced from other studies. These relationships provide for definition of the R curve by two points, derived from the E values of the two DT test configurations

quired" metal separation force must exceed the increase in the mechanically applied force.

R curves which are flat or of very low slope indicate that unstable fracture propagation is possible. If fracture instability is possible, then there is no need to consider the structural aspects of compliance—either rigid or compliant structures will be subject to fracture initiation at predictable values of K_{Ic} or K_{Id} as applicable.

There is a wide gamut of potential engineering failure conditions which can involve geometric instability. In general, these are conditions which result from high compliance response to the load system. An example is the development of a flaw for internally pressurized vessels or piping of sufficient length to cause localized bulging in the flaw area. The consequent plastic loads acting on the flaw ends can cause fracture extension for ductile metals. The difference in initial flaw length that can lead to such extension is related to the *R* curve features of the metal. Very large differences in the critical flaw length (inches to feet) for initiating continued tear extension would be represented by major differences in *R* curve slopes.

These aspects apply in a most important manner for gas transmission pipelines. In this case, the engineering consideration is the reduction in the fracture extension velocity, to the point that the gas pressure release occurs at a rate sufficient for unloading of the hoop stress acting on the propagating tear. Low-slope *R* curve metals may be expected to permit relatively high rates of fracture extension while high-slope *R* curve metals should decrease this rate dramatically, leading to fail-safe conditions. The engineering problem is the definition of the adequate (minimum) *R* curve slope features which provide for fracture arrest (by pressure release), for the specific diameter and pressurization level of the pipe. While C_v shelf level values are being used for such correlations, more exact definitions should be possible by means of full-thickness DT tests in the proposed configurations. An additional advantage of the DT test is that it provides an exact definition of the temperature of transition to shelf levels for the thickness involved. The C_v specimen can define this point only by another correlation. The requirements for two types of C_v correlations introduces undue complexity in the analyses.

The use of high strength commercial steels in ships poses questions of ductile tear extension possibilities. Commercial, low cost, Q&T steels which are being considered may feature *R* curves of very low slopes. The high compliance of ship structures in amidship regions provides ideal conditions for fracture extension in such materials. It is reasonable to suspect the safety of such structural configurations and to consider utilizing metals of reasonably high-rise, *R* curve characteristics for the critical amidship region. The general specifications for steels of these grades (A514-A517) cover a broad range of alloy types and are not definitive as to steel quality aspects which relate to shelf characteristics. Thus, the shelf features of these steels may vary over a wide range without a definable lower limit. A lower limit must be set to provide a significant degree of plane stress fracture resistance for the section size. Specification procedures for ensuring that this is the case are not being utilized at the present time.

The *R* curve relationship to directionality factors deserves mention. In special cases, directionality effects may be used as an engineering design tool. In such cases, an assessment should be made of the benefit to be achieved by controlled rolling and orienting the metal so that the "strong" direction is placed normal to the fracture extension path. The attendant increases in fracture extension resistance are not minor, but are of half order of magnitude scale or larger. Thus, directionality effects may serve as solutions to special problems—without increase in cost. An example is the possible use of low cost straight-away-rolled steels in spiral orientation for the case of gas transmission pipelines. Another example is the use of such orientation in fore-aft direction for ships—the importance of correct orientation should be clearly recognized.

The *R* curve indexing to the RAD adds the feature of zoning which includes the *type of structure*. For example, low shelf regions of the RAD represent plane stress fracture resistance which may be adequate for relatively noncompliant structures. With increase in compliance features and total available structural energy, there is a need to select steels from the high shelf locations of the RAD. Thus, the high shelf regions may be defined as the high compliance-high energy structural regime. Conversely, the low shelf regions may be defined as the low compliance-low total energy structural regime. Regions below the ratio line, which applies to the specific section size, may be defined as the unstable fracture (for any type of structure) regime.

The metallurgical zoning of the RAD, which was discussed previously in terms of significance to the ratio lines zone, can now be related to the aspects of plane stress fracture toughness for the zone above the ratio line for the thickness involved. Reference should be made to the "metallurgical corridors" described in Fig. 32. The remarkable effects of furnace practice variables in controlling strength level —*R* curve relationships become apparent. For example, very large increases in fracture extension resistance may be attained in the yield strength range of 70 to 130 ksi (50 to 90 kg/mm^2) by the use of steels of higher corridor features. The corridors defined in the figure derive from alloy content, furnace, and deoxidation practice variables. Increasing strength causes a decrease in plane stress fracture resistance which is characteristic of the corridor. The corridors represent steel quality and as such are also "cost corridors," i.e., increased cost is involved in moving to higher corridors.

The metallurgical factors which determine corridor and shelf levels are well understood. If the engineer appreciates these factors, there should be no aversion for payment of a reasonable premium for the purchase of intermediate strength steels of higher corridor and, therefore, higher *R* curve slope features, when required. The concept of "buying strength at lowest cost" may be counterproductive and of catastrophic consequences. Questions of relative structural requirements and metal cost can now be analyzed, so that the end result is a fracture-safe structure of lowest cost.

Applications to Computer-Aided Design

The consolidation of scientific knowledge that led to the evolution of the described engineering principles for fracture-safe design is of very recent origin. The studies of the 1940s and 1950s were of preparatory nature, i.e., they served the purpose of sorting of options and finding the way to the desired consolidation of metal, mechanical and structural factors. The consolidations and, therefore, the evolution of fracture-safe design practices were attained in the 1960s, and most notably since 1968. In summary, the new integrated engineering technology of fracture-safe design has been available for relatively few years. The accelerating pace in the solution of primary issues in recent years indicates promise for resolving the remaining problems (section size effects, *R* curve aspects, heat-affected zone properties) in the near future.

Because of the very recent evolu-

tion of this new engineering technology, its application has been largely restricted to organizations which include fracture research specialists. The generality of the new principles, and the simplicity of application based on analytical diagrams of the FAD-RAD type, indicates a clear potential for use on a broad front. Thus, all practicing engineers may make direct use of these procedures for their specific problems. For example, the metallurgist is concerned with the improvement of metals, the welding engineer is more directly concerned with weld and heat-affected zone quality, etc.

The importance of these new developments to the field of structural design can be appreciated by examination of the present state of design practices. The usual considerations are restricted to strength (σ_{ys}), modulus (E) and density (ρ). The application of fracture-safe design principles is a final step, after the general design of the structure is completed. If the new principles are to be used properly they must be included in concept analysis, in predesign selection of options and in the initial design process—not as a "last item" consideration. The available options for application of these principles may be severely limited after the design has been decided.

The exciting new developments in the design field have centered on the recent evolution of Computer-Aided Design (CAD) practices. The σ_{ys}, E, ρ parameters are utilized in computer programs for rapid machine definition of "least weight" designs. As such, there is no basic improvement in design practices—the benefit is in expediting of the calculations. Substantial improvement in design practices should evolve by inclusion of fracture-safe design principles in the machine analysis process.

The analytical diagrams of the FAD-RAD type have the appropriate features for reduction to digital form required for CAD processing. For example, finite zones of the RAD can be referenced to σ_{ys}—fracture resistance parameters (K_{Ic}/σ_{ys} and R curve slope). The corridor locations of the references zones define the generic type of steel, relative cost, weldability, etc. A digital reference network for the RAD would provide for direct machine iteration to these factors, in concert with the conventional parameters.

The possibilities for evolving a fully rational CAD procedure, which includes the RAD and FAD-defined parameters, are being examined for ship design by Professor K. Masabuchi and associates at the Massachusetts Institute of Technology. The aim of the studies is to evolve machine analyses which converge to optimized solutions for "least weight and fracture-safe" designs. Other factors such as relative metal costs, weldability, required inspection quality, etc. can be considered by including additional computer cards. The achievement of these aims is only a matter of time. Since the analyses can be made sequentially by designers, they can be made in closed-loop form by computers. The main advantage of machine processing is that a wider set of options can be examined in a short time.

Closure

An index of the advances which have been described can be gauged by the type of questions posed at the start and end of this review. At the start, the questions involved "how to use the C_v test," at the end, "how to use computers"! These attainments are the result of contributions by many disciplines—and by scientists, technologists and engineers, within the disciplines. The achievements of the research specialists should now be matched by practicing engineers in utilization of the information. The most crucial problem at this time is educational rather than technical, i.e., the transfer of research knowledge from the literature to practice. Because of the rapid advances, the information gap between the specialist and engineer has widened in recent years. A reversal of these trends will require concert effort by both groups —this review is a contribution towards this goal.

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