NRL Report 6787

Dynamic Tear Test Definition of the Temperature Transition from Linear Elastic to Gross Strain Fracture Conditions

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November 29, 1968

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ABSTRACT

Transition temperature concepts for fracture-safe design have been based on the relatively narrow temperature range evidenced by the fracture-mode transition from plane strain to plane stress. Fracture mechanics theory has suggested that large increases of section size should provide sufficient mechanical constraint for retention of plane strain conditions through the transition temperature range. Recent investigations based on Dynamic Tear (DT) tests of thick-section reactor-grade steels (A-533B) have provided clear evidence that the plane strain to plane stress transition is not eliminated. These findings are of major consequence to concepts of fracture-safe design and show that temperatures significantly above NDT + 100°F can provide a conservative margin of safety, even for the condition of very large flaws and plastic overload. The DT test, as conducted using a small specimen, defines the temperature range of transition from linear elastic to gross strain mechanical conditions of fracture and, therefore, the required analytical treatment for flaw size-stress calculations. Fracture mechanics concepts are brought into consonance with transition temperature concepts of long standing, and the importance of considering fracture initiation in terms of limiting dynamic fracture toughness values is emphasized.

PROBLEM STATUS

This report completes one phase of the problem. Work on other aspects of the problem is continuing.

AUTHORIZATION

NRL Problem M01-18
Project RR 007-01-46-5420

Manuscript Submitted August 6, 1968.

DYNAMIC TEAR TEST DEFINITION OF THE TEMPERATURE TRANSITION FROM LINEAR ELASTIC TO GROSS STRAIN FRACTURE CONDITIONS

INTRODUCTION

Much effort is underway to assess the fracture toughness of engineering steels and to relate this to tolerable flaw sizes that will not precipitate catastrophic structural failure. Among the approaches developed to cope with this problem, linear elastic fracture mechanics is receiving the greatest research emphasis. This approach was formulated within the last decade as a means to characterize the critical flaw size-stress level relationship for a material where plastic flow is limited by plane strain conditions. The method employs a linear elastic description of the stress field around a cracklike defect which can be related to instability conditions by a single parameter, K_{Ic}. The approach is generally applicable to ultrahigh strength steels which fracture under elastic loading in the presence of relatively small flaws. Attempts are being made to apply the analysis to the lower strength and generally tougher steels. These lower strength steels are not restricted to use in the relatively low temperature, frangible region which is suitable for linear elastic analysis. Accordingly, effort has been required to extend the fracture mechanics approach to elastic-plastic behavior, for which the flaw boundary is associated with a plastic region that is large with respect to the flaw, yet the overall section is elastically loaded. Analytical treatment of this condition requires a radical departure from ordinary linear elastic analysis; this is presently in the early stages of development. Another area which is deemed to require specialized analysis deals with gross plastic strain, i.e., the regime where failure may occur in the presence of a flaw when the entire section is plastically loaded. Very little effort is being expended to analyze this mode of failure along the lines of fracture mechanics. For the gross strain condition, engineering designs are normally governed by strength-of-materials analyses which do not take flaws into account. Safety is considered to be attained by limiting stresses to such magnitudes as will prevent plastic overload (e.g., buckling, bursting) in the absence of flaws.

The engineering community faces a dilemma in the absence of a fracture toughness characterization test which provides for estimating stress level-flaw size relationships for fracture prevention relating commonly to the linear elastic, elastic-plastic, and gross strain fracture regimes. As an attempt to resolve this problem, NRL has developed a new tool for determination of fracture toughness, known as the Dynamic Tear (DT) test. This test method and related analysis procedures provide for expansion of fracture-safe design beyond the range of validity of linear elastic fracture mechanics and into the plastic regime. Since analytical exactness is not presently attainable for the nonfrangible cases by any method, approximate definitions of failure conditions obtained by the use of a simple test represent a net overall gain. Additionally, the element of confusion as to applicable mechanical conditions involved in the fracture process is largely eliminated.

The DT analysis procedure expresses the material's tolerance for flaws in a framework of fracture energy and fracture appearance. For conditions involving linear elastic fracture, correlations between the plane strain toughness, K_{Ic}, and the DT energy have been evolved (1). The need for accuracy in flaw size determinations diminishes rapidly with the increase in toughness when moving through the elastic-plastic regime and into that of gross plastic strain. The DT approach is based on a simple engineering test procedure that is more discriminating with respect to variations in

toughness than other common engineering tests, such as the Charpy-V (C_v) test. The DT test method gives additional meaning and interpretation to the Drop Weight Test and the Fracture Analysis Diagram (FAD) (2).

The DT test has been used at NRL to investigate a wide variety of representative commercial steels in the thickness range of 1 to 3 inches. Correlations of DT transition temperature curves with the Drop Weight Test NDT temperatures have been developed for purposes of integrating the data. The Drop Weight NDT defines a transition temperature above which shear lips develop at the fracture surface and provide a mechanism for absorbing the energy release of a small popin with consequent arrest. The development of shear lips indicates the loss of plane strain conditions near the free surfaces and the start of a rapid, temperature-induced transition in the direction of full 45-degree-slant fracture. The same events control the performance of the DT test. The discussion of the DT test in this report will be restricted to a particular material which exhibits this transition-type behavior, namely, A-533B steel. This steel is the most commonly used nuclear reactor pressure vessel material for pressurized water and boiling water reactor systems. Applications of the DT procedure to ultrahigh strength steels and to intermediate strength steels which do not exhibit a temperature transition to high toughness are discussed in Ref. 1.

DYNAMIC TEAR TEST DESCRIPTION

The basic DT test procedure consists of impacting a simply supported specimen having a specially prepared brittle electron-beam crack-starter weld on the tension side (Fig. 1). Six specimen sizes are currently in use. The geometry of the specimens is given in Table 1. Comparison studies are also being conducted using a 5/8-in. DT specimen featuring a deep machined notch and a knife-edge-pressed tip in place of the electron-beam weld. The purpose of the brittle crack-starter weld is to simulate a natural popin which may occur in service. This popin, coupled with dynamic loading, creates the most adverse mechanical circumstance that may be encountered in a structure, i. e., limit severity conditions for fracture initiation. It should be remembered that even a statically loaded pressure vessel can behave as if dynamically loaded insofar as the tip of the crack is concerned. A sudden extension in crack front of very small magnitude due to the presence of a locally embrittled region (such as a strain aged zone) or a short duration load pulse (such as water hammer) may be all that is required to make the flaw behave as if it were dynamically loaded. If the dynamic fracture toughness is low, the described instability is then propagated through the structure without arrest.

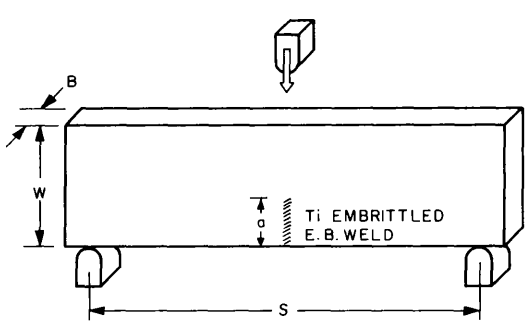

Fig. 1 - Dynamic Tear (DT) specimen

Table 1
Dynamic Tear Test Specimen Geometry

Specimen Designation (in.)	B (in.)	W (in.)	S (in.)	a (in.)	Weight of Steel Spec. (lb)
5/8 DT*	5/8	1-5/8	6-1/2	1/2	2
1	1	4-3/4	16	1-3/4	24
2	2	8	26	3	127
3	3	8	26	3	190
6	6	12	60	3	1220
12	12	15	84	3	4580

*Alternate 5/8" DT features a 1/2" machined notch terminating in a knife edge pressed tip (5/8" DPN-DT).

The crack-starter welds for steel specimens are prepared by diffusing a small titanium wire through the thickness for a depth *a* as indicated in Fig. 1. This forms a highly brittle, iron-titanium alloy which requires very little energy to initiate a crack and propagate it into the base metal. The top edge and the sides of the weld are notched in a trapezoidal pattern for ease in starting the crack.

The various DT specimens are fractured using pendulum or drop-weight machines, and the energy for fracture is recorded. The 5/8-in. and 1-in. specimens are fractured on pendulum machines with capacities of 2000 ft-lb and 5000 ft-lb, respectively. The energy value recorded is the difference between the initial and final potential energies of the pendulum, just as in the Charpy-V test. The larger specimen sizes are usually tested on drop-weight machines having instrumented tups. A record of specimen load vs time is obtained from the instrumented tup and is used to calculate the energy value. An independent method of determining the same energy value as found with the instrumented tup is by use of calibrated lead bricks that absorb the residual energy of the weight after fracture. The energy recorded with the latter method is the difference between the initial potential energy of the weight and the energy absorbed by the lead bricks.

The above methods of measuring energy include several extraneous contributions to the desired quantity of energy to fracture. These contributions include the specimen kinetic energy after fracture, the energy involved in deforming the contact surfaces, and the energy lost to the foundation. For tough specimens the indentation and kinetic energy terms are small compared to the overall energy absorbed. For brittle specimens the total energy is small. In this case the kinetic energy (estimated by assuming the specimen halves to rotate with the tup velocity) is probably larger than the fracture energy and thereby provides an "effective" zero on the energy scale which is an order of magnitude smaller than the maximum energy absorbed by a tough specimen. Since primary interest centers on the relative energy rise over a temperature interval as a measure of toughness increase, it can be argued that the inclusion of extraneous energies will not detract from the overall energy-temperature trend due solely to the fracture process .

A typical DT energy curve for A-533B steel and other steels having sharp transition temperature features is shown in Fig. 2. The energy curve rises slowly with temperature up to the NDT point with specimens exhibiting pure cleavage fracture. Above the NDT temperature point, the energy rises rapidly up to a shelf level that is attained about 150° F above NDT. The rise in fracture energy with temperature from the NDT point is associated with the development of shear lips which progressively increase in size until fully ductile, slant fracture is attained on the shelf as indicated schematically on the figure for steels of 1- to 3-in. thickness. The fracture mode in the central region of the specimen changes from cleavage to high energy dimpled rupture as the shelf temperature region is reached. At this point the brittle weld popin would be immediately arrested because of high toughness (and associated crack tip blunting) were it not for the fact that the testing machine continues to drive the fracture in the gross plastic strain mode until separation is achieved.

The relation of the Drop Weight NDT test (DWT) to the DT test is also shown in Fig. 2. The NDT point always occurs at the toe region of the DT energy curve and denotes the point at which shear lips begin to develop. The NDT is often indexed by means of a correlated energy level or "fix" with the Charpy curve. This point, however, does not consistently fall at the toe region of the C_v curve and may lie anywhere from the toe to the upper shelf region on the Charpy curve for different steels. The initial rising portion of the DT curve follows the general form of the Robertson Crack Arrest Temperature (CAT) curve for 1- to 3-in. plates indicating the same degree of dynamic fracture toughness increase with temperature. This correspondence is expected because the CAT curve in the Robertson test evolves from the same sequence of events, i.e., transition from linear elastic to elastic-plastic fracture modes. The Robertson test is limited to investigations of fracture propagation for conditions of nominal elastic loads. It is not possible to conduct Robertson tests for plastic overload conditions. Therefore, the gross strain regime is not defined, although it is inferred by extrapolation. The extrapolation of the CAT curve was evolved by consideration of explosion-bulge crack-starter tests and are so defined by the FAD extension of the CAT curve (2).

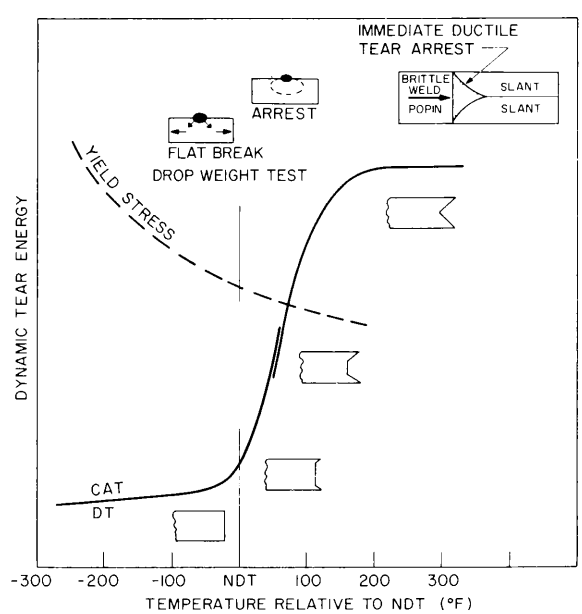

Fig. 2 - Typical DT energy curve for a steel having transition temperature features. The schematic relationships between energy and fracture appearance are illustrated.

FRACTURE MECHANICS EVALUATION OF A-533B

Relatively few fracture mechanics K_{Ic} values are available for A-533 steel because very large specimens are required for obtaining valid plane strain data in the temperature region of prime interest, i.e., room temperature and above. More data should soon be available, particularly in conjunction with the Atomic Energy Commission's Heavy Section Steel Technology (HSST) Program.

For purposes of illustration, data are presented in Fig. 3 for A-302B steel, which is the forerunner of A-533B. The K_{Ic} data were obtained by Wessel (3) using 1X, 2X, and 4X WOL, as well as some spin burst tests, for A302B in the normalized and tempered condition. The data exhibit a rather shallow rise in toughness from liquid nitrogen temperature up to -100° F. Above this temperature (approximately NDT - 100° F) a rapid increase in toughness is indicated. An entirely satisfactory explanation of this rise in terms of crack tip plasticity analysis is not yet available. Part of the indicated rise may be due to the fact that the data above NDT - 100° F are not valid according to the current standards of the ASTM Committee E-24 on Fracture Testing of Metals. If this is the case, the toughness increase with temperature is relatable to a loss in the required plane strain conditions at the crack tip, thereby resulting in K_c mixed-mode or plane stress conditions. A schematic curve showing this behavior is illustrated in Fig. 3. Obviously, the temperature dependence of the schematic curve is related to the section size in question. The K_c departure from the K_{Ic} curve may occur in the valid K_{Ic} region of measurement whenever the specimen size is decreased below the required limit. The implications of K_c behavior are of obvious benefit to structural integrity when it can be shown that a material will be used in section sizes thinner than those required to achieve plane strain conditions. However, fracture mechanics has not yet set forth a rigorous analysis for the K_c situation to predict the critical flaw size-stress level relationship for fracture.

Some investigators are of the opinion that the upswing in toughness in the region in Fig. 3 considered invalid by ASTM standards is real and will be confirmed as thicker specimens having the required constraint become available. Others are of the opinion that thick specimens will result in an extension of the shallow rise in toughness with temperature as now evidenced with the valid data. The reconciliation of these opinions is of critical importance if fracture mechanics is to be of use in predicting conditions for

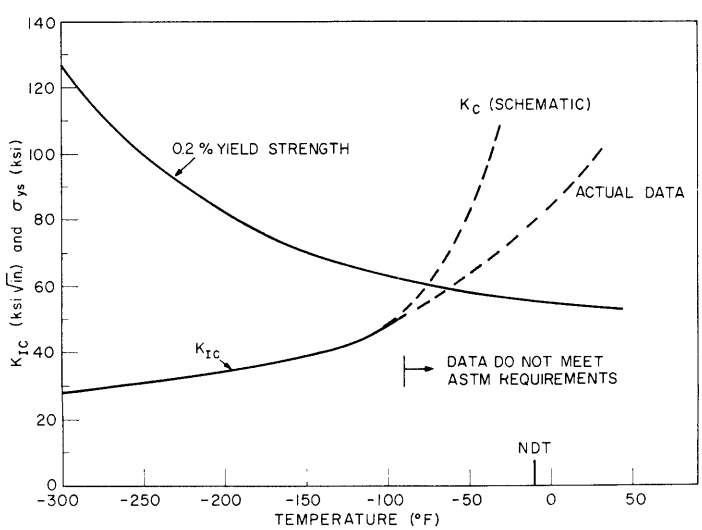

Fig. 3 - Plane strain fracture toughness (K_{Ic}) data for A-302B steel in the normalized and tempered condition

fracture. Unstable surface flaw depths for static loading associated with the toughness at the upper limit of presently established validity of Fig. 3 data (approximately 0.7 K_{Ic}/σ_{ys}) are of the order of 1 in. to 0.1 in. for stress levels ranging from 1/2 yield to yield and for different flaw shapes as determined from the following equation of Irwin (4):

$$K_{Ic} = \frac{1.1}{\sqrt{Q}} \sigma \sqrt{\pi a} \tag{1}$$

where

σ = applied stress in the vicinity of the flaw,

a = surface flaw depth, and

$1.1/\sqrt{Q}$ = specimen geometry and flaw shape factor obtained from tables.

If the toughness increases slowly so that the room temperature K_{Ic} value is only double or triple what it was at NDT - 100° F, then the critical flaw depths could still be of the general order of 1 in. for a particular static loading. On the other hand, if the toughness swings up strongly with increasing temperature around the NDT point, then the critical flaw sizes become so huge that it is unnecessary to calculate their magnitude with a high degree of accuracy. This is the current research question that is being investigated with fracture mechanics. Another relatively unexplored area deals with the effects of dynamic loading on decreasing the fracture toughness value in comparison to the conventional statically determined K_{Ic} value. The significance of dynamic loading will be discussed later.

DYNAMIC TEAR TESTS OF HEAVY SECTION A-533 STEEL

The effect of mechanical constraint associated with thick sections could cause material which exhibits tough, 45-degree-slant fracture in thin sections (1 to 3 in.) at a given temperature to behave in a low energy, brittle fashion in thick structural applications at normal elastic loading in the presence of large flaws. To establish if this, in fact, is the case and to obtain a rapid answer to the question, NRL is conducting a series of DT tests on various thick plates with specimens ranging in size from 5/8-in. to full-plate thickness. Such tests have been completed for a 6-3/8-in. plate of A-533B Class II. Mill chemical analysis as determined by the supplier (Lukens Steel Co.) was

C	Mn	P	S	Si	Ni	Mo
0.20	1.39	0.11	.013	0.20	0.53	0.48

The mechanical properties of the plate are listed in Table 2. It is to be noted that the plate properties overlap the specifications for both Class I and Class II material, so that the plate may be considered as lying in the upper range of Class I material. The latter is the most commonly used material for reactor vessel construction. The plate is representative of mill practice in that it is part of a production melt conforming to steel making and rolling practices that are identical to those used by Lukens in supplying the reactor vessel construction programs.

The results of a series of 5/8-in. DT specimens are shown in Fig. 4. All fracture path orientations are RW (5) (longitudinal specimen). The solid lines are the best fit through the 5/8-in. DT data points taken from four thickness locations. Specimens using both the brittle electron beam crack starter as well as the pressed notch were used with no apparent difference in results. The scatter band of the data is shown but not the individual data points, which comprise over 100 values from two plate locations about 5 ft apart. The data indicate small differences in toughness on the upper shelf with location (as shown by the letters on the curves representing thickness locations). Also shown in Fig. 4 are photographs of the fracture faces at selected temperatures. These relate to data points which fall on the curves labeled C and D and E. Note the absence of

Table 2
Mechanical Properties of 6-3/8-in. A-533 Grade B Class II Plate Material

		YS (ksi)	UTS (ksi)	Percent Elongation	Percent R. A.
Longitudinal	Top	78. 8	98. 5	24	60. 6
	Bottom	76. 9	97. 5	20	46. 3
Transverse	Top	77. 9	99. 5	20	44. 3
	Bottom	80. 7	100. 0	24	59. 1

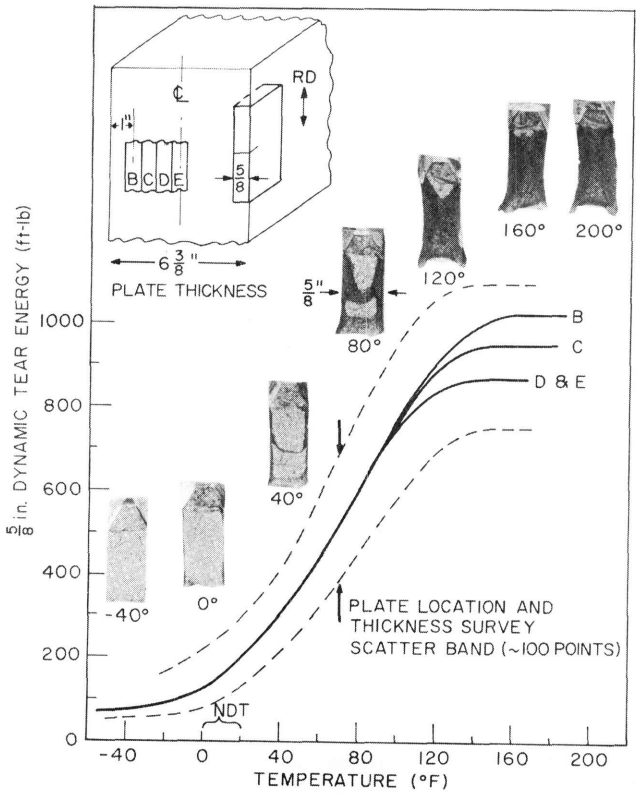

Fig. 4 - Best fit curves of the 5/8-in. DT energy values from four thickness locations of a 6-3/8-in. thick A-533B Class II steel plate. The fracture surfaces at selected temperatures are shown. All data are for the RW fracture orientation.

shear lips at 0°F and the rapid development of shear lips above the NDT region. Arrest markings are noticeable beginning with the fracture at 40°F. The arrest markings become more pronounced with increasing temperature until the initial popin through the brittle weld at 120°F is no longer accepted by the base material and the fracture reverts to full slant condition. The energy values of Fig. 4 show an order of magnitude increase from the NDT, or DT toe region, to the upper shelf, thereby reflecting the increase in toughness indicated by the fracture appearance. It is important to note that the fractures at the shelf temperature are fully ductile and require a high degree of gross plastic deformation to propagate.

Photographs of 3-in. and 6-3/8-in. fractures are presented in Fig. 5. All fracture orientations are RW except the one 6-3/8-in. fracture marked WR. The 6-3/8-in. DT fractures represent full plate thickness; the 3-in. DT specimens were obtained by splitting the broken halves of the full thickness specimens, and they therefore represent material from surface to centerline (1/2 T). The fracture appearance of the 3-in. and 6-3/8-in. specimens is closely similar to that of the 5/8-in. and 1-in. specimens (not shown) in the transition region up to 120° F. However, at 120° F, where the 5/8-in. DT specimen indicates full shear, the 3-in. and 6-3/8-in specimens still show flat regions. These flat zones were determined to be cleavage from electron fractograph studies. This behavior can be rationalized on the following basis. Assuming the size of the shear lip to depend on loss of constraint caused by proximity to a free surface, one would expect approximately the same size shear lips for all size specimens at a given temperature provided they represent an identical metallurgical condition. This appears to be true for the 5/8-in., 3-in., and 6-3/8-in. DT specimens at 120° F. The shear lips in the latter two sizes are approximately the same, being 5/16 in. Since 5/16-in. shear lips for the 5/8-in. DT specimen would cover the entire surface, it does not seem unexpected to find full shear in this size and not in the others. The tests indicate that the fracture toughness of the material is very similar from surface to center judging from the 3-in. DT specimens, which have approximately the same size shear lips on each side. As noted above, one side of the 3-in. specimen represents plate surface material, and the other represents the center plate.

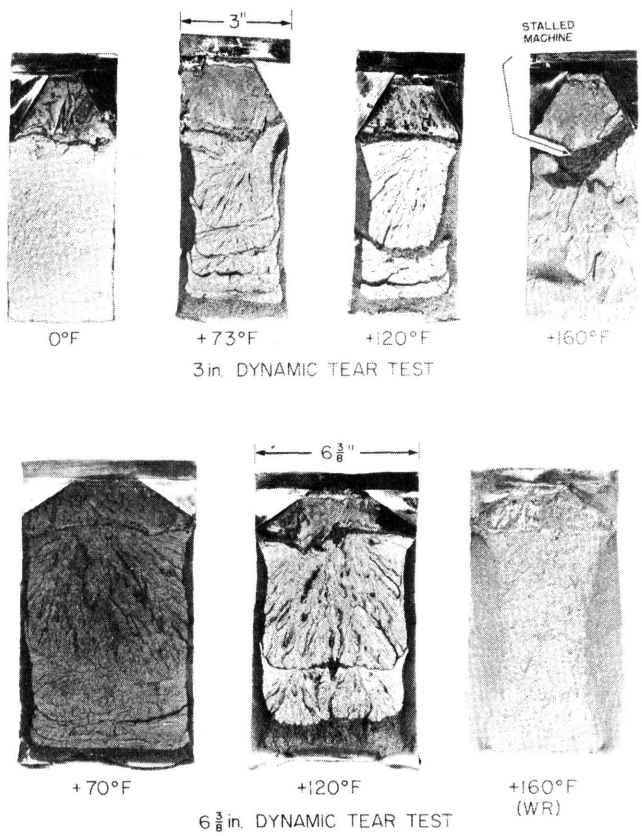

Fig. 5 - DT fracture faces for a 3-in. and 6-3/8-in. specimen from a 6-3/8-in. plate of A-533B Class II steel. All fracture orientations are RW except the one specifically marked WR.

It is important to note that the 3-in. and 6-3/8-in. DT specimens indicate that full fibrous fracture (ductile rupture) is attained only 40° F above the same behavior for the 5/8-in. and 1-in. DT specimens. The 3-in. DT test at 160° F required more energy than was available in the intermediate size testing machine; consequently, the fracture was arrested soon after initiation, as indicated. The specimen was later cooled in dry ice and broken open. The fracture initiated in a fibrous manner immediately beyond the brittle crack starter. A similar behavior was obtained for an RW 6-3/8-in. DT specimen tested at 170° F. This specimen likewise resulted in a fracture arrest which stalled the larger testing machine (Fig. 6). The specimen was not broken open at low temperature because, as could be observed in the crack opening, the fracture began in a fibrous fashion immediately beyond the brittle crack starter. Another index of the gross plastic strain conditions for the 170° F test is the severe lateral contraction which was evident at the base of the crack starter weld.

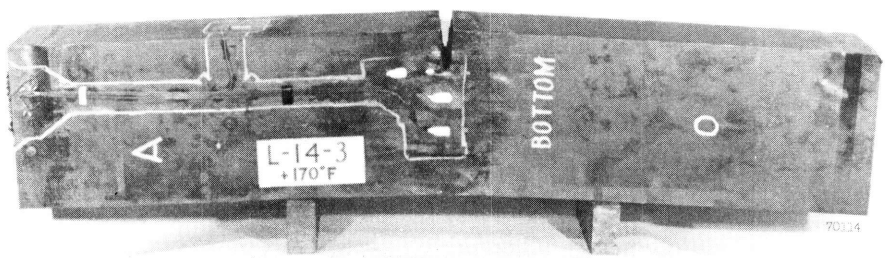

Fig. 6 - Illustration of the dramatic resistance to fracture of the 6-3/8 in. DT specimen (RW orientation) from A-533B Class II plate which stalled the testing machine at 170° F test temperature

A complete fracture for the 6-3/8-in. specimen was obtained at 160° F, shown in Fig. 5, by shifting to the transverse or WR orientation. Since material is characteristically weaker in the WR than in the RW orientation, complete fracture was obtained without stalling the testing machine. The entire fracture is fibrous with no evidence of cleavage. Note, however, that the fracture is not completely slant but contains a 3-in., flat fibrous region in the center. Severe lateral contraction of 3/16 in. was noted for both side surfaces of this specimen. The fracture conditions are obviously similar to those of the partially fractured 6-3/8-in. RW specimen shown in Fig. 6. The only difference is that the limit of the drop weight energy was exceeded for the RW test direction, and full fracture could not be obtained.

The energy values associated with the various 1-in., 3-in., and 6-3/8-in. -specimen fractures are presented in Fig. 7 superimposed on the 5/8-in. DT energy band shown in Fig. 4. These results are presented as the ratio of energy to unbroken specimen area (not including brittle weld area) in order to compare the results from different specimen sizes on a common basis. This method was used as a first approach to normalizing the data; other methods of representing the data are being considered. Based on tests of other materials, this method of representation has indicated that the thicker specimens result in higher shelf energy values on an energy/area basis. This is to be expected in view of the larger plastic enclave for the thicker specimens in the plane stress fracture region at shelf temperatures. The most important conclusion that may be derived from the Fig. 7 plots is that all specimen sizes exhibit a rapid rise in fracture energy along the same temperature interval as the 5/8-in. DT band. It is clear that the 5/8-in. specimen can be used to predict thick section behavior (i.e., the temperature range of rapid rise). All specimen sizes exhibit full fibrous fracture on the shelf requiring large amounts of plasticity to propagate the fractures. The rapid energy rise coupled with the change

in fracture appearance indicates a probable collapse of plane strain conditions and a dramatic fracture toughness transition with temperature. From results to date, it does not appear that the effects of mechanical constraint decrease fracture toughness significantly in the thickness range studied. Additional studies are underway with 12-in. A-533 B steel in connection with the AEC HSST Program, and preliminary results indicate the same trends displayed for the 6-3/8-in. material.

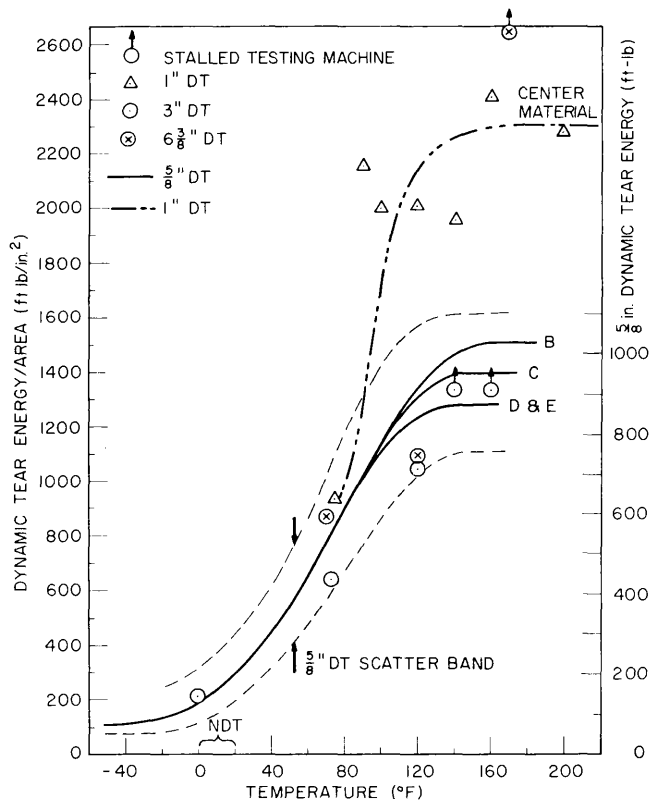

Fig. 7 - DT energy values for four different size specimens of A-533B Class II steel. The energy data are normalized to a unit of fracture surface area. The individual 1-in., 3-in., and 6-3/8-in. DT energy values plotted with respect to the 5/8-in. DT data band obtained in Fig. 4. All fractures are for the RW orientation.

The implications of these data are that a common transition from cleavage to ductile fracture occurs and that the metal refuses to separate in a cleavage mode beyond a critical temperature region. This transition represents the basic effect of controlling microscale fracture processes (cleavage mode cannot be activated) as described by Tetelman and McEvily (6). Fracture mechanics theory, which considers the metal as a continuum, recognizes changes due to temperature only in terms of the effect on yield strength. As such, there is no rationale for introducing transition effects connected to basic changes in the micromode fracture processes which are related to metallurgical and defect structure theory. The effects of increased constraint apparently apply to the static K_{Ic} measurement region below NDT, because the cleavage mode may be activated at these temperatures. However, by an increase in temperature in the order of 120°F above NDT, the cleavage mode exclusion region is reached, and restraint effects become of lesser significance. In effect, crack blunting and gross plastic strain are enforced by the refusal of the metal grains to activate cleavage fracture.

COMPARISON OF DYNAMIC TEAR AND FRACTURE MECHANICS TOUGHNESS TRENDS

The K_{Ic} data of Fig. 3 plus other data from Wessel (3, 7) for A-302B are combined in Fig. 8 with the DT data band for A-533B steel. Comparison between different materials is necessary because both types of data are not yet available for one heat of steel. It is expected that the DT energy curve will rise similarly for the A-302B as it does for the previously described A-533B plate.

A number of inferences can be drawn from Fig. 8. First, it appears that the temperature region below NDT can be adequately analyzed by linear elastic fracture mechanics if structural operation is contemplated here. Such a region can be delineated with the DT test as being below the zone where the DT test exhibits a rapid increase in energy values. Secondly, the thick section DT data (which fall in the band shown) indicate an irrepressible transition in dynamic fracture toughness in the 100 to 150° F interval above NDT. This trend appears consistent with an observed loss in constraint in the K_{Ic} tests as the NDT temperature is approached, thereby projecting a transition (rapid rise) in the K_{Ic} curve with temperature. The K_{Ic} transition occurs at a temperature below the DT transition because of the static nature of the K_{Ic} tests. The absolute accuracy of the inferred "ΔT" temperature relationships indexed to the NDT or DT toe region remains to be verified by large size K_{Ic} tests, such as the 12-in. compact K_{Ic} tests being conducted by Wessel in association with the HSST program.

The general implications of the data, irrespective of modifications as to details, are that pressure vessels constructed of 6-3/8-in. A-533B will experience a marked increase in fracture toughness due to transition effects. This thickness is common for boiling water reactor vessels. Judging from the behavior of the thick section DT specimens on the shelf, nuclear reactor plants can be operated with a conservative safety margin

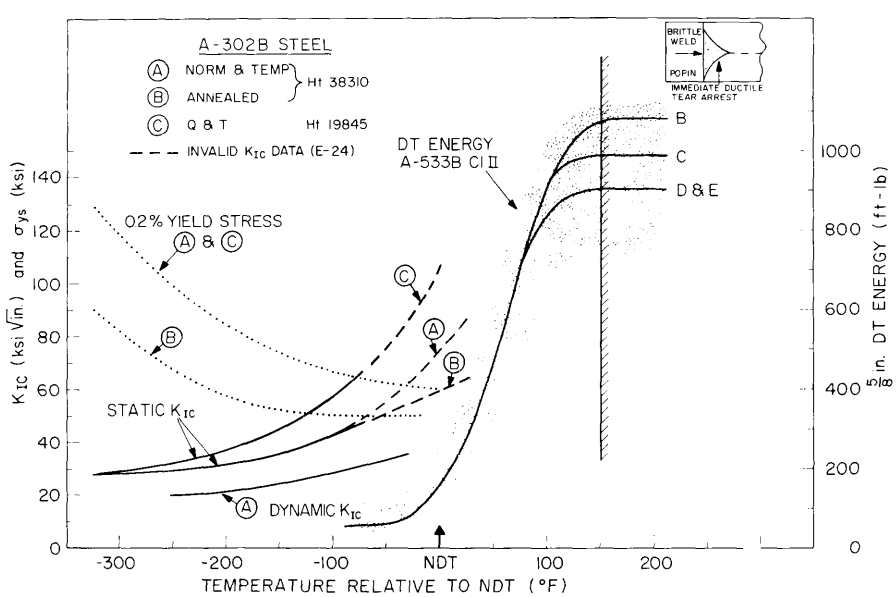

Fig. 8 - Comparison between fracture toughness behavior with respect to temperature as determined by the DT energy for A-533B steel and by fracture mechanics K_{Ic} values for A-302B steel. The NDT positions for steels B and C have been approximated from C_v data and ordinarily expected temperatures. The temperature scale is normalized by reference to the NDT temperature of the steels.

(which should include acceptability of modest plastic overloads), at temperatures in excess of 100° F above NDT provided irradiation damage does not severely drop the shelf energy. The vessels may be safe for most practical elastic loading conditions at temperatures above NDT + 60° F. Little concern need be given to flaw conditions at temperatures significantly in excess of NDT + 100° F, since through-thickness flaws would require gross plastic deformation to propagate in a plane stress mode. This type of loading in the presence of flaws of such dimensions is not expected during plant operation. It is difficult to make a case for urgency to obtain solutions based on gross strain fracture mechanics in this tough region for the above application. The prospect for such solutions, even if vigorously pursued, is many years away. Research effort should prove most rewarding if directed to the lower half of the DT transition temperature region, where the flaw size determinations are certainly much more critical, particularly for the dynamic case.

The transition region is complicated by the effects of dynamic loading. Fracture mechanics studies are being considered for this region; the importance of such studies can be inferred from the behavior of the DT test. As the statically determined K_{Ic} values sweep up in the transition region, the critical flaw sizes increase as K_{Ic}^2 (Eq. (1)) and may range from several inches to over 1 foot depending on the ratio K_{Ic}/σ_{ys} and the relative stress σ/σ_{ys}. Yet it is well known from service experience (2) that flaws much smaller than this (say an arc strike) are sufficient to trigger fracture at the NDT temperature under dynamic conditions, whereas much larger flaws can remain dormant under static conditions. This results from the decrease in K_{Ic} for the dynamic as compared to the static case. It must be emphasized that overall dynamic structural loading is not required to make the flaw behave as if it were dynamically loaded, as explained earlier.

Dynamic K_{Ic} investigations for strain rate sensitive steels have been initiated by several laboratories. These are presently restricted to the region where valid K_{Ic} numbers can be obtained with specimens no thicker than several inches. A preliminary curve of dynamic fracture toughness values (K_{Id}) has been obtained by Mager (8) for the normalized and tempered A-302B steel (heat 38310) shown in Fig. 8. These data were obtained from 1-in. compact tension specimen (CTS) tests at a loading rate or $\dot{K}$ of 10^4 ksi-in $^{1/2}$/sec and the K_{Id} numbers have dropped to 60% of their static values. From Eq. (1) it may be calculated that the critical flaw depth under dynamic loading will decrease as $(K_{Id}/K_{Ic})^2$; therefore, the critical flaw sizes under dynamic loading have decreased to 36% of their static values at the temperatures involved (below NDT). This calculation does not necessarily relate to the lower limit of dynamic flaw size since this will depend on the loading rate and must be investigated as a function of this parameter. It is important to carry the analysis up to and beyond the NDT temperature to establish the reduction in critical dynamic flaw sizes as compared to the static case since this region is more significant from an operating viewpoint than the area below NDT. In this regard dynamic K_{Ic} tests of thick specimens have been initiated by Westinghouse.

IRRADIATION EFFECTS

The effects of irradiation on A-533B steel may be expected to raise the NDT temperature and lower the upper shelf level toughness. A shelf toughness relating to huge flaws and gross plastic strain can no longer be guaranteed when the shelf level is sufficiently lowered. This effect is analogous to the "strength transition" (1) for high strength steels which display lower shelf toughness values with increasing yield strength. Reduced toughness can be readily determined from the fracture appearance as well as specimen energy absorption. Full slant fractures for 5/8- and 1-in. specimens, then, are not obtained at the shelf though the entire fracture surface displays dimpled rupture from a fractographic viewpoint. The DT analysis for the case of decreasing shelf levels is discussed in terms of Ratio Analysis Diagram procedures in Ref. 1. Briefly, present estimates indicate that

a shelf level DT energy value below 2000 to 2500 ft-lb for the 1-in. DT specimen (or approximately 400 to 500 ft-lb for the 5/8-in. DT) would place the irradiated materials of 80 to 100 ksi yield strength in the region of questionable toughness requiring more refined calculative techniques.

GENERAL CONSIDERATIONS

As additional data are obtained and analyzed it is becoming evident that low strength steels which feature very high fracture toughness in plane stress fracture (on the temperature transition "shelf") must be treated by a fracture analysis method based on transition temperature concepts. The behavior of such steels is entirely different from that of ultrahigh strength steels which have been investigated extensively by plane strain, linear elastic fracture mechanics. The latter type do not show a meaningful temperature transition to mechanical conditions of elastic-plastic or gross plastic strain except in relatively thin sheet form. When such a transition is developed, the fracture propagation resistance is relatively low compared to that of the low strength steels. The microfracture processes for the ultrahigh strength steels are not controlled by a dramatic change from cleavage to ductile rupture; the fracture processes involve intragranular and other modes of microcrack growth and coalescence.

The importance of the transition temperature aspects of the "high shelf," low strength steels is the suppression of plane strain, linear elastic fracture mechanics conditions by basic microfracture events which cannot be avoided by the imposition of mechanical constraint. The limiting transition temperature at which the ductile micromode fracture processes become dominant should be determined under conditions of dynamic fracture.

The DWT-NDT and the toe region of the DT test transition curve have a common meaning as well as a common reference temperature. The subject reference temperature indicates that the transition temperature range has been reached and that the dramatic events of increasing dynamic fracture toughness to full ductility will follow in the subsequent interval of 120 to 150° F. The best available evidence to date indicates that the static K_{Ic}, plane strain transition (rapid rise in plane strain fracture toughness) begins to evolve in the interval of approximately 100° F below the NDT-DT toe region reference temperature. The static K_{Ic} transition then appears to project through the subject reference temperature with temperature dependence features (K_{Ic} curve slope) which require very large K_{Ic} test specimens for definition. Whether this transition culminates to a limiting condition of a K_{Ic} shelf or to plane stress K_c conditions will be indicated by these large tests. From a practical engineering viewpoint the thick section DT tests serve to index a temperature region above the reference temperature for which gross strain fracture conditions are attained irrespective of applied loading rate.

A subtle but very important consideration emerges from the fact that fracture mechanics as presently evolved is based primarily on the analytical treatment of the plane strain condition. Calculations of mixed-mode or plane stress K_c conditions represent approximations of uncertain value, except as to order of magnitude. In this regard, consideration of transition temperature effects lead to the following possible conclusions regarding the applicability of linear-elastic fracture mechanics theory:

1. It is not currently applicable in the elastic-plastic or gross strain regimes and should not be used.

2. In the interim that elastic-plastic or gross strain fracture mechanics may be evolved, use of K_c approximations is justified on the basis of retention of a common language. In this context, definitions of "effective" high values of K_{Ic}/σ_{ys} ratios serve equally well as definitions of approximations. For conditions of high fracture toughness the accuracy of the effective ratio or of the approximated K_c value is of minor consequence. The fracture mechanics "K" language then simply serves to define fracture-safe conditions.

An example of the use of the K_{Ic}/σ_{ys} ratio approximation is provided by extrapolation of a K_{Ic} temperature dependence curve such as depicted in Fig. 8. As the transition temperature range is entered, valid K_{Ic} values for a specified large thickness may not be obtainable beyond some limiting temperature (e.g., NDT or NDT + 50° F). Extrapolation of the indicated trends will then infer high K_{Ic}/σ_{ys} ratios and correspondingly large critical flaw sizes. While such extrapolations may be declared "invalid," these are nevertheless useful in defining that plane stress conditions will dominate for the section size of interest. Similarly, if dynamic fracture propagation can be developed for a thin specimen containing a small flaw (say DWT-NDT test) the effective K_{Id}/σ_{yd} ratio is obviously low. Approximations of the static K_{Ic}/σ_{ys} and dynamic K_{Id}/σ_{yd} ratios provide for useful comparisons, despite the approximations involved. As static and dynamic fracture toughness measurements are made and correlated to NDT or DT data, useful indices may be expected to emerge in terms of "K" language.

The alternatives to the described correlation and extrapolation procedures are to declare nonplane strain fracture mechanics as nonexistent and to fall back on classical transition temperature fracture tests correlated to prototype structural performance tests. In either case the process of interpretation of fracture safety for materials of high fracture toughness depends on correlation or direct observation of the fracture modes involved. If these facts are understood, it will be evident that fracture safety for steel structures depends on metallurgical factors and that the mixed-mode or plane-stress conditions are the goals to be attained for structures intended for critical applications. In effect this is a restatement of the transition temperature approach, irrespective of the language used to define fracture toughness.

SUMMARY

Classical transition temperature concepts have been based on evidence that the plane strain to plane stress fracture mode transition is fully developed over a relatively narrow temperature range. Fracture mechanics theory has suggested that increased section size provides sufficient mechanical constraint for the retention of plane strain conditions through the transition temperature range. Evidence is now presented, based on Dynamic Tear (DT) tests of thick section reactor grade steels (A-533B), that the plane strain to plane stress transition is not eliminated. Under conditions of dynamic fracture toughness testing, the transition is correlated to a basic change in microfracture modes from cleavage to ductile rupture of the individual grains. The dynamic fracture transition of thick section steels can be characterized by small inexpensive DT test specimens because the temperature region at which ductile rupture becomes the only microfracture mechanism is common to thin and thick section material. The microfracture mode transition has an established scientific base in defect structure theory.

These findings are of major importance to concepts of fracture-safe design, from engineering as well as basic considerations. The most important engineering deduction is that temperatures in excess of NDT + 100° F, for high shelf toughness material, provide a conservative margin of safety against fracture. The primary inference to analytical aspects of fracture mechanics is that extensions of linear elastic analyses to

cover the elastic-plastic and finally the gross strain modes must be evolved for the transition range. The DT test serves to define the temperature range of transition from linear elastic to the gross strain modes and, therefore, the nature of the required analytical treatment for these regimes.

The temperature region significantly below NDT can be analyzed by linear elastic fracture mechanics based on K_{Ic} test specimens of practical size. On the basis of existing evidence, the gross strain regime at the shelf level temperatures does not require refined flaw size calculations since flaws of the order of the plate thickness still require plastic deformation loading to propagate in this region. The transition region between NDT and shelf temperatures requires further definition by fracture mechanics tests of large size to determine the extent of applicability of the linear elastic theory. Beyond the limits of linear elastic conditions, quantitative treatment must wait for new solutions to the mechanical conditions involved in elastic-plastic and gross strain fracture, which unfortunately cannot be expected in the near term.

The importance of considering fracture initiation, in terms of limiting dynamic fracture toughness values, is emphasized by analyses of available data for reactor pressure vessel steels. Dynamic crack tip conditions reduce the size of critical flaws from much larger values controlled by static initiation. Evidence for this behavior in the transition region is brought out by the small Drop Weight and DT tests and is corroborated by service experience. It remains an urgent task of fracture mechanics to explore these dynamic effects and thereby extend the useful limit of linear elastic theory into the transition region.

In some cases it is necessary to make immediate decisions concerning fracture-safe operating limits for existing and planned structures. In this regard the importance of the fact that the 6-3/8-in. DT tests, as well as preliminary 12-in. DT results, imply a strong temperature transition in dynamic fracture toughness becomes evident. A large increase in section size (5/8 in. to 6-3/8 in. and apparently to 12 in.) does not eliminate or increase the transition temperature range appreciably. Therefore, reactor pressure vessels could be assumed to survive plastic overload in the presence of large flaws with a conservative margin of safety at or near the shelf level region, i.e., at temperatures in excess of 100° F above the NDT point. The vessels are considered safe for most practical flaw size-elastic stress conditions above NDT + 60° F. It is important to note that the increased fracture toughness due to the transition is so marked that the temperature increment from NDT + 60° F to over NDT + 100° F introduces a high degree of conservatism in assessing probabilities of failure. It also permits consideration of fracture safety under conditions of gross plastic strain of the vessel. In effect, these findings document the validity of the FAD for steels of large section size.

If attainable, the gross strain condition should be used for reactor and other pressure vessels for critical service. In the event that this is not possible, because of irradiation shifts in NDT and a severe lowering of upper shelf level toughness, then a refined approach based on fracture mechanics is required. This approach should incorporate parameters relating to dynamic toughness, which is the limiting condition to be considered in safety analyses.

ACKNOWLEDGMENT

The authors acknowledge the efforts of the members of the Strength of Metals Branch in obtaining the test data.

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Security Classification

DOCUMENT CONTROL DATA - R & D

(Security classification of title, body of abstract and indexing annotation must be entered when the overall report is classified)

1. ORIGINATING ACTIVITY *(Corporate author)* Naval Research Laboratory Washington, D. C. 20390	2a. REPORT SECURITY CLASSIFICATION Unclassified 2b. GROUP

3. REPORT TITLE
DYNAMIC TEAR TEST DEFINITION OF THE TEMPERATURE TRANSITION FROM LINEAR ELASTIC TO GROSS STRAIN FRACTURE CONDITIONS

4. DESCRIPTIVE NOTES *(Type of report and inclusive dates)* This report completes one phase of the problem. Work on other aspects of the problem is continuing.

5. AUTHOR(S) *(First name, middle initial, last name)*
F.J. Loss and W.S. Pellini

6. REPORT DATE November 29, 1968	7a. TOTAL NO. OF PAGES 20	7b. NO. OF REFS 8
8a. CONTRACT OR GRANT NO. NRL Problem M01-18 b. PROJECT NO. RR 007-01-46-5420 c. d.	9a. ORIGINATOR'S REPORT NUMBER(S) NRL Report 6787 9b. OTHER REPORT NO(S) *(Any other numbers that may be assigned this report)*	

10. DISTRIBUTION STATEMENT
This document has been approved for public release and sale; its distribution is unlimited.

11. SUPPLEMENTARY NOTES	12. SPONSORING MILITARY ACTIVITY Department of the Navy (Office of Naval Research), Washington, D. C. 20360

13. ABSTRACT

Transition temperature concepts for fracture-safe design have been based on the relatively narrow temperature range evidenced by the fracture-mode transition from plane strain to plane stress. Fracture mechanics theory has suggested that large increases of section size should provide sufficient mechanical constraint for retention of plane strain conditions through the transition temperature range. Recent investigations based on Dynamic Tear (DT) tests of thick-section reactor-grade steels (A-533B) have provided clear evidence that the plane strain to plane stress transition is not eliminated. These findings are of major consequence to concepts of fracture-safe design and show that temperatures significantly above NDT + 100° F can provide a conservative margin of safety, even for the condition of very large flaws and plastic overload. The DT test, as conducted using a small specimen, defines the temperature range of transition from linear elastic to gross strain mechanical conditions of fracture and, therefore, the required analytical treatment for flaw size-stress calculations. Fracture mechanics concepts are brought into consonance with transition temperature concepts of long standing, and the importance of considering fracture initiation in terms of limiting dynamic fracture toughness values is emphasized.

DD FORM 1 NOV 65 1473 (PAGE 1)

S/N 0101-807-6801

Security Classification

Security Classification

14. KEY WORDS	LINK A ROLE	LINK A WT	LINK B ROLE	LINK B WT	LINK C ROLE	LINK C WT
Dynamic Tear test						
Fracture mechanics						
Steel						
Structural safety						
Transition Temperature						
Fracture-safe design						
A-533B steel						
Pressure vessel material						
Nuclear reactor material						

DD FORM 1 NOV 65 1473 (BACK)
(PAGE 2)

Security Classification