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FRACTURE TOUGHNESS OF REACTOR PRESSURE VESSEL STEELS

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By

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ABSTRACT

The effect of loading rate on fracture toughness of A533 steel was studied. Dynamic loading was found to elevate the yield strength to an extent that resulted in a significant decrease in the specimen size required for measuring valid plane strain toughness. The direct influence of rate on K_{Ic} , however, was modest.

The minimum specimen dimensions suggested by the ASTM E-24 Committee for high strength metal appears to be too conservative for this intermediate strength steel. Dimensions less than half the recommended ones were able to separate size-dependent from size-independent toughness values.

The effect of specimen thickness at each of the testing temperatures could be described by the Irwin $K_{Ic}-K_{Ic}-\beta_c$ relationship. Using this expression, a K_{Ic} value for infinitely thick specimens was defined, and this value was consistent with the lower bound of arrest toughness for tests made at +75, 0, and -75°F.

INTRODUCTION

This program was undertaken in order to develop a technique for measuring K_{Ic} of intermediate strength steels at room temperature and above with specimens small enough to be irradiated. Although linear elastic fracture mechanics has been largely confined to high strength, low toughness materials, there is no conceptual barrier to extending this discipline to the testing of such lower strength, tougher materials. There is a serious practical obstacle, however: as the toughness increases relative to the yield strength, larger specimens are required, both to preserve the validity of the elastic analysis through which K_{Ic} is defined, and to obtain a plane strain stress state at the crack tip.

Developing a testing technique for measuring fracture toughness values which could be unequivocally identified as K_{Ic} involved first, the selection of a specimen type and testing procedure which yielded the maximum K_{Ic} measurement capacity for the specimen size. A crack-line loaded specimen offered the most promise; and in particular, the contoured double-cantilever beam (CDCB) specimen was selected. It not only was small, but also offered the possibility of making several independent measurements on one specimen; and, specifically, it made possible (or at least greatly facilitated) the measurement of crack arrest toughness. Having selected a type of specimen and test procedure, it was necessary to ascertain the minimum dimensions which would yield valid K_{Ic} measurements. The validity of the data was established experimentally on the material of interest (A533 steel). The work of ASTM Committee E-24 directed toward the establishment of standard procedures for K_{Ic} testing provided some guidance in selecting the criteria for interpreting test

results. The recommendations of the committee, however, are based on testing experience with higher strength materials. Fracture toughness testing of low strength materials must be regarded as a new area in which the "rules" have yet to be established. The recommendations of ASTM E-24, which serve formally to define K_{Ic} in terms of standardized testing techniques cannot be adopted without question for the testing of mild steels.

The culmination of this program was the demonstration of the validity of measurements of the room temperature crack initiation and arrest toughness of A533 steel with one-inch and two-inch thick specimens. The initiation data were not necessarily valid K_{Ic} numbers according to the criteria of ASTM E-24 Committee (1), and, of course, there is no standard method of measuring arrest toughness. The data, however, do define meaningful fracture toughness values for this steel at room temperature and also at 0° and -70° F. The values are meaningful, first, because they can be used to predict the fracture behavior of sections thicker than the actual specimens tested, i. e. they serve to define a K_{Ic} value in the original sense of a minimum thickness independent toughness parameter. Second, the data embrace the effect of strain rate on the fracture toughness of the A533 steel.

The results obtained in this program have already been presented in interim reports. Therefore, rather than present a chronological review of the data this report will discuss the meaning of the data in terms of their contribution to the general problem of fracture toughness measurement of mild steel.

BASIC CONCEPTS

Definitions of $\mathcal{G}$, K , r_y and K_{Ic}

In linear elastic fracture mechanics the stress field at the tip of a crack is described by the stress intensity factor, K . A single parameter characterization is possible because the two-dimensional elastic stress field sufficiently close to the crack is dominated by a mathematical singularity, and the stress distribution is the same regardless of how the remote loads are applied to the structure containing the crack. The stress intensity factor is directly proportional to the applied load: but the proportionality constant, which has dimensions of $(\text{length})^{-3/2}$, depends on the geometry of the structure containing the crack and, in particular, on the crack length. The relationship between K and the applied load or nominal stress as a function of the crack length is available for different specimens and structural configurations, for example in ref. (2).

The crack extension force (or strain energy release rate), $\mathcal{G}$, measures the decrease in elastic stress field energy per unit increase in crack area. For plane stress elastic deformation, $\mathcal{G}$ and K are related by:

$$\mathcal{G} = K^2 / E \quad (1)$$

where E is Young's modulus. Thus, within the limits of two-dimensional linear elasticity, K and $\mathcal{G}$, are equivalent representations of the crack tip stress field. In terms of the applied load, P , the crack extension force for a through-the-thickness crack in a plate of thickness, B , is given by

$$\mathcal{G} = \frac{P^2}{2B} \frac{dC}{da} \quad (2)$$

where C is the compliance (deflection per unit load) of a structure or specimen containing a crack of length a . The dependence of β on the applied load, therefore, can be obtained experimentally by measuring the compliance of a specimen as a function of crack length. Combining eqs. (1) and (2) gives

$$K = P \left(\frac{E}{2B} \frac{dC}{da} \right)^{1/2} \quad (3)$$

This equation expresses the proportionality constant between K and P in terms of the specimen compliance, which can be determined experimentally.

The stress intensity factor, K , is defined in terms of a linear elastic model; in real materials, however, plastic deformation occurs at the crack tip. If the size of the crack tip plastic zone is small, the validity of K as a meaningful parameter is preserved since the plastic zone is contained wholly within the portion of the elastic stress field dominated by the crack tip singularity. The surrounding elastic field (which is characterized by the stress intensity factor) then controls the size and shape of the plastic zone. The effect of the plastic zone is primarily to shift the origin of the elastic stress field, and this effect can be taken into account by adding a correction to the crack length. If, on the other hand, the plastic zone is large enough to be influenced significantly by the gross state of stress, the stress intensity factor ceases to be a meaningful parameter for characterizing fracture behavior. To preserve the applicability of the linear elastic analysis, the lateral dimensions of the specimen--length, width, crack length, unbroken ligament size--must be large with respect to the plastic zone size. For different specimens,

more or less formal rules governing specimen dimensions have been set up on the basis of testing experience of high strength materials. One of the purposes of this study, as stated in the Introduction, is to establish these minimum dimensions for intermediate strength steel.

An accurate description of the crack tip plastic zone must be based on an elastic-plastic stress analysis. An approximate description of the size and shape of the plastic zone, however, can be obtained by applying one of the yield criteria to the linear elastic stress field. In plane stress (i. e. the through-the-thickness stress is zero) the formally defined "radius" of the crack tip plastic zone, r_y , is given as

$$r_y = \frac{1}{2\pi} \left(\frac{K}{\sigma_Y} \right)^2 \quad (3a)$$

where σ_Y is the yield stress of the material. For plane strain (i. e. through-the-thickness strain is zero) the plastic zone is thought to be about one-third this size. Obviously, the size (and shape) of the formally calculated plastic zone depends on whether the stress state is assumed to be one of plane stress or of plane strain, and in practice the fracturing process is strongly dependent on the degree of constraint through the thickness. At the specimen surfaces deformation occurs under a state of plane stress; the plastic zone is large, and separation commonly occurs on planes which are oblique to the specimen surfaces. In a thin specimen, the entire fracture surface may be oblique to the specimen surfaces. In a thicker specimen the oblique fracture may appear as shear lips extending only a limited distance in from the sides of the plate. In the thicker specimen, a state of plane strain deformation is approached on proceeding

inward from the specimen edges. The plane strain plastic zone is smaller, and separation occurs, macroscopically, on a plane perpendicular to the specimen edges. The change in stress state from plane stress to plane strain must be gradual; however, the line of demarcation between slant fracture (shear lips) and flat fracture is typically quite sharp. It is quite common, therefore, to associate the shear lips with plane stress fracturing and the flat central portion with plane strain fracturing. The existence of a macroscopically flat fracture surface, however, does not necessarily prove that fracture occurred under plane strain conditions.

Fracture toughness is expressed in terms of a critical value of the stress intensity factor, K_c , corresponding to the onset of rapid fracture or the occurrence of detectable slow crack extension. The measured value of K_c depends on the specimen thickness; as the proportion of flat to slant fracture increases, K_c decreases. With a sufficiently thick plate specimen, the plane strain center portion dominates fracture initiation, and further increases in thickness produce no further reduction in K_c . This minimum value of toughness is denoted as the plane strain fracture toughness, K_{Ic} .

Size Requirements for Measuring Valid K_{Ic}

The measurement of a valid K_{Ic} requires that the plastic zone size be small enough relative to the plate thickness (B) that a central portion of the crack front deforms in plane strain. Since the plastic zone size at the critical load is proportional to the quantity $(K_c / \sigma_Y)^2$ as indicated in eq. 3, a thickness criterion should have the form

$$B = \alpha (K_c / \sigma_Y)^2 \quad (4)$$

where α is an experimentally determined parameter which should not differ much for different materials. Based on tests of high strength materials, a value of $\alpha = 2.5$ is currently recommended by ASTM Committee E-24 as a criterion for the validity of a K_{Ic} measurement. Whether the same value of α is valid for tougher, low strength materials has not been established. The simple assumption that the plastic zone size is proportional to $(K_c / \sigma_Y)^2$ ignores any effect of strain hardening on the plastic zone.

The question of specimen size is particularly urgent for pressure vessel steels. For operating temperatures significantly above room temperature the low value of yield stress and the high (expected) plane strain toughness, indicate that the measurement of K_{Ic} would require a specimen several inches thick. Because of the heavy sections employed in pressure vessels, brittle crack propagation under plane strain conditions might be possible in service. The testing of specimens in the full thickness of a pressure vessel wall is too expensive to be practical, especially for irradiated material. Thus, it is necessary to extrapolate the results obtained on thin specimens to predict the fracturing behavior of thick sections and/or to extrapolate toughness measurements at low temperatures to predict behavior at the operating temperature. To date, most of the direct fracture toughness testing of mild steels has been confined to temperatures well below room temperature. If it proves possible to measure a valid K_{Ic} with reasonable sized specimens at the temperatures of interest, this value would apply, of course, to thicker sections. If not, the extrapolation of the results to thicker sections would involve a greater uncertainty than an extrapolation of valid K_{Ic} data from lower temperatures. These considerations illustrate the desirability of pressing specimen capacity for K_{Ic} testing as far as possible.

Effect of Loading Rate on Size Requirements

A feature of particular concern in the testing of mild steels is their strain rate sensitivity. The rate sensitivity has a direct bearing on specimen size requirements by virtue of the rate dependence of yield stress. Fig. 1 shows 0.2 percent offset yield stress of A533 steel measured over a range of loading rates. (The rates are expressed in terms of the reciprocal of the loading time up to the yield point.) The yield stress at -70°F was approximately 5 to 7 psi greater than at 0°F over the whole range of loading rates. At both of these temperatures the increase in yield strength was relatively modest up to 0.1 to 1.0 sec^{-1} , and then increased at about 6 ksi per ten-fold increase in loading rate. The room temperature data showed less rate sensitivity over the same range of loading rates. At the highest rates, however, the room temperature curve also appears to approach a slope of 6 ksi per order.

According to eq. 4, the thickness required to measure a particular value of K_{Ic} should decrease inversely as the square of the yield stress. Thus, by conducting fracture toughness at high loading rates, the attendant increase in yield strength due to the more rapid formation of the plastic zone should decrease the required specimen size. The effect is illustrated quantitatively in Fig. 2, which is based on the data of Fig. 1. In constructing Fig. 2 it was assumed that a thickness requirement like eq. (4) was applicable, and that a thickness B_0 was required to measure a particular value of K_{Ic} in a static room temperature test. The figure shows the relative thickness, B/B_0 , required to measure the same value of toughness at the higher loading rates and lower temperatures for which the yield stress data were obtained. It can be seen that a significant reduction in required thickness can be obtained by dynamic testing (at rates of the

order of 10 milliseconds loading time) as compared to a static rate (100 seconds loading time). The effect of such a rate change at room temperature is greater than the effect of reducing the test temperature to -70°F for a static test.

Effect of Loading Rate on K_{Ic} and Significance of Arrest Toughness

Because loading rate affects the plastic flow behavior of materials, it would not only be expected to contribute to specimen size requirements, but since fracturing occurs within the plastic zone, it would also be expected to influence the measured value of K_{Ic} as well. Indirect measurements of K_{Ic} as a function of loading rate, as presented by Krafft and Irwin (3), suggest that the effect of rate would be large. Their data shown in Fig. 3 indicate that the dynamic toughness might be as low as half the static value. In attempting to collect dynamic data, there is a practical limit to the rate at which specimens can be loaded, without encountering experimental difficulties because of inertia and a dynamic response of the load cell and other measuring instruments. Crack tip behaviors under such rapid loading conditions can be simulated, however, by measuring the toughness of running cracks. The crack-tip plastic zone forms extremely rapidly over the interval in which the crack front jumps from its original to its final location. This jump is marked by two values of instability: the first one being associated with crack initiation and the second one with crack arrest. On applying a load to the cracked member, the size of the crack tip plastic zone at plane strain instability decreases as the loading rate is increased even if K_{Ic} is unchanged by rate. The plastic zone size associated with the running crack, however, is small even when compared with high loading rates. Hence, measurements of the arrest instability yields a value of dynamic toughness,

and since it is associated with a free running crack, it would be expected to be the specific value of minimum toughness for the testing temperature. The arrest toughness, K_{Ia} , might also be expected to be less dependent on testing conditions than K_{Ic} . The latter is a function of the method used for precracking, e. g. the amplitude of ΔK used for fatigue precracking. The influence of precracking technique may result from differences in crack smoothness, size of the plastic zone, or residual stress at the crack tip. All three of these should be solely dependent on material for arrest toughness, making the latter behavior more invariant than K_{Ic} .

Use of Side Grooves to Promote Plane Strain Fracturing

A through-the-thickness variation in toughness caused by the changing stress state is an inevitable feature of fracture toughness testing. The use of side grooves alters the through-the-thickness variation. Whether or not side grooves can (or should) be used to reduce the thickness required to measure a valid K_{Ic} is not settled. Clearly, side grooves restrict plastic flow at the plate surfaces; and, if the grooves are not too deep or sharp, they should have no effect on the stress state near the center thickness. General experience has shown that side grooves may eliminate shear lips and increase the abruptness of instability, i. e. the sharpness of the "pop-in" on the load deflection record. Fig. 4 shows some effects of side grooves on the fracture toughness of 7075-T651 aluminum alloy tested early in this program. For fractures initiated from arrested fast cracks, in 1/4, 1/2, and one-inch thick plates, the same toughness was observed for all depths of side grooves, and the fracture surfaces showed no shear lips. When specimens without grooves were tested in the same manner, the fracture surfaces had shear lips, and the toughness was greatly increased and dependent on the thickness. How-

ever, for fracture initiation from fatigue cracks, ungrooved specimens gave the same toughness values as did the (unfatigued) face grooved specimens. These results indicate that the side grooves produced a more uniform response across the crack front and that the grooves had no effect on the measured toughness, but they do not prove that side grooves could reduce the thickness required to measure K_{Ic} (from a fatigue crack).

The fact that side grooves alter the through-the-thickness stress distribution would suggest that the use of specimens with face grooves requires a three-dimensional stress analysis. Side grooves do not introduce the three-dimensional element into the fracture problem; it is implicit in the non-uniform response across the crack front. In ungrooved specimens, a K_{Ic} value can be obtained when a central segment of the crack front "pops-in" independent of the edge portions. Provided that crack motion is initiated at the center thickness, grooved specimens should give an equally valid K_{Ic} measurement. In both cases, the minimum thickness which allows measurement of K_{Ic} must be ascertained experimentally. Whether or not the use of side grooves can lower the minimum thickness by reducing the relative influence of the edge regions which deform in plane strain deserves investigation.

TESTING PROCEDURE

The testing procedure developed, as a part of this program, for measuring both initiation and arrest toughness with reasonably small specimens has been discussed previously in some detail (4)(5). Although these procedures will not be reviewed here, two specimen sizes, as identified by "m" number, were used, and it is helpful to compare their dimensions as shown in Fig. 5. Essentially all of the data were collected

on $m = 3$ specimens, but because these were relatively large for irradiation studies, some tests were also conducted in which the toughness of a plate was measured with both types of specimens to be certain that they yielded identical data.

FRACTURE TOUGHNESS TESTS RESULTS

Fracture toughness tests were conducted at room temperature, 0° and 70° . The results obtained at room temperature are shown in Fig. 6. A few specimens with deep side grooves ($b_n/b = 0.50$) were tested; however, the results of these tests were not distinctly different from results on specimens with $b_n/b = 0.75$ which comprise the remainder of the data. Rather, the measured values of initiation toughness fell into distinct groups based on the gross specimen thickness, b .

Two-inch thick specimens tested over a range of loading rates from 1.0 sec.^{-1} to 100 sec.^{-1} and one-inch specimens tested at 100 sec.^{-1} and at 1000 sec.^{-1} together indicated a single level of initiation toughness of $85\text{-}90 \text{ ksi-}\sqrt{\text{in.}}$ over a total range of testing rates which spans about four orders of magnitude. The absence of a rate and thickness dependence in these data suggests that $85\text{-}90 \text{ ksi-}\sqrt{\text{in.}}$ is a meaningful room temperature toughness value.

One-inch thick specimens tested at loading rates less than about 100 sec.^{-1} showed a progressive increase in toughness from the $85\text{-}90 \text{ ksi-}\sqrt{\text{in.}}$ level up to $115 \text{ ksi-}\sqrt{\text{in.}}$ at the slowest testing rate. This trend must reflect an increasing departure from plane strain caused by the progressive lowering of yield strength with decreasing loading rate.

The arrest toughness values are plotted directly below the corresponding initiation values. It should be borne in mind, however, that the

strain rate governing crack arrest is determined by the crack propagation velocity, not by the loading rate which has immediate significance only for crack initiation. Hence, there should be no dependence of arrest toughness on the reciprocal loading time. This expected behavior was confirmed by the tests on two-inch thick specimens and the one-inch tests at rates greater than about 5.0 sec.^{-1} ; all of these data defined a single arrest toughness level of $75\text{-}80 \text{ ksi}\sqrt{\text{in.}}$ which is of the order of ten percent less than the initiation toughness. Arrest toughness values for the slowest tests on one-inch thick specimens were, however, distinctly higher than the level defined by the other tests. Since crack initiation in these tests clearly did not occur under conditions approaching plane strain, it may be conjectured that the propagating cracks were arrested before a plane strain situation could be established.

Considered as a whole, the room temperature data define a rate independent level of initiation toughness ($85\text{-}90 \text{ ksi}\sqrt{\text{in.}}$) and single level of arrest toughness ($75\text{-}80 \text{ ksi}\sqrt{\text{in.}}$). One-inch thick specimens gave the same results as two-inch specimens, provided that the one-inch specimens were tested at a rate fast enough to avoid an excessive departure from plane strain.

Two groups of tests were run at 0°F . The specimens in the first group were fatigue cracked at room temperature; the specimens in the second group were fatigue cracked at lower temperatures. The data from the first group are shown in Fig. 7. The plot suggests an average value for initiation toughness of about $70 \text{ ksi}\sqrt{\text{in.}}$; however, there was considerable scatter above and below this value. There was no clear tendency for the test results from two-inch thick specimens to give lower toughness values than the thinner specimens. Within the scatter, there was no

obvious systematic dependence on loading rate. The arrest toughness values also showed considerably more scatter than was observed in the room temperature tests; however, there did seem to be a fairly well defined lower bound of arrest toughness at $55-60 \text{ ksi}\sqrt{\text{in.}}$

On the assumption that some of the scatter in the above tests might be attributable to the fatigue cracking practice, a second group of specimens were precracked at a lower rate at a temperature between -20 and -60°F . The results obtained with these specimens are plotted in Fig. 8. The initiation toughness values in this group were on the average distinctly lower than in the first group, but the results defined an arrest toughness of approximately $60 \text{ ksi}\sqrt{\text{in.}}$ in agreement with the lower bound value observed in the first group. Thus, the principal effect of the more conservative fatigue cracking practice was to reduce the initiation toughness to values more near to the arrest, but to leave the crack arrest toughness unchanged. Although there are not enough data from the second group to establish a definite statistical trend, the behavior is somewhat suggestive of the room temperature results. The two specimen thicknesses appear to have given the same toughness at high loading rates. As the loading rate was decreased, there was a tendency for the toughness measured with one-inch thick specimens to increase, while at the lowest loading rates, two tests on two-inch thick specimens gave quite low values.

The data obtained at -70°F are plotted in Fig. 9. Initiation toughness values measured on one-inch thick specimens showed a mild strain rate dependence for loading rates ranging from 0.01 to 100 sec.^{-1} . Two points at higher loading rates obtained in drop weight tests showed somewhat elevated values. One test with a two-inch specimen at 0.001 sec.^{-1} loading rate gave a toughness value very slightly below the trend line

established by tests on one-inch specimens. All measured arrest toughnesses were in the range 55-60 ksi- $\sqrt{\text{in.}}$ which is approximately the same level that was observed at 0°F.

The tests conducted at -70°F should represent a closer approach to plane strain fracture conditions than the tests at the higher temperatures. In addition they extended over the greatest range of loading rates and showed the least scatter. The initiation toughnesses were only moderately dependent on loading rate, and the arrest toughness was only slightly below the initiation toughness. Approximately the same level of arrest toughness (60 vs. 55-60 ksi- $\sqrt{\text{in.}}$) was measured at 0°F as at -70°F and the results were again essentially independent of loading rate.

All of the data in Figs. 6 to 9 were collected on specimens contoured to $\underline{m} = 3$ which requires a blank size of approximately four by six inches (see Fig. 5). Smaller specimens would be desirable for irradiation studies so that the toughness obtained on $\underline{m} = 8$ specimens, requiring a blank size of approximately 2.4 by 4 inches were compared with that obtained on $\underline{m} = 3$ specimens. Unfortunately, a new plate of A533 steel had to be used for this study so that tests on the $\underline{m} = 3$ specimens had to be repeated. To be certain that valid K_{Ic} values would be measured, the tests were conducted at -70°F. Both slow (0.1 sec.⁻¹) and fast (100 sec.⁻¹) tests were made, and as shown in Fig. 10, both specimen sizes gave the same results. The scatter at the lower loading rate, however, was much higher than that at the high rate. Particularly surprising was the very low toughness obtained on this plate as compared with that shown in Fig. 9 for another lot of the same heat of steel.

In summary, a single level of arrest toughness was fairly well defined at each of the three test temperatures. At room temperature the

arrest toughness was 75-80 ksi- $\sqrt{\text{in.}}$, at 0°F and -70°F the arrest toughness was 55-60 ksi- $\sqrt{\text{in.}}$. At room temperature the initiation toughness, defined by tests on two-inch thick specimens and by the higher rate tests on one-inch thick specimens was 85-90 ksi- $\sqrt{\text{in.}}$. The initiation toughness values measured at 0°F ranged from just above the arrest level of 55-60 ksi- $\sqrt{\text{in.}}$ up to nearly 90 ksi- $\sqrt{\text{in.}}$. Specimens fatigued at reduced temperatures generally showed initiation values at the low end of the range. At -70°F there was a very small but systematic variation of initiation toughness with loading rate; the average value was about 60 ksi- $\sqrt{\text{in.}}$.

In general, the arrest toughness levels were better defined than the initiation. On the whole, the variations in crack initiation toughness did not show an immediate correlation with thickness or strain rate, except in the case of one-inch thick specimens tested at room temperature where the specimen thickness was clearly inadequate to meaningful toughness data at low rates. At the lower temperatures one-inch and two-inch thick specimens appeared to give equally valid toughness measurements in the range of loading rates employed.

SPECIMEN SIZE REQUIREMENTS

The results presented in the preceding section were not identified as plane strain fracture toughness values. The slower tests on one-inch thick specimens gave results which were clearly not plane strain toughnesses, but this does not necessarily imply that the remaining results were valid. As mentioned above, to establish that a given thickness is sufficient to measure a plane strain toughness, it is necessary to demonstrate that further increases in thickness do not lead to the measurement of progressively lower values. Strictly speaking, such a demonstration

would require testing specimens of several different thicknesses at a given temperature and strain rate.

The plastic zone size at a given value of K is, at least to a first approximation, inversely proportional to the square of the yield strength (eq. 3a). Since it is the specimen thickness relative to this zone size which determines the degree of plane strain constraint, changes in yield strength produced by varying the test temperature or strain rate produce changes in the effective specimen thickness for measuring a given value of K_{Ic} . The concept of an effective thickness can be made formal by defining the effective thickness, B^* , as:

$$B^* = B \left(\frac{\sigma_Y}{\sigma_0} \right)^2 \quad (5)$$

where B is the actual specimen thickness, σ_0 is the static room temperature yield strength, and σ_Y is the yield strength at the strain rate and temperature of the test. The square of the yield strength ratio, then, can be considered as a correction factor for the thickness which accounts for changes in temperature and rate.

The thickness requirement for a plane strain toughness measurement can be estimated from the dependence of the measured toughness, K_{Ic} , on the effective thickness, B^* , as defined above. If the fracture toughness decreases significantly with decreasing strain rate and increasing loading rate, such a procedure would either over-estimate the minimum required thickness or prove inconclusive; i. e., a strong temperature and rate dependence of toughness could not lead to an incorrectly small value for the thickness, which might in turn result in erroneously high values for K_{Ic} . On the other hand, if the temperature or rate dependence of

toughness is small, the dependence of K_c on effective thickness will reflect primarily the thickness transition and therefore can be used to assess the minimum required thickness.

As has already been pointed out, the toughness data presented in the preceding section showed no strong rate dependence which was not likely attributable to a thickness transition. Moreover, the toughness levels at 0° and -70° were virtually the same. Therefore, an analysis of these data in terms of effective thickness offers promise as a method for assessing specimen thickness requirements.

The yield stress data plotted in Fig. 1 permitted the calculation of an effective thickness for each of the initiation toughness measurements presented in the preceding section. For calculating B^* in eq. (5), B was taken as the gross specimen thickness (i. e. one inch or two inches), σ_o was taken to be 71 ksi, and the value of σ_Y was taken from Fig. 1 by matching the reciprocal loading time of the yield stress measurement with that of the fracture toughness test. In Fig. 11, the measured values of initiation toughness (from Figs. 6, 8, and 9) are plotted against $(1/B^*)$, the reciprocal of the effective thickness calculated as indicated above. As can be seen in Fig. 11, although only two different specimen thicknesses (one-inch and two-inch) were tested, the values of $(1/B^*)$ are generously distributed over the range 0.3 to 1.0 inch^{-1} .

The room temperature initiation toughness data plotted in Fig. 11 are represented by a mean value of $85 \text{ ksi}\sqrt{\text{in.}}$ for effective thicknesses in the range $(1/B^*) = 0.4$ to $(1/B^*) = 0.8 \text{ inch}^{-1}$. These data include all of the tests on two-inch thick specimens and the tests on one-inch specimens at the higher loading rates. For $(1/B^*)$ greater than 0.8 inch^{-1} there is a marked progressive increase in K_c . These data points cor-

respond, of course, to the slower tests on one-inch thick specimens which were already pointed out in Fig. 6 as exhibiting an excessive departure from plane strain.

The tests at 0° and -70° F defined essentially the same level of initiation toughness, and the plot of K_c versus $(1/B^*)$ in Fig. 11 defines a single curve for the two temperatures. For $(1/B^*)$ ranging from 0.3 to 0.08 inch^{-1} all of the data are in reasonable agreement with a mean value, $K_c = 60 \text{ ksi-}\sqrt{\text{in.}}$. There are too few points at larger values of $(1/B^*)$ to establish the onset of a significant toughness increase.

The thickness requirement for the measurement of plane strain toughness, eq. (4), can be rewritten in terms of the parameter β_c defined as

$$\beta_c = \frac{1}{B} \left(\frac{K_c}{\sigma_Y} \right)^2 \quad (6)$$

The parameter β_c is, in effect, a ratio of the plastic zone size (at fracture initiation) to the specimen thickness. Thus, expressing the thickness requirement in terms of a maximum allowable value of β_c is identical to the criterion of eq. (4).

The current recommendation of ASTM Committee E-24 is $\beta_c \leq 0.4$; i. e. a measured value of toughness, K_c , can be identified as the plane strain fracture toughness, K_{Ic} , only if $\beta_c \leq 0.4$. The locus of points corresponding to $\beta_c = 0.4$ is plotted in Fig. 11. All of the room temperature data lie above $\beta_c = 0.4$ curve and are thus invalid according to the criterion of ASTM Committee E-24. To measure a valid K_{Ic} of the order of $85 \text{ ksi } \sqrt{\text{in.}}$ at room temperature would require $(1/B^*) < 0.27$ which corresponds to a specimen thickness of three inches for a dynamic

test and nearly four inches for a static test. For the lower temperature data only the two-inch thick specimens and the one-inch thick specimens tested at the highest loading rates are valid in terms of the criterion.

The absence of a dependence of K_{Ic} on effective thickness at values of β_c greater than 0.4, suggests that the criterion of ASTM Committee E-24 is too stringent for this material. The curve corresponding to $\beta_c = 1.0$ appears to be an adequate criterion for the data plotted in Fig. 11. This curve classifies virtually all of the lower temperature results as valid, in agreement with the observation that a single value of 60 ksi- $\sqrt{\text{in.}}$ is representative of the fracture toughness over the entire range of effective thickness in these tests. For the room temperature data the curve $\beta_c = 1.0$ provides a somewhat conservative criterion. The data indicate that a value of 85 ksi- $\sqrt{\text{in.}}$ could be measured up to $(1/B^*)$ of about 0.8 inch⁻¹ while the criterion $\beta_c = 1.0$ indicates a maximum allowable value of $(1/B^*) = 0.67$ inch⁻¹.

The analysis in terms of effective thickness indicates that an adequate criterion for this material is $\beta_c = 1.0$ which is equivalent to the thickness requirement: $B = 1.0 (K_{Ic} / \sigma_Y)^2$. This defines a minimum required thickness which is less than half that currently recommended by ASTM Committee E-24 which suggests the requirement, $B = 2.5 (K_{Ic} / \sigma_Y)^2$. Wessel (6) had presented data for A302-B steel quenched and tempered to a 50,000 psi yield strength level. A limited number of tests at -200°F, where K_{Ic} was about 45 ksi- $\sqrt{\text{in.}}$, also suggested that valid measurements could be made with B as small as 1.0 to 1.5 times $(K_{Ic} / \sigma_Y)^2$.

In its strictest sense, the plane strain fracture toughness K_{Ic} might be considered to be the limiting value of K_c as β_c tends toward zero. In line with this viewpoint, Irwin has suggested a relationship

between K_c and K_{Ic} based on experimental results from several different materials (7). The relationship can be written in the form:

$$\left(\frac{K_c}{K_{Ic}}\right)^2 = 1 + \frac{1.4}{B^2} \left(\frac{K_{Ic}}{\sigma_Y}\right)^4 \quad (7)$$

The formula associates K_{Ic} with an infinite plate thickness ($B \rightarrow \infty$). This definition of K_{Ic} yields a somewhat lower value of K_{Ic} than is allowable with the ASTM criterion. Indeed, if one assumes $\beta_c = 0.4$, and that this yields a toughness K_c , then the calculated K_{Ic} is lower than the measured value by about seven percent.

When the value of K_{Ic} is calculated according to eq. (7) on the assumption that all the data of Figs. 6, 8 and 9 are K_c values, the results plotted in Fig. 12 are obtained. The effect of the thickness correction is shown most strikingly in the room temperature test results, particularly for the one-inch thick specimens (see Fig. 6). The apparent rate dependence shown by the uncorrected data has been eliminated; the corrected data for one-inch specimens define a toughness of 70 ksi- $\sqrt{\text{in}}$. over the whole range of loading rates. The corrected results for two-inch thick specimens define a slightly higher toughness. Applying the thickness correction to the test results for 0°F and -70°F had only a modest effect because of the relatively lower values of $(K_c / \sigma_Y)^2$ at the lower temperatures. For both low temperatures the results plotted in Fig. 12 indicate a K_{Ic} value of 55-60 ksi- $\sqrt{\text{in}}$. in good agreement with the lower boundary of the uncorrected arrest toughness measurements.

The behavior of the thickness correction is also illustrated in Fig. 13. Assuming K_{Ic} values of 70 and 55 ksi- $\sqrt{\text{in}}$., respectively, for the ambient and low temperature tests (from Fig. 12), eq. (7) was used

to generate curves of K_c versus $1/B^*$ in Fig. 13. The $K_{Ic} = 70 \text{ ksi-}\sqrt{\text{in.}}$ curve fits the one-inch room temperature data very well, but underestimates the results from two-inch specimens. Thus, calculated value, $K_{Ic} = 70 \text{ ksi-}\sqrt{\text{in.}}$, which is based primarily on data for which $\beta_c > 1.0$, is conservative with respect to the measured data obtained on two-inch thick specimens (i. e. $\beta_c < 1.0$). The latter suggest $K_{Ic} = 85 \text{ ksi-}\sqrt{\text{in.}}$. The K_c vs. $1/B^*$ curve based on $K_{Ic} = 55 \text{ ksi-}\sqrt{\text{in.}}$, Fig. 13, represents the 0°F and -70°F results quite well. It is again seen that the lower boundary value of arrest toughnesses at both ambient and low temperatures defines a value consistent with the calculated K_{Ic} . These values are plotted in Fig. 13, at some arbitrary value to the left of the drop-weight tests, with the two-inch thick data points at a $1/B^*$ position equal to roughly half that of the one-inch thick point.

Applied to the data presented here, eq. (7) provides a method for converting a measured toughness, K_c , into a plane strain toughness, K_{Ic} , which is conservative in comparison to the values indicated by the uncorrected data. As a value for K_{Ic} which represents a lower bound for crack toughness in thick sections, a value $70 \text{ ksi-}\sqrt{\text{in.}}$ at room temperature and $55 \text{ ksi-}\sqrt{\text{in.}}$ at 0°F and -70°F is obtained. Although it is probable that the actual toughness lies between these values and the directly measured values; the difference between the corrected and uncorrected toughness numbers is not large, and the corrected ones appear to be conservative.

The most striking feature of these test results was the modest influence of loading rate on K_c , and the slight differences between initiation and arrest toughness values. Where a rate dependence did occur, it could be attributed to a thickness influence.

As indicated in an earlier section, the running crack should seek out the velocity corresponding to the minimum crack toughness, and hence crack arrest should require that the stress intensity factor fall below this minimum value. In spite of this, the arrest toughness values were virtually the same as for initiation. There was a tendency for successive initiation and arrest toughness values obtained on a single specimen to show a progressive increase. Such an increase must reflect a progressive change in the condition of the crack, for example, a roughening of the fracture surfaces. Just as the initiation toughness is found to be influenced by the fatigue pre-cracking procedure, so might the arrest value depend on the amount of "roughness" developed by the running crack. Indeed, if an enhanced toughness produced by "roughening" were to develop sufficiently rapidly, the crack would be expected to arrest at a toughness level only very slightly below the initiation value. To eliminate the contribution of crack "roughness", one might compare each arrest toughness value with the immediately following initiation toughness. The tests run in this study furnished forty-eight instances where such a comparison could be made. The difference between such an initiation and arrest toughness ranged from 9 to 36 percent of the initiation value, with an average difference of about twenty percent at each of the three testing temperatures. If a fast fracture initiated from a fatigue crack propagated without "roughening" one might guess that the arrest toughness would be of the order of twenty percent less than the initiation toughness. This is still a much smaller rate effect than Krafft's data would suggest; yet, unless "roughening" could always be counted on to enhance the toughness, this twenty percent toughness loss could be of critical practical importance.

Another factor which should be considered is variation of the condition of the crack across the specimen thickness. Normally, K_{Ic} is calculated from a "pop-in" load. The "pop-in" corresponds to unstable propagation of the center portion of the crack, which (when the thickness is sufficient) represents a condition of plane strain. Arrest of the "pop-in" crack, however, is almost certainly dominated by the departure from plane strain conditions near the specimen surfaces, rather than the value of K at the center of the plate. Hence, arrest of the "pop-in" crack does not represent a meaningful arrest toughness of the material. In the tests described here the initial instability resulted in propagation of the crack across the full specimen thickness, rather than just in the center portion. This behavior may be attributed to the effect of the side grooves in restricting plastic flow at the plate surface.

SUMMARY

High loading rates, because they elevate the yield strength, significantly reduce specimen size requirements. The magnitude of yield strength elevation is sizable; four orders being equivalent to more than a change in testing temperature of 140°F .

Side grooves appear to be essential for measuring arrest toughness. These values are only meaningful when the crack "pops" across the full specimen thickness. The drag introduced by shear lips probably prevent such uniform pops.

For values of $\beta_c = 1/B (K_c / \sigma_Y)^2$ less than about 1.0, the measured values of initiation toughness were independent of thickness and loading rate. Thus, the criterion $\beta_c \leq 1.0$, instead of the more con-

servative criterion, $\beta_c = 0.4$, better described the thickness requirement for plane strain fracturing in this material. Using this criterion, most of the room temperature data and all of the 0° , and -70°F data, represent valid toughness measurements. They define $K_{Ic} = 85-90 \text{ ksi} \cdot \sqrt{\text{in.}}$ at room temperature and $K_{Ic} = 55-60 \text{ ksi} \cdot \sqrt{\text{in.}}$ at both 0° and -70°F . The arrest toughness were $75-80 \text{ ksi} \cdot \sqrt{\text{in.}}$ at room temperature and $55-60 \text{ ksi} \cdot \sqrt{\text{in.}}$ at the lower temperature.

A more conservative method of defining K_{Ic} was to use the $K_{Ic} - K_c - \beta_c$ relationship suggested by Irwin. This gave $K_{Ic} = 70 \text{ ksi} \cdot \sqrt{\text{in.}}$ at room temperature and $K_{Ic} = 55 \text{ ksi} \cdot \sqrt{\text{in.}}$ at 0° and -70°F . These values are also consistent with the arrest toughness measurements. While it is probable that these values underestimate the actual toughness, the difference between these numbers and the measured ones is not too large, and the corrected values represent a reasonable lower bound for K_{Ic} .

The data indicated little strain rate sensitivity of K_{Ic} . The effect of loading rate on toughness was modest, and the difference between initiation and arrest values was minimal. While artificial enhancement of arrest toughness by crack "roughening" cannot be ruled out as a possibility, there was no positive evidence to support this interpretation. Despite variations in the length of unstable extension preceding arrest, arrest toughness values generally showed less scatter than the initiation.

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Fig. 1 Yield Strength of A533 Steel as a Function of Loading Rate.

Fig. 2 Effect of Loading Rate on the Specimen Thickness (B) required to Measure a Given Value of K_{Ic} . $B = B_0$ for Static Test at Room Temperature. (Based on data of Fig. 1).

Fig. 3 Effect of Strain Rate on K_{Ic} (-12°F , 0.20% C Steel, $\sigma_Y = 34$ ksi at 75°F).

Fig. 4 Effect of Side Grooves on the Measured Fracture Toughness of 7075-T651 Aluminum Alloy.

Fig. 5 Relative Sizes of $m = 3$ and $m = 8 \text{ in.}^{-1}$ Fracture Toughness Specimens. (Specimens (a) and (b) have same profile except for loading hole configuration.)

Fig. 6 Room Temperature Fracture Toughness of A533 Steel.

Fig. 7 Fracture Toughness of A533 Steel at 0°F. Specimens Fatigue Pre-cracked at Room Temperature at 5 to 10 μ -in./cycle.

Fig. 8 Fracture Toughness of A533 Steel at 0°F. Specimens Fatigue Pre-cracked at -20 to -60°F at 1 to 2 μ -in./cycle.

Fig. 9 Fracture Toughness of A533 Steel at -70°F . Specimens Pre-cracked at -20 to -60°F at 1 to 2 $\mu\text{-in./cycle}$.

Fig. 10 Fracture Toughness of A533 Steel at -70°F Measured With Different Specimens.

Fig. 11 Dependence of Fracture Toughness on Effective Thickness (Data from Figs. 6, 8, 9).

Fig. 12 Thickness Corrected Fracture Toughness of A533 Steel. (Initiation Data from Figs. 6, 8, 9).

Fig. 13 Correlation of Thickness Correction With Initiation and Arrest Toughness Values.