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A DOUBLE CANTILEVER BEAM SPECIMEN
FOR DETERMINING THE PLANE STRAIN FRACTURE
TOUGHNESS OF METALS

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A DOUBLE CANTILEVER BEAM SPECIMEN FOR DETERMINING THE PLANE STRAIN FRACTURE TOUGHNESS OF METALS

INTRODUCTION

The purpose of this paper is to describe the results of a study determining the feasibility of using a double cantilever beam (DCB) specimen for plane strain fracture toughness measurements. These measurements in low-strength, high-toughness materials require large notch-bend, round notch-tensile, and single- and double-notch plate specimens to provide plane strain restraints. Often the required size is too large for investigating environment effects in restricted spaces or for obtaining specimens from limited material sources. For example, there is a great need for information on the fracture behavior of materials exposed to neutron radiation in hot water, high-temperature gas and liquid metal environments. Since existing irradiation facilities have a very limited capacity, fracture studies using large, conventional specimens are either difficult or impractical. A typical DCB specimen (Figure 1) is sufficiently small to satisfy many of these testing needs involving size and space limitations.

Surface energy studies of materials have been performed with the DCB specimen by a number of investigators.⁽¹⁻⁵⁾ Extensive analysis of the specimen mechanics has been carried out by Berry⁽⁶⁾ and Gilman.⁽⁷⁾ As noted by Berry, unstable crack propagation for any crack length is impossible in the DCB specimen if the specimen is composed of a material possessing a strain energy release rate or crack resistance force independent of crack velocity. Since most structural materials exhibit some degree of rate sensitive flow, crack instabilities can occur in the DCB specimen due to a decrease in the crack resistance force with increasing crack velocity. If the increase in crack length (resulting from unstable propagation) is small relative to the length of the specimen, a number of fracture data may be obtained from a single specimen of moderate size. A schematic DCB test record showing load versus extension is given in Figure 2.

FIGURE 1

DCB Specimen and Loading Arrangement

FIGURE 2

Typical DCB Specimen Fracture Behavior
at Constant Crosshead Speed

A comparison of nominal stresses at the crack tip, for other than very short cracks, shows that the stress in the length direction (bending stress) exceeds the stress in the height direction for cantilevered arms. Since opening mode crack propagation occurs perpendicular to the principal stress, the bending stress in the arms of the DCB specimen will cause the crack to veer from the midplane of the specimen and will cause one of the arms to break off. Berry⁽⁴⁾ has solved this problem by machining grooves at middepth down both sides of the specimen. These grooves maintain the crack in the desired midplane position and provide important restraints on the plastic zone at the crack tip.

SUMMARY AND CONCLUSIONS

The important findings of this study are as follows:

- The appropriate relationship defining the mechanics of the DCB specimen (except for very long or very short crack lengths relative to the length of the specimen) is

$$y = \frac{fAc^n}{E} ,$$

where $A = \frac{C_1}{I}$ and the constants C_1 and n depend only on beam depth.

- K_{Ic} tends to increase as w/b (the ratio of net to gross thickness) becomes small.
- As the beam depth becomes large relative to the thickness, K_{Ic} varies with crack length because of elastic buckling of the specimen arms. Increasing the thickness diminishes this effect.
- The side groove root radius must be small enough to prevent the formation of shear lips. However, the requirements for root sharpness are much less stringent than for notched tensile or bend specimens.
- The difference between the fracture toughness at crack initiation, K_{Ic} , and crack arrest, K_{Ia} , provides a measure of a material's ability to arrest a moving crack.

MECHANICS OF THE DCB SPECIMEN

From dimensional considerations, the relationship between the applied force f , the crack length c , and the resultant total deflection of both arms y , can be derived as

$$y = \frac{f}{E} \varphi, \text{ where } \varphi = \varphi(I, c, h) \quad (1)$$

where E = modulus of elasticity

I = moment of inertia

h = beam depth.

This relationship has been evaluated by compliance measurements on 20 DCB specimens of 7075-T6 aluminum, each having different combinations of gross thickness, beam depth, and side groove depth (Table I). The side groove root radius was varied in the last four specimens in order to observe this effect on the fracture toughness (discussed in the next section). The length of the specimens was parallel to the rolling direction of the plate. The relationship between φ and c was determined for each of the specimens from the slope of the load extension line from Instron recordings and from the crack length measured with a traveling microscope to an accuracy of ± 0.001 in. Because the crack length from one side of the specimen differed by as much as 0.030 in. (but in most cases no more than 0.005 in.), an average value was used. The crack length was increased simply by increasing the load until propagation occurred. The increase in crack length as a result of unstable propagation was generally small so that the amount of data acquired from each specimen provided a statistical basis for determining the relationship between φ and c .

A typical plot between φ and c (Figure 3) suggests an equation of the form

$$\varphi = A c^n, \quad (2)$$

where A and n are constants. Equation (2) agrees with Berry's⁽⁴⁾ results in his surface energy determinations for plastics. Equation (2) fits the data well for all of the specimens except for very short and very long cracks relative to the specimen length. For cracks both shorter than

TABLE I
TABULATION OF SPECIMEN DIMENSIONS

Specimen	Beam Depth, in.	Gross Width, in.	Side Groove Depth, in.	Side Groove, Root Radius, in.	Moment of Inertia, 10^{-3} in. ⁴
1	0.500	0.125	0.030	≤ 0.001	1.275
2	0.500	0.250	0.094	≤ 0.001	2.329
3	0.500	0.251	0.028	≤ 0.001	2.591
4	0.500	0.376	0.155	≤ 0.001	3.194
5	0.500	0.375	0.122	≤ 0.001	3.452
6	0.500	0.376	0.090	≤ 0.001	3.663
7	0.500	0.376	0.059	≤ 0.001	3.812
8	0.500	0.501	0.217	≤ 0.001	3.885
9	0.500	0.501	0.185	≤ 0.001	4.410
10	0.500	0.501	0.152	≤ 0.001	4.530
11	0.500	0.501	0.119	≤ 0.001	4.786
12	1.000	0.125	0.030	≤ 0.001	10.320
13	1.000	0.253	0.060	≤ 0.001	20.590
14	1.000	0.376	0.090	≤ 0.001	30.283
15	1.000	0.502	0.120	≤ 0.001	39.921
16	0.500	0.250	0.060	≤ 0.001	2.497
17	0.500	0.251	0.060	0.010	2.497
18	0.500	0.251	0.060	0.050	2.497
19	0.500	0.501	0.120	0.010	4.782
20	0.500	0.501	0.120	0.050	4.773

FIGURE 3
Compliance Test Results
on Specimen No. 16

1-3/4 in. and within about 5/8 in. of the specimen end, significant deviations from Equation (2) occur.

Values for A and n are given in Table II, along with their 95% confidence limits. Figure 4 shows the variation in A with moment of inertia.

TABLE II
RESULTS OF FIT TO EQUATION $\varphi = Ac^n$

<u>Specimen</u>	<u>A</u>	<u>n</u>
1	1004 ± 31	2.742 ± 0.0210
2	554 ± 23	2.738 ± 0.0250
3	505 ± 21	2.710 ± 0.0270
4	433 ± 21	2.685 ± 0.0330
5	358 ± 10	2.749 ± 0.0190
6	355 ± 14	2.701 ± 0.0250
7	332 ± 12	2.748 ± 0.0220
8	339 ± 21	2.753 ± 0.0380
9	304 ± 12	2.733 ± 0.0250
10	280 ± 13	2.724 ± 0.0290
11	256 ± 14	2.741 ± 0.0370
12	239 ± 20	2.467 ± 0.0560
13	121 ± 12	2.455 ± 0.0660
14	90 ± 8	2.383 ± 0.0630
15	66 ± 10	2.422 ± 0.1022
16	515 ± 26	2.710 ± 0.0350
17	506 ± 19	2.747 ± 0.0250
18	525 ± 28	2.720 ± 0.0340
19	258 ± 9	2.739 ± 0.0210
20	263 ± 10	2.755 ± 0.0230

For a given beam depth, the relationship is found to be

$$A = \frac{C_1}{I}, \quad (3)$$

where C_1 is a constant. Equation (1) can now be rewritten as

$$y = \frac{fC_1c^n}{EI} \quad (4)$$

which is similar to the simple cantilever beam formula,

$$y = \frac{2fc^3}{3EI}. \quad (5)$$

FIGURE 4

Variation of "A" with Moment of Inertia

By doubling the beam depth (from 0.5 in. to 1.0 in.) the value of C_1 is also increased by nearly a factor of 2. However, for both beam depths, the value of C_1 is larger than would be predicted by the simple beam formula (0.667). This variation in C_1 with beam depth is a result of strain in the unbroken part of the specimen. Unfortunately, no rigorous solution for the stress state in the specimen for this condition exists.

As Table II shows, the crack length exponential n also varies with beam depth, is constant for a given beam depth, and departs from the simple beam value of 3 for increasing beam depth. The values of C_1 and n would be expected to approach 0.667 and 3, respectively, as the beam depth approaches zero.

APPLICATION OF THE DCB SPECIMEN TO FRACTURE TESTING

The strain energy release rate, G , for a specimen can be determined experimentally in the absence of a mathematical analysis.⁽⁸⁾ The appropriate form of Equation (1) is

$$G = \frac{f^2}{2wE} \frac{d}{dc} (\varphi). \quad (6)$$

Incorporating Equation (4) into (6) yields

$$G = \frac{ny^2E}{2wA} \left(\frac{Af}{Ey} \right)^{\frac{n+1}{n}}. \quad (7)$$

Given the corresponding constants, A and n , Equation (7) can be used to determine G independent of crack length. The strain energy release rate, G , may then be converted to the stress intensity factor, K , using the relationship $K = \sqrt{EG}$ for plane stress or $K = \frac{\sqrt{EG}}{1 - \nu^2}$ for plane strain, where ν is Poisson's ratio.

When fracture occurs, the results can be described in terms of G_c or K_{Ic} (or their plane strain counterparts, G_{Ic} or K_{Ic}). In this paper (for the sake of consistency) the data are given in terms of K_{Ic} . The justification for using plane strain toughness is discussed later.

Fracture toughness data for five materials (including the aluminum specimens used for the compliance measurements) are presented to demonstrate both the advantages and limitations of the DCB specimen. Table III gives the chemical compositions and the room temperature properties of four of these materials. Unfortunately, these data were not available for the 7075-T6 aluminum. All the specimens tested (other than the aluminum calibration specimens) had essentially the same dimensions as Specimen 16 in Table I. The time rate of change of the stress intensity factor, $\dot{K}$, at crack initiation for the tests was about 300 psi/in./sec.

The 20, 7075-T6 aluminum specimens prepared to determine variations in A and n with specimen geometry also aided in observing geometry effects on fracture toughness. Stress intensity for crack initiation is plotted as a function of crack length in Figure 5 for the various

geometries given in Table I. The curves represent the best fit through the approximately 700 fracture toughness values obtained. In general, the K_{Ic} values fall in the range of 26 to 29 ksi/in., which are fairly representative of K_{Ic} values for 7075-T6 aluminum obtained by other methods.^(9,10) There were, however, trends deserving further comment.

TABLE III
MATERIALS

<u>Material</u>	<u>Composition, wt%</u>	<u>Room Temperature Mechanical Properties</u>
A-302B SS	0.19 C, 1.26 Mn, 0.55 Mo, 0.20 Ni, 0.14 Cr, 0.31 Si, 0.020 P, 0.018 S, balance Fe	66 ksi 0.2% YS, 86 ksi UTS, 26.4% elongation (in 1 in.), 71.4% RA $C_V 15 = 35^\circ F^{(c)}$
Zircaloy-2	1.44 Sn, 0.118 Fe, 0.092 Cr, 0.046 Ni, 133 ppm C, 9 ppm H, 1146 ppm O, balance Zr	55.6 ksi 0.2% YS, 82 ksi UTS 21.5% elongation (in 2 in.)
Mild Steel ^(a)	0.20 C, 0.76 Mn, 0.16 Ni, 0.019 P, 0.040 S, 0.03 Si, balance Fe	34.3 ksi LYS, 61.6 ksi UTS, 39.5% elongation (in 2 in.) 65.3% RA $C_V 15 = 20^\circ F$
Ni-Mo-V ^(b)	0.30 C, 2.6 Ni, 0.44 Mo, 0.09 V, 0.48 Mn, 0.22 Si, 0.03 P, 0.03 S, balance Fe	105.3 ksi 0.2% YS, 129.9 ksi UTS, 10% elongation (in 1 in.), 25.1% RA $C_V 15 = 210^\circ F$

(a) This material formed a portion of Specimen 39 in the University of Illinois rapid crack propagation studies, and was kindly supplied by W. F. Hall.

(b) This material was kindly supplied to the author by A. J. Brothers of the Materials and Processes Laboratory, General Electric Company, Schenectady, New York. It was obtained from a 6 in. square notch bend specimen used in extensive fracture studies on this steel.

(c) Charpy V-notch transition temperature at the 15 ft. lb energy level.

Examination of two groups of specimens; namely, the 3/8 in. wide specimens (4, 5, 6, and 7) and the 1/2 in. wide specimens (8, 9, 10, and 11), reveals a tendency for K_{Ic} to increase as w/b becomes small. For the 3/8 in. wide specimens, K_{Ic} becomes essentially constant for w/b greater than 0.351; while for the 1/2 in. wide specimens, K_{Ic} attains a constant value for w/b greater than 0.134. For specimens having a 1/2 in. beam depth, sharp (0.001 in.) root radius, and w/b greater than 0.250, K_{Ic} fell in the range of 27 to 29 ksi/in.

FIGURE 5
Fracture Toughness Results as a Function of Crack Length
for the Twenty 7075-T6 Aluminum Specimens.

It can be seen in Figure 5 that specimens with the 1 in. beam depth, particularly Specimens 12, 13, and 14, exhibit a pronounced decrease in K_{Ic} as the crack length increases. For these three specimens, K_{Ic} varies approximately as $c^{-1/2}$. A dependence this strong cannot be attributed to an error in the compliance measurements; however, Specimen 12 exhibited a noticeable amount of twist during testing due to elastic buckling in the arms. This instability subjects the crack tip to a shearing action described as Mode III by Paris and Sih.⁽¹¹⁾ It is thought that this buckling (which becomes more severe with increasing crack length) encourages the apparent K_{Ic} decrease as the crack length increases, since an increase in the Mode III stress field would decrease the required opening mode stress field for crack propagation. Buckling is remedied by increasing the specimen thickness (Specimen 15).

The effect of side-groove root radius on fracture toughness is revealed by Specimens 18 and 20. After Specimen 18 (1/4 in. width, 0.050 in. side groove root radius) separated into two halves, a sharp "flip-flop" shear lip was observed at the base of each groove. Since no shear lip occurred for specimens with a 0.001 in. root radius, this shearing along the sides of Specimen 18 is believed to cause the inordinately high value of K_{Ic} . A similar shear lip also occurred along the sides of Specimen 20 (1/2 in. width, 0.050 in. root radius). However, the relative contribution of this shear lip on K_{Ic} was smaller than for Specimen 18 because of the greater crack width of Specimen 20. The effect of a 0.010 in. root radius is mildly discernible for a 1/4 in. width (Specimen 17), but is not apparent for a 1/2 in. width (Specimen 19). Therefore, the root radius must be selected to give a shear lip which is small relative to the net width, w . The required side groove root radius for 7075-T6 aluminum, where w is 0.130 in. or larger, can be much greater than the required notch root radius in conventional notch tensile or bend specimens.

By conducting successive crack initiations over a range of temperatures, the variation in K_{Ic} with temperature from very few specimens can be determined. Figure 6 shows the results of such a test on a single 7075-T6 aluminum specimen. This specimen was machined so that the crack propagated transverse to the rolling direction.

FIGURE 6

Effect of Temperature on the Fracture Toughness
at Crack Initiation and Arrest for 7075-T6 Aluminum.

It is interesting to note that the fracture toughness reached a minimum at -60°C rather than at the lowest temperature. In fact, the fracture toughness at -140°C is slightly higher than at room temperature. Furthermore, the K_{IC} at room temperature is slightly higher for the transverse direction than for the rolling direction. The stress intensity corresponding to crack arrest is also plotted against temperature (Figure 6). The physical significance of these arrest values as well as the ones in subsequent figures will be discussed in the next section.

Obtaining fracture toughness data on a material with a much higher toughness and lower yield strength than 7075-T6 aluminum is more difficult. Such is the case for A-302B in the normalized condition. The results of tests on two DCB specimens of this material are shown in Figure 7. A larger specimen probably would have enabled toughness data to be obtained at higher temperatures. In addition, a strongly rate-sensitive fracture

toughness results in rather long crack jumps, particularly as the temperature is increased. As a result, only four or five data points were obtained from each 7 in. specimen.

FIGURE 7

Effect of Temperature on the Fracture Toughness
at Crack Initiation and Arrest for Normalized A-302 B

The effect of temperature on the fracture toughness for crack propagation, both parallel and perpendicular to the rolling direction of annealed Zircaloy-2 plate, is shown in Figures 8 and 9. A comparison of these data with those obtained by Hoagland and Bement⁽¹²⁾ from 1/2 in. wide notch bend specimens indicates that the degree of side groove restraint is sufficient to ensure plane strain conditions. In fact, more restraint is available in the

FIGURE 8

Effect of Temperature on The Fracture Toughness at Crack Initiation and Arrest for Crack Propagation Perpendicular to the Rolling Direction in Annealed Zircaloy-2 Plate

FIGURE 9

Effect of Temperature on The Fracture Toughness
at Crack Initiation and Arrest for Crack Propagation
Parallel to The Rolling Direction in Annealed Zircaloy-2 Plate

DCB specimen than in the larger notch bend specimens. This is shown by the lack of gross deformation up to a temperature of -70°C in the rolling direction specimens and -50°C in the transverse direction specimens. Above these temperatures the load-extension behavior is nonlinear prior to crack initiation. Such nonlinearities can be the result of two causes: slow crack growth, or plastic bending of the arms. A correction to the data can be applied for plastic bending. The load-extension line is obtained by partially unloading the specimen (Figure 10). By extrapolating to zero load, the true extension values (or alternately) by extrapolating to zero load, the true extension value for calculating fracture toughness at crack initiation and arrest.

FIGURE 10

Method of Correcting Extension When Plastic Bending Occurs

Figures 11 and 12 show test results for a mild steel and a Ni-Mo-V rotor steel, respectively. Fracture properties on identical materials have been investigated comprehensively by Eftis and Krafft,⁽¹³⁾ who used SEN (single edge-notch) specimens, and by Mowbray,⁽¹⁴⁾ who used both SEN specimens and NB (notch bend) specimens. These comparisons illustrate the agreement of DCB results with results from more conventional methods.

As with the A-302B steel, it was not possible to cover a large range of temperatures with the mild steel for the specimen geometry used. However, over the 60 °C temperature range (Figure 10) there is excellent agreement between the DCB values and the isothermal fracture toughness $K_{Ic}(T)$ values of Eftis and Krafft interpolated at a value of $\dot{K}$ of 300 psi/in./sec. This agreement indicates that the amount of restraint developed in the DCB specimen is at least as great as that developed at pop-in in the SEN specimens. It has been argued that plane strain fracture conditions exist at pop-in in the SEN specimen.⁽¹⁵⁾

FIGURE 11

Effect of Temperature on The Fracture Toughness of a Mild Steel

FIGURE 12

Effect of Temperature on The Fracture Toughness of a Rotor Steel

DCB tests agree favorably with both the SEN and small NB tests of Mowbray in Figure 11. Tests on large NB specimens (3.0 in. beam width) by Mowbray suggest that smaller specimens may not truly yield plane strain fracture toughness of this material. However, the total absence of shear lips and the flat, crystalline fracture exhibited by the DCB specimen implies a plane strain condition, since plane stress conditions give rise to the formation and growth of shear lips commonly seen along the sides of notched plate or notch bend specimens.⁽¹⁶⁾

CONSIDERATION OF THE MEANING OF THE K_{Ic} AT ARREST

Numerous experiments have been performed to demonstrate that flow stress increases with increase in strain rate. Strain rate effects are manifested in the fracture process as a dependence in K_{Ic} on the loading rate. For example, a decrease in K_{Ic} as $\dot{K}$ increases (which is directly proportional to the nominal stress rate) has been established for a number of steels (including the mild steel considered in this paper).^(13, 16, 17) Since the energy release rate is almost equal to the energy absorption rate at the point of instability, any increase in crack velocity (from zero to a finite value) will cause an unstable condition because of a decrease in the instantaneous energy absorption rate. The excess energy (difference between the energy released and absorbed) will act as kinetic energy, and the crack will propagate until the kinetic energy is consumed.

Dynamic solutions for a moving crack in a DCB specimen are provided by Gilman⁽⁷⁾ and Berry.⁽⁶⁾ An instantaneous energy absorption rate, γ , dependent on crack velocity, establishes crack instability in these solutions if the energy absorption rate decreases as the crack velocity increases. This modification introduced into Berry's solution can be written.

$$\left(\frac{V_c}{V_s}\right) = \frac{35c}{3whEy^2} \left[\frac{3EIy^2}{2} \left(\frac{1}{c_o^3} - \frac{1}{c^3} \right) - \int_{c_o}^c \left(G_{Ic} + \frac{d\gamma}{dV_c} V_c \right) dc \right]^{1/2} \quad (8)$$

where (for simplification) the simple beam formula has been used in place of the exact Equation (4). In this equation, V_c is the crack velocity, V_s is the longitudinal wave velocity, c is the instantaneous crack length, c_o is the initial crack length, h is the beam depth, and γ is the instantaneous energy absorption rate incorporating all irreversible energy absorbing processes occurring as the crack propagates. The other symbols are previously defined. In addition to the error caused by the use of the simple beam formula, additional error will result unless V_c is very much less than V_s , a condition which is required to keep relativistic and vibrational effects small.

Since the purpose of this argument is to demonstrate (qualitatively at least) the meaning of G at arrest (G_{Ia}), a simple hypothetical case is appropriate in which the relationship between γ and V_c is defined as

$$\frac{d\gamma}{dV_c} = \frac{-k}{V_c} \quad (9)$$

The constant, k , in Equation (9) defines the degree of sensitivity of γ on V_c . Thus, larger k values means γ will decrease more rapidly with increasing V_c . By inserting Equation (9) into Equation (8), a relationship between V_c/V_s and c can be obtained (Figure 13). As expected, the greater the sensitivity of γ to V_c (larger k) the further the crack will progress. Figure 13 can be used to compute G_{Ia} and to observe the dependence of G_{Ia} on k . This dependency is shown schematically in Figure 14, together with the k dependence of the minimum γ attained at maximum crack velocity. In this hypothetical case, G_{Ia} falls about half way between G_{Ic} and γ min. Although the position of G_{Ia} between G_{Ic} and γ would probably be different for an actual material, this same general behavior will be obeyed as long as $\frac{d\gamma}{dV}$ is negative. The difference between G_{Ic} and G_{Ia} can now be considered a relative measure of the material's ability to arrest a running crack.

The arrest data reported earlier in this paper confirm that only a small and relatively temperature-insensitive difference exists between the initiation and arrest values for 7075-T6 aluminum and Zircaloy-2. Although adequate information relating the fracture toughness to a rate parameter for Zircaloy-2 is not presently available, speculation is that such behavior is influenced by the temperature-independent, rate-sensitivity index of zirconium deformation behavior between -196°C and room temperature.⁽¹⁸⁾ A similar speculation can be made for aluminum. This behavior is in contrast to the three steels where a much larger, strongly temperature dependent difference exists between G_{Ic} and G_{Ia} .

The mild steel data also provides an opportunity to compare the DCB arrest values with the minimum adiabatic K_{Ic} values (Figure 11) obtained by Eftis and Krafft.⁽¹³⁾ In Figure 11, K_{Ia} falls between the values for initiation and $K_{Ic(Q)min}$. The scatter of the arrest data prevents any firm conclusion about the exact position of the arrest curve, but it does demonstrate the expected behavior discussed previously.

FIGURE 13

Schematic Showing the Effect of a Hypothetical Rate Sensitivity Index, k , on the Crack Velocity Crack Length Profile in a DCB Specimen

FIGURE 14

Schematic Depicting the Effect of a Hypothetical Rate Sensitivity Index on G_{Ia} Compared to the Minimum Energy Absorption Rate, γ , During the Fracture Event

DISCUSSION

The DCB technique has been applied to facilitate the acquisition of fracture toughness data under conditions of limited space availability. Although this study has shown the need for specimen geometry adjustment to reduce unwanted effects, the feasibility of the testing method has been demonstrated. Two favorable features are the amount of fracture toughness data that can be obtained from one specimen and the restraint against excessive deformation near the crack tip provided by the sharp grooves.

The effectiveness of the side grooves in providing plane strain conditions at crack initiation is indicated by the good agreement in plane strain toughness data between DCB specimens and larger, conventional specimens for at least four of the five materials discussed. This restraint arises from an elastic intensification of the stress parallel to the crack front immediately near the side groove tip. This restraint imposes a highly tri-axial stress state in the plastic zone across the net width of the specimen. A similar effect occurs in notch rounds. However, in plate or bar specimens with no side grooves, the specimen must be thick enough so that the plane stress regions near the surfaces are small relative to the interior region which tends towards plane strain.⁽¹⁹⁾ Excessive plastic deformation may still wash out the stress intensification effect of the side grooves in a DCB specimen if the yield strength of the material is low and the specimen size is too small. Experimentation is necessary to achieve the correct specimen design for a given material.

The present technique has two drawbacks which deserve attention and further study. These liabilities are the K_{Ic} dependence on crack length demonstrated for certain of the 7075-T6 calibration specimens and the difficulty encountered in testing tough materials (such as the normalized A-302B and mild steel). The first problem may be avoided by proper specimen design. The beam thickness should be large enough for a given beam depth to provide lateral stability against buckling. The second problem may be reduced in two ways: by increasing the section modulus ($\frac{2I}{h}$) or by

decreasing the net width. The first method seems preferable since, as was found from specimens 4 through 11, K_{Ic} tends to increase as w/b becomes very small.

At the beginning of this investigation, K_{Ic} at arrest had doubtful significance. It is now felt that this quantity provides a useful measure of the relative ability of various materials to arrest a moving crack. These comparisons must be based, however, on specimens of the same geometry and on loading systems of the same stiffness, (as hard as possible). For example, although the fracture toughness for rotor steel is slightly higher than for Zircaloy-2 at -60°C , the ability of Zircaloy-2 to arrest a running crack is greater because of the relatively small difference between K_{Ic} and K_{Ia} for Zircaloy-2 compared to the rotor steel. The ability of a material to arrest a moving crack is particularly important in pressure vessels and piping. Therefore, these applications warrant selecting materials on the basis of fracture arrest ability and resistance to fracture initiation as well.

It is interesting to note that if the instantaneous energy absorption rate does not decrease with increasing crack velocity but remains constant or perhaps even increases, then it should be possible to closely control the crack velocity in a DCB specimen. Since the crack can be accurately located if the load and extension are known, the velocity of the crack can also be accurately measured. This capability suggests an interesting method for studying slow crack growth, stress corrosion cracking, delayed failure, and a number of other areas where fracture is time-dependent. Furthermore, the specimen and testing method provides an excellent means for determining the fracture toughness profile through a weld, associated heat affected zones, and base material. There are, undoubtedly, many other examples where the ability to measure fracture toughness gradients directly is of considerable value.

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