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FRACTURE-TOUGHNESS TESTS ON A HIGH-STRENGTH BERYLLIUM FROM ROOM TEMPERATURE TO 300°C

Mark Perra

Iain Finnie

March 10, 1975

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LAWRENCE LIVERMORE LABORATORY
University of California Livermore, California 94550

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Foreword

Mark Perra performed the experiments described in this report while employed by LLL in the summer of 1973. He is presently a graduate student in the Department of Mechanical Engineering at the University of California at Berkeley. Iain Finnie is a consultant to LLL. He is a Professor of Mechanical Engineering at the University of California at Berkeley.

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FRACTURE-TOUGHNESS TESTS ON A HIGH-STRENGTH BERYLLIUM FROM ROOM TEMPERATURE TO 300°C

Abstract

As one step in the characterization of its mechanical, metallurgical, and physical properties, the temperature dependence of the fracture toughness of a high-yield-strength grade of hot-pressed beryllium block was determined over the range between room temperature and 300°C. Fatigue cracks were generated in 8.92-mm-thick double-cantilever beam specimens by a reversed-loading method. Room-temperature fracture-toughness data from tests on specimens with Elox-machined^{*} notches and fatigue precracks were compared. Electron fractographs of fatigue and fast-fracture regions

were obtained. Load-displacement records generated from tests on specimens with natural cracks were non-linear, whereas tests on specimens with machined cracks gave the expected linear records. Evidence points to a crack-closure phenomenon rather than plasticity as the source of the non-linearity. The fracture-toughness data indicate that for this material the mean value of the toughness increases from 9.0 MPa·m^{1/2} (8.1 ksi·in^{1/2}) to 10.8 MPa·m^{1/2} (9.7 ksi·in^{1/2}) between 23°C and 300°C. These values are among the lowest ever reported for beryllium.

Introduction

Beryllium is an attractive material for applications in which moderately high yield strength, high elastic modulus, low density, and good fatigue resistance are desirable. The principal feature of beryllium which interferes with full realization of its potential as a structural material in specialized applications is its tendency to brittle fracture at low stress intensities. At room temperatures, tensile elongations of less than 3% are commonly

reported for hot-pressed beryllium; no shear lips are evident on fracture surfaces and cleavage mechanisms dominate. The propensity of beryllium to brittle fracture presumably arises from its ease of cleavage on several planes (most conspicuously the basal plane) and the resistance of its lattice to the generation and motion of dislocations having a component resolvable along the c-axis.

The influence of elevated temperatures on the resistance of beryllium to fracture is largely undetermined, although this is a problem of considerable practical significance. Such information would be of value, for instance, in a program for

^{*}Reference to a company or product name does not imply approval or recommendation of the product by the University of California or the Energy Research & Development Administration to the exclusion of others that may be suitable.

controlling the fracture of welded parts. Welded beryllium parts frequently fail as a result of the propagation of cracks in tensile fields which are present in the vicinity of the welds. In at least some cases, preheating can diminish cracking in welded beryllium parts.

The amount of data available in the literature on the elevated-temperature fracture toughness of beryllium is small. The results of Harrod et al.¹ for S-200 grade beryllium indicate that between 23°C and 260°C the fracture toughness increases from about 20 MPa·m^{1/2} (18 ksi·in^{1/2}) to 38 MPa·m^{1/2} (34 ksi·in^{1/2}). The value for 250°C was based on the results of a single test. More recently, the data of Shabbits and Logsdon² on S-200 grade beryllium indicated no appreciable increase in toughness between 23°C and 150°C but showed a rapid increase in

toughness between 150°C and 260°C going from about 10 MPa·m^{1/2} (9 ksi·in^{1/2}) to about 16 MPa·m^{1/2} (14 ksi·in^{1/2}) respectively. These values were obtained from fatigue-precracked compact-tension specimens oriented with the crack plane perpendicular to the pressing direction, and were based on the results of three tests at 150°C and two tests at 260°C. In this temperature range, no data were generated from specimens oriented with the crack plane parallel to the pressing direction.

The present study has been undertaken with the objective of determining the temperature dependence of the fracture toughness of a high-yield-strength grade of beryllium between room temperature and 300°C, and is part of a broader program of characterizing the mechanical, metallurgical, and physical properties of this material.

Materials and Methods

SPECIMENS

The beryllium used in this study was a vacuum-hot-pressed block of high-yield material supplied by Brush-Wellman. The chemistry was within the limits of S-200 grade, but a deliberate attempt was made to produce a high yield-strength. Selected chemical, physical, and mechanical properties of this material are listed in Table 1. The physical and tensile properties are those obtained by Gelles,³ The chemical analyses listed in Table 1 are those established by Brush-Wellman and by the Battelle Columbus Laboratory. In comparison with most commercial grades of beryllium, this material has a rather low density, a small grain size,

and a high iron content. Based on iron content, as determined by wet chemical methods at Battelle, and on aluminum content, as determined by neutron-activation analysis, the iron/aluminum ratio is 3.2.

The specimens were cut from the high-yield block in two crack-plane orientations with respect to the pressing direction. There were 15 transverse specimens (i. e., crack-plane parallel to the pressing direction). These were numbered GL1 to GL15. In addition, there were 10 longitudinal specimens (i. e., crack-plane perpendicular to the pressing direction). These specimens were numbered DE1T to DC010T.

The double-cantilever beam (DCB) specimens, which were 8.91 mm thick,

Table 1. Selected chemical, mechanical, and physical properties of high-yield-grade beryllium.

Chemical Composition			
	Brush-Wellman	Battelle Columbus Laboratory	
	(ppm)	Chemical Determination (ppm)	Determination by neutron-activation analysis (ppm)
BeO	20,040	—	14,100
Fe	1,680	1,500	—
C	1,230	1,430	—
Al	620	400	469
Mg	80	60	—
Si	47 ^a	200	—

Tensile Properties (8.5×10^{-3} mm/sec at room temperature)	
Property	Value
0.2% offset yield strength (MPa)	338
Ultimate strength (MPa)	386
Total elongation (%)	2.1
Reduction of area (%)	2.6
Density (% of theoretical)	99.45 to 99.71
Grain size (μm)	8.8 ± 0.03

were pin loaded in tension. Figure 1 is a diagram of such a specimen. Changes of the dimension δ were recorded during tests. An Elox slot, shown enlarged below the figure, was 0.102 mm wide and 1.27 mm long. The distance between the line connecting the centers of the loading pin holes and the tip of the Elox slot was 10.4 mm. Two side-grooves functioned to maintain directional stability of the crack. The depth of the Elox side-grooves was 0.610 mm and the diameter at the root of the groove was 0.102 mm.

FATIGUE PRECRACKING

Some specimens were tested with the notch in the as-Eloxed condition but, most

frequently, fatigue cracks were grown beyond the tip of the notch. In view of the amount of success in the controlled growth of fatigue cracks using reversed loading, which has been reported by Jones et al.¹ and by others,^{2,5} this technique was used in the present work. Fatigue cracks were produced by reversed loading, with the ratio of maximum-to-minimum applied load equal to -1. The controlled-displacement mode of a hydraulic materials-testing system was used, at a frequency of load reversal of 20 Hz. A minimization of unwanted displacements in the rigid load train was accomplished by using tight-fitting loading pins and pretorqued connections. Because no universal joints were used, care was taken in alignment

Fig. 1. A double cantilever beam (DCB) specimen. Dimensions are in mm.

to minimize eccentric loading on the specimens. The alignment of the loading grips allowed the loading pins a deviation from parallel of no more than 0.08 degree.

The specimens were fatigued in reversed loading for anywhere from 5×10^4 to 8×10^4 cycles at a preselected displacement. The procedure was repeated with progressively increasing displacements until a fatigue crack was initiated. These cracks were grown to about 1.27 mm in length. Maximum fatigue loads varied up to 844 N, which corresponded to a fatigue stress intensity K_f of approximately $6.5 \text{ MPa} \cdot \text{m}^{1/2}$ ($5.9 \text{ ksi} \cdot \text{in}^{1/2}$).

FRACTURE TOUGHNESS TESTING

Specimens were pulled to fracture in an 89 kN Instron machine at a crosshead

speed of $2.12 \times 10^{-3} \text{ mm/sec}$. The tensile load was measured with a 4.45 kN load cell. A clip gage measured the relative displacement of the knife edges at the loading end of the specimen. That is, the gage measured changes in the dimension of b in Fig. 1.

Specimens for the high temperature tests were heated in an argon atmosphere with a conventional Instron furnace. Temperature measurement was made by a platinum/platinum-rhodium thermocouple in contact with the specimen. The change of specimen temperature during a test never exceeded 2°C . The clip gage used for measuring displacement was made of a type 416 ferritic stainless steel. A strain-gage adhesive which would remain stable at temperatures up to 310°C was selected. The clip gage was calibrated at high temperatures (150°C and 300°C) by means of the specially designed apparatus shown in Fig. 2. A quartz rod and a concentrically fitted quartz tube were extended from the exterior to the interior of the Instron furnace through an access port. Outside the furnace, the tube was attached to the body of a barrel micrometer while the quartz rod was affixed to the spindle of the micrometer. In the interior of the furnace the two clip gage arms were seated on mating knife edges attached to the tube and rod respectively. Thus, rotation of the barrel micrometer resulted in a relative displacement of the rod and tube and thereby allowed calibration of the clip gage.

Specimen compliance was calibrated by the same procedure as that used for performing fracture-toughness tests except that the Elox slot was extended by machining. The 0.102 mm-wide slot

(Fig. 1) was increased in length in 1.27 mm increments and the corresponding relation between applied load and opening displacement was then recorded. Because opening displacements were measured at the end of the specimen rather than under the point of loading, a correction factor was applied to the apparent compliance to determine the actual specimen compliance. The correction, derived by Finnie and

published in Prado,⁶ is

$$C_{\text{actual}} = C_{\text{apparent}} - \frac{b\gamma}{Ebh} \left[(1 - \nu) + 2 \left(\frac{a + 0.6h}{h} \right)^2 \right]$$

where γ is the perpendicular distance from the center of the loading pins to the knife edges at the end of the specimen, E is Young's modulus, ν is Poisson's ratio, and b and h are the cantilever thickness and height respectively.

After establishing the compliance of the specimen as a function of crack length, the energy-release rate G is then computed using $dC/d\alpha$, through the standard relation

$$G = \frac{P^2}{2b_N} \left(\frac{dC}{d\alpha} \right)$$

where P is the applied load and b_N is the specimen thickness between the side-groove roots. The opening-mode stress intensity for plane strain is determined from G by

$$K_I = \left[\frac{GE}{1 - \nu} \right]^{1/2}$$

Because maximum fracture loads were used to compute the critical stress intensity factors, the toughness data are designated by K_{max} rather than by K_{IC} .

CRACK LENGTH DETERMINATION

After fatigue precracking or fracture-toughness testing, crack-length measurements were generally made on both sides of the specimen, using an optical microscope at 100X magnification. As will be discussed below, lapping the side grooves before testing allowed improved resolution of cracks. The lapping was performed

Fig. 2. Elevated-temperature clip gage calibration apparatus.

with 0.305 mm-thick brass shimstock and 600- and 900-grit alumina abrasive. The lapping deepened the side grooves by about 0.051 mm and somewhat sharpened the root. Because optical techniques of crack-length determination frequently gave results lacking consistency and reproducibility, other techniques were utilized, including ultrasonic, radiographic, and dye-penetrant methods. In addition, crack length determinations were made with a scanning electron microscope.

During testing, crack lengths were also obtained indirectly by comparing the experimentally-determined specimen compliance with the standard plot of compliance vs crack length.

FRACTOGRAPHY

Scanning electron fractographs were made of fatigue surfaces and of low- and high-temperature fast-fracture surfaces.

Results and Discussion

FATIGUE PRECRACKING

By applying fatigue stress intensities of $6.5 \text{ MPa} \cdot \text{m}^{1/2}$ ($5.9 \text{ ksi} \cdot \text{in}^{1/2}$) using reversed loading with the maximum compressive load equal to the maximum tensile load, fatigue cracks were generally observed to initiate and grow to approximately 1.27 mm in length in 8×10^4 cycles. At slightly smaller fatigue stress intensities, for example $5.6 \text{ MPa} \cdot \text{m}^{1/2}$ ($5 \text{ ksi} \cdot \text{in}^{1/2}$), no crack was observable after a similar number of cycles. It is very likely that this value for K_f does not satisfy the ASTM requirement of $K_f < 0.6 K_{IC}$ for this material. Difficulty was encountered in propagating fatigue cracks in a controlled manner at the stress intensities employed, and erratic growth rates were frequently observed.

It should be pointed out that when compressive loads are applied to specimens, the stress field near the crack tip is not necessarily compressive. A tensile-stress field can be generated in the vicinity of a compressively loaded material

discontinuity as a result of the applied load. Whether the stress is tensile or compressive may depend strongly on the length of the crack and on the geometry of the body. To illustrate this point, we can model the DCB specimen under applied compressive loads as an elastic beam on a rigid foundation (lower portion of Fig. 3). By the symmetry of the specimen and of the applied loads, the displacement in the Y direction at $x = a_i$ must vanish. The simple support at $x = a_i$ enforces this condition in the model. By examining the model at $x = a_i + \beta$ one can infer that an opening force is present at the tip of the crack in the cracked specimen due to the applied compressive load. If we had chosen as a model an elastic beam on an elastic foundation, this result would remain the same. Consequently, caution must be observed in applying large compressive loads to these specimens when employing reversed-loading techniques. The nature of the stress field at the crack tip is unknown and the extent of the damage at the crack tip from potentially large

tensile stress intensities is undetermined. This difficulty does not arise in reversed loading of bend specimens providing the fatigue crack does not cross the neutral axis of stress.

LOAD-DISPLACEMENT RECORDS

Figure 4 is a plot of the observed compliance (corrected for the location of displacement measurement) versus machined crack length. Also plotted is the analytic expression determined by Mostovoy⁷, i. e.

$$C = \frac{8}{Eb^3} \left[(a + 0.6h)^3 + \frac{3(1+\nu)ah^2}{4} \right]$$

Fig. 3. DCB specimen under a compressive load modeled as an elastic beam on a rigid foundation.

Fig. 4. Experimentally determined compliance calibration corrected for the location of displacement measurement.

The data for these plots are tabulated in Appendix A. It can be seen from Fig. 4 that the experimental and the analytical results agree very well. The compliance calibration at 300°C also agrees well with these results, indicating that between 23°C and 300°C compliance is not a function of temperature. The load-displacement records for these compliance tests were linear from small loads up to loads which were comparable to those used for crack propagation in a fracture-toughness test. Within the sensitivity of the clip gage, displacements were completely reversible after unloading the specimen.

For specimens with natural cracks, however, the load-displacement-record linearity depended strongly on the length of the crack. For cracks which extended only short distances (≤ 1.5 mm) beyond the tip of the Elox slot, the record was essentially linear up to the point of fracture, and satisfied ASTM linearity requirements. Nonlinearity became more pronounced for longer cracks. An example

Fig. 5. Load displacement record for specimen GL-4.

of this behavior is given in Fig. 5 which is the load-displacement record for specimen GL-4. For long crack-lengths, the curvature of the record decreases with increasing loads; linearity is most closely approximated at higher loads.

As shown in Fig. 5, curve f, if a subcritical load (that is, a load which was less than that required to extend the crack) was applied to a specimen with a long natural crack, and if the specimen was subsequently unloaded, the unloading curve did not coincide with the loading curve. This curve also indicates no appreciable residual displacement after a complete loading-unloading cycle.

Because the nonlinearity shown in Fig. 5 persisted over substantial fractions of the load deflection records, and because nonlinearity was never observed with machined cracks tested under identical conditions, it cannot reasonably be attributed to errors in the measurements of loads or displacements. Faulty seating of the clip gage or the loading pins must be ruled out as possible sources for this effect. Further, because the nonlinearity decreased with increasing loads and because there were no observable residual displacements on unloading from

subcritical loads, crack-tip plasticity and plastic flow in the vicinity of the loading pins must also be ruled out.

The nonlinearity manifested itself as increasing specimen compliance with increasing load. This phenomenon has been observed by Jones, Bubsey, and Brown⁴ in beryllium bend-specimens and has been discussed in Appendix C of their paper. These authors suggested that the nonlinearity was associated with the crack-closure phenomenon described by Elber.⁸ Elber has defined a crack as being closed when the compliance of the body is the same as the compliance of an identical but uncracked body under the same load system. Assume for the moment that a crack of length a in an unloaded elastic body is closed due to internal forces which exist in the body. Then, as loads are applied which overcome the closing forces and gradually open the crack, the compliance will also gradually increase from that of an uncracked body to that of a body with an open crack of length a .

If we postulate the existence of crack-closing forces, we can explain the dependence of compliance on applied load as exhibited by these specimens. The crack may be considered to consist of both open and closed parts; that is, as shown in Fig. 6, part a,

$$a = a_0 + a_c$$

As a result of the postulated closing forces, a_0 increases monotonically with P and the compliance smoothly increases with P , as shown in Fig. 6, part b. It is apparent that the nonlinearity would be especially prominent for DCB specimens with long cracks (a being several times larger than a_1).

The influence of crack closure on compliance in the DCB specimen was demonstrated by the following technique: A 0.102 mm-wide Elox slot was machined in a specimen to simulate a long ($a = 30$ mm), open crack. A tensile load 267 N was applied to the specimen. This would correspond to a stress intensity of $4.4 \text{ MPa} \cdot \text{m}^{1/2}$ ($4 \text{ ksi} \cdot \text{in}^{1/2}$) if it were applied to a specimen with a natural crack

Fig. 6. The influence of gradual crack opening on the compliance of the DCB specimen.

of the same length. The load-displacement record was linear. Shimstock slightly thicker than the slot was then placed between the separated arms of the specimen, with the stock wedged as close to the notch tip as interference would allow. Subsequent load-displacement records showed strong nonlinearities similar to that of Fig. 6, part b.

Measurements of opening displacements near the tip of a long (about 20 mm), natural crack were made as a function of applied load. The reference points for displacement measurement were attached rigidly to both arms of the DCB specimen at a small distance (about 2.5 mm) behind the crack tip. The signal from the high-sensitivity displacement transducer was displayed directly on a digital voltmeter. The resulting plot of load vs displacement, Fig. 7, indicates that substantial loads

Fig. 7. Load vs crack opening near the tip of a long natural crack.

(about 220 N) had to be applied before any resolvable displacement occurred. Further increases in load caused appreciable displacements. These data are consistent with the hypothesis that a crack-closing phenomenon did play a role in the observed nonlinearities.

The problem of the origin of the postulated crack-closing forces in a naturally cracked specimen remains to be explained. It is likely that the crack faces do not fit perfectly on unloading because of plastic deformation left in the wake of the crack. For beryllium, the amount of deformation must be small, but it may be sufficient to cause this effect. As the applied load was removed, the putative lack of fit would prevent the arms of the specimen from returning to their original position with respect to one another. Consequently, a residual compressive-force would be transmitted across the points of contact of the crack faces. In this manner, crack-face interference could generate internal forces leading to nonlinearities of the load-displacement record.

In these experiments, microcracking frequently occurred and revealed itself as small steps in the load-displacement record. (See, for example, Fig. 5, curve c). These events were detected by an acoustic transducer. No optically observable increase in crack length was produced by microcracking. As pointed out by Jones et al.,⁴ microcracking is to be expected in beryllium in the light of its very low basal-plane fracture propagation energy.

CRACK-LENGTH DETERMINATION

None of the several methods used in the determination of crack length in the

DCB specimen proved to be completely satisfactory. Initially, an endeavor was made to establish crack length on the basis of observations with an optical microscope. This method had several drawbacks. For example, in the as-Eloxed condition the surface of the root of the side groove proved to be unsatisfactory for crack observation due to its high reflectivity and unevenness, as shown in the top part of Fig. 8. Lapping the groove improved this condition, as shown in the bottom portion of Fig. 8, but accurate measurement remained difficult. Because the depth of focus was limited, only a narrow band of the side groove could be in focus at a time and cracks which wandered up the side of the grooves were difficult to locate and follow.

Ultrasonic methods of crack length determination were attempted. These methods occasionally indicated that the crack front was not straight. An example

As - eloxed condition (100 x)

After lapping and fatigue precracking (200 x)

Fig. 8. Root of side groove before and after lapping and fatigue precracking.

Fig. 9. Scan of reflected ultrasonic signal from plane of the crack for specimen GL-1. Reflection from side grooves interferes with crack observation.

of this is shown in Fig. 9, which is a record of the intensity of the reflected signal from the plane of the crack for specimen GL-1. Note that the reflected signal from the side grooves interferes with crack detection. Ultrasonic techniques and x-radiographic techniques both suffered from inadequate resolution in this application.

The scanning electron microscope was also used to check the measurements obtained by optical methods. Cracks were revealed which would not be visible in an optical microscope. One can infer from Fig. 10, which is a 10,000X electron micrograph of a crack in the root of a side groove, that optical measurements of crack length are likely to be biased toward short values. The large number of determinations required eliminated this as a routine method, however.

An indirect method of crack-length measurement involved comparing the experimentally determined compliance (defined as the inverse of the slope of the

linear portion of the load deflection record) with the standard plot of compliance vs crack length. For the cases in which the nonlinearity extended up to the critical load, the maximum compliance (corresponding to the minimum slope) was used. Of course, the assumption that maximum compliance correctly reflects specimen compliance can be valid only if plasticity does not play a role in causing the nonlinearity. If crack closure was the dominant source of the nonlinearity, then the use of maximum compliance is valid. The correlation between the longest of the two optically determined crack lengths (obtained from the two sides of the specimen), and the crack length as determined from the observed compliance, was superior to the correlation between the average of the optical measurements

Fig. 10. Scanning electron micrograph of a fatigue crack in the root of a side groove (10,000X).

Fig. 11. Mostovoy's solution with displacement measured at end of specimen; maximum crack length vs experimentally-observed uncorrected compliance; and average crack length vs experimentally-observed uncorrected compliance (Specimen GL-1).

and the compliance-crack length. This is demonstrated in Fig. 11 which plots (1) maximum crack length vs experimentally observed uncorrected compliance; (2) average crack length vs experimentally observed uncorrected compliance; and (3) Mostovoy's solution with displacement measured at the end of the specimen. As shown in Fig. 11, the superior correlation with maximum observed crack length was taken to be evidence that optically determined measurements were biased toward short values.

NUMERICAL VALUES OF FRACTURE TOUGHNESS

It was not possible to compute K_{IC} values from the recorded data because of the non-linearity of the load-displacement records.

Maximum loads were used to compute the critical stress intensity factors. As mentioned previously, the maximum load did not always correspond to the initiation of detectable cracking during a test. If acoustic-emission techniques had been employed in the determination of load for the calculation of fracture toughness, significantly smaller values of fracture toughness would have been derived. As discussed above, specimen compliance was used as an independent check on crack length. Occasionally, the discrepancy between the length of the crack as determined optically and the length as determined from specimen compliance exceeded 2.5 mm. When this occurred, the uncertainty in crack length (i. e. the parameter a) was considered to be unacceptable and the data from that test were ruled to be inadmissible. The experimental data and the computations of K_{max} are tabulated in Appendix B. Values of K_{max} vs temperature for both transverse and longitudinal specimens are plotted in Figs. 12 and 13 respectively.

No significant differences are observed for K_{max} between the transverse and longitudinal orientations. The fracture-toughness values may increase slightly with increasing temperature but the data are inconclusive in view of the scatter. The toughness values for room temperature tests range from 6.8 to 11.0 $MPa \cdot m^{1/2}$ (6.1 to 9.9 $ksi \cdot in^{1/2}$). The average of 24 room temperature tests is 9.0 $MPa \cdot m^{1/2}$ (8.1 $ksi \cdot in^{1/2}$). These values are among the lowest that have ever been reported for beryllium.

Figure 14 is a plot of K_{max} vs crack length for the transverse specimens at 23°C. An increasing trend in K_{max} with

Fig. 12. K_{max} vs temperature for transverse specimens.

Fig. 13. K_{max} vs temperature for longitudinal specimens.

Fig. 14. K_{max} vs crack length for transverse specimens at 23°C.

crack length is apparent for 10 mm $< a < 15$ mm. A similar trend has been observed by Prado.⁶ In that report, the K_{max} values derived for the first two or three data points generated from each DCB specimen were neglected. The implied justification for omitting these points was the possible crack-tip damage induced by the wedge-loading technique which was used to initiate cracks. Using this technique, cracks were generated by impact wedging in a machined notch. The crack was arrested under an unspecified compressive load which was applied ahead of the notch. Because a trend of increasing K_{max} with crack length occurred in our studies when wedge loading was not used, alternate explana-

tions must be considered. One explanation is that discrepancies of this nature could be due to systematic errors in the compliance calibration. That is, if the computed values for dC/da (which is proportional to K^2) were systematically underestimated for 10 mm $< a < 15$ mm, low K values for this range of crack lengths would result. However, the close correspondence of the experimental compliance-calibration curve and Mostovoy's solution (Fig. 4) renders this explanation improbable.

Two other possible explanations would be the systematic overestimation of crack length for long cracks, and eccentric specimen loading. A fourth possibility is that the value of the stress intensity

factor, K , is influenced at fracture by the non-singular terms in the series expansion for stresses near the tip of a crack. It has been suggested by Finnie and Smith⁹ that the second term in the series expansion has an influence on the directional stability of cracks. Larsson and Carlsson¹⁰ have demonstrated that for the same K values, different types of specimens will have varying sizes of plastic zones in the case of small-scale yielding at the crack tip. Rice¹¹ has shown that in the series expansion the magnitude of the second term relative to the first, or singularity term, is responsible for this effect. There is at this stage little clear-cut experimental evidence to show the influence of subsequent terms in the series expansion on fracture toughness. However, in the double-cantilever-beam test, the relative value of the second series-expansion term, relative to K , increases rapidly as crack length increases. The effect of this aspect of the specimen on fracture toughness values needs further attention.

A simple experiment could test the proposition that crack tip damage causes the dependence of K_{max} on crack length. In this experiment, a long Elox slot would be machined in a specimen so that the initial crack-length would be greater than those crack lengths for which the dependence was observed. If subsequent fracture-toughness testing revealed no dependence of K on a , irrespective of how the starter crack was generated, then crack-tip damage would have to be discounted as a potential cause of the dependence. If, on the other hand, a dependence was observed, crack-tip damage would be indicated.

That the distribution of points in Fig. 12 may not be a true indication of scatter

can be seen in Fig. 14. For any given crack length in Fig. 14, the extreme values do not differ by more than 2.2 MPa·m^{1/2} (2 ksi·in^{1/2}). Returning to Fig. 12, in which crack length is an implicit variable, we note that the extreme values for the same set of data differ by 4.4 MPa·m^{1/2} (4 ksi·in^{1/2}).

A limited study was performed at room temperature to determine the difference between fracture toughness of specimens in the fatigue-precracked and in the as-Elox conditions. The results are presented in Fig. 15, which is a plot of fracture toughness vs notch-root radius. (It is assumed in constructing the figure that the notch-root radius of fatigue-precracked specimens is vanishingly small on this scale.) No significant influence of notch-root radius on K_{max} can be detected with these data. Data for a particular test were ruled to be inadmissible if a crack became directionally unstable and caused arm failure. Generally, the maximum loads reached during tests in which arm failure occurred were significantly larger than those reached in tests in which the crack remained in the side groove. Arm failure occurred frequently for specimens tested in the as-Elox condition. Also, there was apparently a greater tendency to arm failure at elevated temperatures than at ambient temperatures. In about half of the tests performed at elevated temperatures, arm failure occurred on the first attempt. Cracks in these specimens occasionally did not arrest. For example, specimen GL-6 (Appendix B) was tested at 150°C with a 1.27 mm long fatigue crack. The crack stayed in the side groove, did not arrest, and

Fig. 15. Influence of notch preparation on fracture toughness (K_{max} vs notch root radius).

propagated through the entire length of the specimen.

FRACTOGRAPHY

An examination of fatigue- and fast-fracture surfaces was made using a scanning electron microscope. In Fig. 16 (taken at 500X) a fatigue area shows isolated regions in which cleavage planes make large angles with respect to each other. The remainder of the fatigue surface appeared to be highly textured and

Fig. 16. Scanning electron micrograph of a fatigue region (500X).

Fig. 17. Scanning electron micrograph of a fatigue region (5000X).

rough, as can be seen in Fig. 17 (taken at 5000X). The various facets in these micrographs exhibit a wide range of sizes. There is apparently some debris on the fatigue surface. It is possible that the peculiar appearance of these fatigue surfaces could, in large part, result from damage accumulated during the reversed loading procedure.

Fractographs of fast-fracture areas of specimens tested at room temperature and at 302°C are shown in Fig. 18 and

Fig. 18. Scanning electron micrograph of a fast fracture region at 23°C (1000X).

Fig. 19. Scanning electron micrograph of a fast fracture region at 302°C (1000X).

Fig. 19, respectively. Both fractographs reveal predominantly flat transgranular cleavage. There were no apparent differences between room-temperature and elevated-temperature fast-fracture surfaces.

The poor homogeneity of the grade of beryllium being tested is demonstrated in Fig. 20 in which can be seen evidence of foreign material which had been included in the hot pressed block. X-radiographs taken of these specimens

Fig. 20. Scanning electron micrograph of an inclusion which has been intersected by a rapidly propagating crack (1000X).

exhibited a substantial density of strongly absorbing particles, giving the radiographs a stippled appearance. The

scatter in the fracture toughness data generated in this study could in large part be due to these inhomogeneities.

Conclusions

From the results of this study, the following conclusions can be made:

- Using reversed-loading procedures, fatigue cracks 1.27 mm in length were grown (typically, in 8×10^4 cycles) at a fatigue stress intensity of $6.5 \text{ MPa} \cdot \text{m}^{1/2}$ ($5.9 \text{ ksi} \cdot \text{in}^{1/2}$). Cracks would not initiate or grow at lower stress intensities. Uncertainty as to the extent of plasticity and twinning in the vicinity of the crack tip may limit the usefulness of this technique in standard fracture-toughness testing.
- The values of $K_{I\max}$ obtained from specimens with machined notches did not differ significantly from those obtained from fatigue-precracked specimens.
- The specimens exhibited nonlinear load-displacement records. The extent of the nonlinearity depended strongly on crack length. For short cracks, the load-displacement records satisfied ASTM requirements for linearity. Long cracks exhibited increasing compliance as applied loading increased.
- The fracture-toughness data generated indicate that for tem-

peratures between 23°C and 300°C the fracture toughness of the material tested is only a weakly increasing function of temperature. At 23°C the mean value of toughness from 24 tests is $9.0 \text{ MPa} \cdot \text{m}^{1/2}$ ($8.1 \text{ ksi} \cdot \text{in}^{1/2}$) with a lower extreme of $6.8 \text{ MPa} \cdot \text{m}^{1/2}$ ($6.1 \text{ ksi} \cdot \text{in}^{1/2}$). At 300°C the mean value of toughness from 7 tests is $10.8 \text{ MPa} \cdot \text{m}^{1/2}$ ($9.7 \text{ ksi} \cdot \text{in}^{1/2}$) with a lower extreme of $9.7 \text{ MPa} \cdot \text{m}^{1/2}$ ($8.5 \text{ ksi} \cdot \text{in}^{1/2}$). The room-temperature values are among the lowest ever reported for beryllium.

- Subcritical crack growth (microcracking) occurred at stress intensities as low as 60% of the critical stress intensity.
- There is no appreciable orientation effect for this material. At room temperature, the mean value of toughness from 18 tests of transverse specimens is $9.1 \text{ MPa} \cdot \text{m}^{1/2}$ ($8.2 \text{ ksi} \cdot \text{in}^{1/2}$) whereas the mean value of toughness from 7 tests of longitudinal specimens is $8.6 \text{ MPa} \cdot \text{m}^{1/2}$ ($7.8 \text{ ksi} \cdot \text{in}^{1/2}$).

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Appendix A

Compliance Calibration

The following data are used in Fig. 4 to compare the experimentally observed specimen compliance and the analytically-determined compliance from Mostovoy's solution.

Table A-1. Compliance calibration data.

a (mm)	C_o ($\mu\text{m}/\text{kN}$)	C_c ($\mu\text{m}/\text{kN}$)	$C_o - C_c$ ($\mu\text{m}/\text{kN}$)	C_m ($\mu\text{m}/\text{kN}$)
10.3	23.5	5.89	17.6	17.4
11.7	29.7	6.75	23.0	22.0
12.9	33.7	7.61	26.1	26.7
15.5	46.9	9.61	37.3	38.3
18.1	64.0	11.9	52.2	53.2
20.5	85.8	14.2	71.5	70.4

Compliance as observed on P- δ record:

$$C_o$$

Compliance correction for location of
displacement measurement:

$$C_c = \frac{6\gamma}{Eb^3} \left[2(a+0.6h)^2 + (1+\nu)h^2 \right]$$

Mostovoy's expression for DCB Compliance: $C_m = \frac{8}{Eb^3} \left[(a+0.6h)^3 + \frac{3(1+\nu)ah^2}{4} \right]$

where

$$\nu = 0.03 \quad h = 9.53 \text{ mm}$$

$$E = 289 \text{ GPa} \quad \gamma = 3.58 \text{ mm}$$

$$b = 8.92 \text{ mm}$$

Appendix B

Calculation of K_{max}

The following data are used to calculate K_{max} :

Table B-1. Tabulation of experimental results.

Specimen	T (°C)	a (mm)	C_o ($\mu M/kN$)	P_{max} (kN)	K_{max}		Remarks
					($MPa \cdot m^{1/2}$)	($ksi \cdot in^{1/2}$)	
GL-1	23	12.7	37	0.890	7.7	6.9	Arm failure uncertain
	23	16.3	50	0.917	9.4	8.5	
	23	17.7	66	0.872	9.5	8.6	
	23	19.4	69	0.828	9.5	8.6	
GL-2							Compliance calibration specimen
GL-3	23	11.3	23	1.05	8.4	7.6	Arm failure occurred at $a = 14.0$, making subsequent tests invalid
GL-4	23	12.7	23	0.992	8.4	7.6	Arm failure
	23	20.3	69	0.926	11.0	9.9	
	23	19.9	80	0.837	10.0	9.0	
	23	22.7	103	0.837	11.0	9.9	
	292	24.4	132	0.926	12.3	11.1	
	286	34.3	286	0.596	11.0	9.9	
GL-5	23	16.8	51	0.935	9.7	8.7	Arm failure uncertain
	23	17.5	69	0.939	10.1	9.1	
GL-6	150	11.8	26	1.12	9.3	8.4	Crack did not arrest; stayed in side grooves entire length of specimen
GL-7							Specimen used in crack-length-determination tests (scan, x-ray and dye penetrant tests)

Table E-1 (continued)

Specimen	T (°C)	a (mm)	C _o (μM/kN)	P _{max} (kN)	K _{max}		Remarks
					(MPa·m ^{1/2})	(ksi·in ^{1/2})	
GL-8							Data not reliable; possible calibration error. Also, load train not aligned properly
GL-9	23	11.9	34	0.881	7.3	6.6	
	23	15.2	51	1.05	10.2	9.2	Crack leaves side groove. Subsequent tests invalid
GL-10	23	11.9	25	0.908	7.5	6.8	
	301	13.5	40	1.19	10.8	9.7	
	30	30.5	183	0.579	9.7	8.7	Arm failure
GL-11	23	10.4	25	0.890	6.8	6.1	
	23	11.7	34	0.952	7.9	7.1	
	303	15.0	34	0.997	9.4	8.5	
	300	17.0	74	1.13	11.8	10.6	Arm failure
GL-12							Intact, with fatigue crack. Saved for possible future testing
GL-13							Tested at 210°C without observing crack. Initial crack length not consistent with initial compliance. Arm failure
GL-14		21.0	86	0.828	10.1	9.1	Intact, with an arrested crack
GL-15							Intact, with fatigue crack. Saved for possible future testing
DE 1T							Arm failure occurred before an observable crack appeared in side groove

Table B-1 (continued)

Specimen	T (°C)	a (mm)	C _o (μM/kN)	P _{max} (kN)	K _{max}		Remarks
					(MPa·m) ^{1/2}	(ksi·in) ^{1/2}	
DE 2T							Substantial micro-cracking occurred with no measurable crack growth. Crack leaves side grooves and re-enters. Data not reliable.
DE 3T							Arm failure occurred when testing machined notch at 300°C
DE 4T							Specimen destroyed by machinist
DE 5T							Intact. Saved for possible future testing
DCO 6T		10.4	23	1.03	8.0	7.2	
		10.8	37	1.03	8.3	7.5	
		13.5	49	1.12	10.2	9.2	Arm failure
DCO 7T		10.4	23	1.26	9.8	8.8	Arm failure
DCO 8T							Intact. Saved for possible future testing
DCO 9T	23	10.4	23	1.19	9.2	8.3	
	298	13.6	46	1.09	10.1	9.1	
	23	13.6	57	1.03	9.5	8.6	Arm failure
DCO 10T	23	10.4	23	0.970	7.5	6.8	
	23	10.7	29		7.9	7.1	Arm failure

$$K_{\max} = \left[\frac{E}{2b_N(1-\nu^2)} \right]^{1/2} \left(\frac{dC}{da} \right)^{1/2} P_{\max}$$

where $\frac{dC}{da} = \frac{8}{Ebh^3} \left[3(a + 0.6h)^2 + \frac{3(1+\nu)h^2}{4} \right]$

and $E = 289 \text{ GPa}$ $b = 8.92 \text{ mm}$
 $\nu = 0.03$ $h = 9.53 \text{ mm}$
 $b_N = 7.70 \text{ mm}$