Lawrence Berkeley National Laboratory
Recent Work

Title
UNIAXIAL BEND STRENGTH AND FRACTURE INITIATION ENERGY IN GLASS-ALUMINA COMPOSITES

Permalink
https://escholarship.org/uc/item/1pg3f308

Author
Biswas, Dipak Ranjan.

Publication Date
1974-02-27
UNIAXIAL BEND STRENGTH AND FRACTURE INITIATION ENERGY IN GLASS-ALUMINA COMPOSITES

Dipak Ranjan Biswas
(M.S. thesis)

February 27, 1974

Prepared for the U. S. Atomic Energy Commission under Contract W-7405-ENG-48

TWO-WEEK LOAN COPY
This is a Library Circulating Copy which may be borrowed for two weeks. For a personal retention copy, call Tech. Info. Division, Ext. 5545
DISCLAIMER

This document was prepared as an account of work sponsored by the United States Government. While this document is believed to contain correct information, neither the United States Government nor any agency thereof, nor the Regents of the University of California, nor any of their employees, makes any warranty, express or implied, or assumes any legal responsibility for the accuracy, completeness, or usefulness of any information, apparatus, product, or process disclosed, or represents that its use would not infringe privately owned rights. Reference herein to any specific commercial product, process, or service by its trade name, trademark, manufacturer, or otherwise, does not necessarily constitute or imply its endorsement, recommendation, or favoring by the United States Government or any agency thereof, or the Regents of the University of California. The views and opinions of authors expressed herein do not necessarily state or reflect those of the United States Government or any agency thereof or the Regents of the University of California.
UNIAXIAL BEND STRENGTH AND FRACTURE INITIATION ENERGY IN GLASS-ALUMINA COMPOSITES

Table of Contents

ABSTRACT
I. INTRODUCTION
II. EXPERIMENTAL PROCEDURE
   A. Preparation of Powders
   B. Fabrication of Specimens
   C. Uniaxial Bend Strength Measurement
   D. Fracture Initiation Energy Measurement
   E. Microstructure and Scanning Electron Micrograph of the Specimen
III. RESULTS AND DISCUSSION
IV. SUMMARY
ACKNOWLEDGMENT
REFERENCES
UNIAXIAL BEND STRENGTH AND FRACTURE INITIATION ENERGY IN GLASS-ALUMINA COMPOSITES

Dipak Ranjan Biswas

Inorganic Materials Research Division, Lawrence Berkeley Laboratory and Department of Materials Science and Engineering, College of Engineering; University of California, Berkeley, California 94720

ABSTRACT

Uniaxial bend strength and fracture initiation energy of a glass-alumina composite system are measured by 4-point bending and double-cantilever-beam methods. The composites are fabricated by vacuum hot-pressing technique. The second phase alumina particles increase the strength and the load necessary for failure in the double cantilever test. The two test methods are discussed making the assumption that the presence of microflaws introduced by diamond sawing at the crack tip of the double-cantilever-beam specimen leads to the similar stress at the point of fracture. The tendency of fine alumina particles to agglomerate reduces the strength of the composite approximately 15% from that without agglomeration.
I. INTRODUCTION

In general, a composite material is a material with several distinct phases. Normally the composite consists of a reinforcing or dispersed phase (e.g. fibers, whiskers, or particulates) supported in a binder or matrix phase. The reinforcing material in composites for structural use is the load bearing medium and the matrix serves as a protector and load transferring medium.

Most composite materials have been developed to improve mechanical properties such as stiffness, toughness and high temperature strength. There are a considerable number of technological important composite materials, particularly among the fibrous composites. Ceramic materials play an important role in many of the composite systems where they are used as a matrix or as a dispersed phase in particulate, whisker, or fiber form.

For a number of years, Fulrath and co-workers have studied the mechanical properties of brittle matrix composites using a model composite system consisting of a glass matrix that is combined with various metal or ceramic particles, as well as nearly spherical voids as the dispersed phase. The coefficient of thermal expansion for glasses varies with composition while the strength variation with composition is small. The main parameters that have been found to affect the mechanical properties of this type of model system are:

1. The volume fraction and the particle size of the dispersed phase,
2. The difference in elastic modulus leading to micromechanical stress concentrations,
(3) The degree of chemical bonding between the matrix and the dispersed phase,

(4) The internal stresses developed due to the difference in thermal expansion between the glass and the dispersed phase,

(5) The porosity and

(6) The surface damage during the preparation of the specimens.

The theoretical strength of a material can be calculated from its molecular constitution and structure. The value obtained in this way is much greater than that observed in a laboratory experiment. The observed strength of ordinary glass is less than one-hundredth of its theoretical strength. This discrepancy between the theoretical and observed strength was first explained by Griffith. He postulated that the low observed strengths were due to the presence of small cracks or flaws in the glass and at the tip of the cracks there is a strong concentration of stresses under applied load. He calculated the rate at which crack growth releases energy and compared with the energy needed to form the new fracture surfaces. He found for plane stress that the critical stress required for failure of a flat plate containing an elliptical flaw is

$$\sigma = \sqrt{\frac{2E\gamma}{\pi c}}$$  \hspace{1cm} (1)

where \(\gamma\) = surface energy per unit area,

\(E\) = Young's modulus of elasticity, and

\(c\) = twice the major axis of an ellipse, i.e. the flaw size.

The surface energy of a material plays an important role in fracture and many investigations have been made to measure and interpret the
meaning of $\gamma$. In general, the surface energy is defined as the energy necessary to create unit area of fresh surface due to rearrangement of bonding forces and can be measured in a number of ways unrelated to fracture. When the value of $\gamma$ is determined by use of the Griffith equation, it is often termed the fracture energy $\gamma$—defined similarly as the energy necessary to create unit area of fresh surface during the fracture process. Two values of $\gamma$ are of particular interest in fracture: $\gamma_I$, pertaining to the initiation of fracture, and $\gamma_P$, that of the propagation of the fracture. The energies for crack initiation and crack propagation are not necessarily the same, as the latter varies with the crack velocity. $\gamma_I$ is related to the strength of the material via the Griffith equation and $\gamma_P$ is related to the fracture propagation characteristics and the crack stopping properties of the material. The experimental procedure available for the measurement of the fracture energy includes the compliance method, the work-of-fracture method, the double-cantilever-beam (DCB) method, and the notched-beam technique. The DCB method has been used for single crystals with well defined cleavage planes by Gilman. He measured the work that is required to cause cleavage per unit area of the cleaved surface. The energy of the new surfaces that were created when the crack moves forward a small distance was found by forming an energy balance for the whole system. This technique is applicable for a slowly moving crack. The energy balance is made at the critical point when crack motion just begins. The equation derived for the surface energies of the cleavage plane in single crystals is
where $F =$ critical force for crack propagation
$L =$ length of the crack
$E =$ Young's modulus of elasticity
$W =$ width
$2t =$ height of the specimen.

The DCB method was also used by Wiederhorn$^{20}$ for glasses which are isotropic and homogeneous materials. The equation used to calculate the fracture energy $\gamma$ was a modified equation$^{24,25}$ of Eq. (2). The same method was used by Lange$^{23}$ to measure the fracture energy of a sodium borosilicate glass-alumina composite system using the same equation as Wiederhorn. The modified equation is of the form:

$$\gamma = \frac{6F^2L^2}{EW^2t^3} \left[ 1 + 1.32 \frac{t}{L} + 0.542 \left( \frac{t}{L} \right)^2 \right]$$  \hspace{2cm} (3)

The glass-alumina composite system is a non-homogeneous system. For this two-phase composite system, a fracture theory was proposed by Hasselman and Fulrath.$^6$ They hypothesized that hard crystalline dispersions within the glass matrix limit the size of the Griffith flaws and thus strengthen the composite. Quantitative relationships were derived for the effect of a dispersed phase on the composite's uniaxial bend strength. The strength of the composite was a function of the volume fraction of the dispersed phase at low volume fractions and
dependent on both volume fraction and the particle size of the dispersed phase at higher volume fraction. For verification of the theory, they used D-glass (70% SiO₂, 16% Na₂O and 14% B₂O₃) and spherical alumina particles and the composites were prepared by vacuum hot-pressing the mixed glass and alumina powders. The volume fraction and the particle size of the dispersed alumina was varied. Cross-bending strengths of the composites were measured and the strength was plotted against the inverse square root of the calculated mean free path between the alumina particles (Fig. 1). The plot showed two distinctly different regions. In region I, the average distance between the particles was greater than the flaw size induced by diamond sawing and the strength was independent of the second phase. In region II, the flaw size was governed by the mean free path between the particles and the strength was a function of volume fraction and particle size of the alumina particles. When the mean free path is equal to the original flaw size, a discontinuity in slope of the strength vs reciprocal square root of the mean free path curve occurred. Therefore the conclusion was that the limitation of Griffith flaws in the glass matrix by the second phase alumina particles leads to the strengthening of the composite system.

On the other hand, Lange used similar glass-alumina composites and measured the bending strength and the fracture energy of the composites. He prepared the glass-alumina composites with non-spherical commercially produced alumina particles in a similar manner to that described by Hasselman and Fulrath. Fracture surface energies were determined in liquid nitrogen using the double-cantilever-beam technique. Four-point flexural strength was measured at room temperature. He attempted to
Fig. 1. Strength plotted against the inverse square root of the calculated mean free path between the alumina particles.
(From D. P. H. Hasselman and R. M. Fulrath, J. Am. Ceram. Soc., 49 (2) 68-72 (1966).)
determine the relationship between the fracture energy, uniaxial bend strength, and microstructure and compared the bend strengths with those measured by Hasselman and Fulrath. The flaw size \( c \) was calculated from a modified Griffith equation using the bending strength from the 4-point bending and the fracture energy from the double-cantilever test. The crack size for the composite was found to increase from that of the glass without a dispersion by about 1 to 3 times the average particle size of the dispersed phase. The crack size was not significantly affected by the volume fraction of the alumina dispersion. This is in conflict with the proposed fracture theory for strengthening of glass-alumina composites by Hasselman and Fulrath. They showed the flaw size is controlled by the volume fraction and particle size of the alumina.

Recently Evans proposed the line tension effect for the strength of brittle materials with a second phase dispersion. The line tension contribution to strength or fracture surface energy occurs when a crack bows out between the obstacles just as a dislocation bows between obstacles. He analyzed the data from Lange and Hasselman and Fulrath from the "line tension effect" concept and showed that the fracture surface energy of a glass-alumina system varies in a similar manner to the calculated line tension contribution. The line tension was the primary contribution to the fracture surface energy. This concept was also used to calculate the fracture strength of glass-alumina composites and he found that for a large obstacle spacing the strength was independent of the inter-obstacle spacing and similar to the strength of the matrix. This agreed well with the experimental data of Hasselman and Fulrath.
The present work was undertaken to measure the uniaxial bend strength and fracture initiation energy for a glass-alumina composite system and to compare the experimental data with that reported by Lange\textsuperscript{23} and Hasselman and Fulrath.\textsuperscript{6} In this work, the test techniques were those used by Lange\textsuperscript{23} and the sample preparation was the same as that of Hasselman and Fulrath.\textsuperscript{6} Instead of commercial alumina particles, this study used size separated single crystal particles obtained by crushing sapphire boules.
II. EXPERIMENTAL PROCEDURE

A. Preparation of Powders

The specimens used for the measurement of uniaxial bend strength and fracture energy consisted of a glass matrix containing crushed sapphire particles. The glass used was of the same composition as the D-glass (70% SiO₂, 16% Na₂O and 14% B₂O₃) used in previous investigations in this laboratory. This particular glass has a coefficient of thermal expansion almost identical with that of alumina which avoided internal stresses in the system.

The glass was prepared by mixing the component oxides (silica, anhydrous sodium carbonate, and anhydrous borax) in a porcelain ball mill containing alumina balls. Isopropyl alcohol was used as a liquid medium for ball milling. After ball milling the suspension was dried and then the mixed oxides were melted in a platinum crucible at 1400°C for 4 hours in air. The glass was cooled and ground to fine powder (average particle size * was about 5-6 µm) in a ball mill using isopropyl alcohol.

The alumina used in this work was prepared from the sapphire single crystal boules. ** The boules were ground to fine powders and then carefully separated into three different size fractions by using the precision screen in an ultrasonic sifter. # Some difficulties were encountered because of the tendency of alumina particles to agglomerate.

The volume fraction and the particle size are necessary to calculate the mean free path between the alumina particles. The assumption made

* Fisher subsieve sizer.
** Union Carbide Corp.
# Allen-Bradley Sonic Sifter.
in the present work was that there was a constant mean free path between the dispersed phase particles. In order to maintain the constant mean free path, three different series of glass-alumina composites (in region II, Fig. 1) were prepared after calculating the proper volume fraction and the particle size of alumina for each series from the following equation by Fullman:

\[
d = \frac{4R(1-\phi)}{3\phi}
\]  

(4)

where \(d\) = mean free path between the particles,
\(R\) = radius of the particles (in this case the average particle size), and
\(\phi\) = volume fraction of alumina.

B. Fabrication of Specimens

For the fabrication of 2" dia. x 3/8" thick compacts, the desired proportions of glass and alumina powders were intimately mixed in a ball mill using isopropyl alcohol for 4 hrs and then dried to evaporate the alcohol. The proper weight of the dry powder was then loaded in a graphite die lined with Grafoil and cold pressed at 1000 psi before loading into the hot-press.

Vacuum hot-pressing was carried out at 725°C and 1000 to 2000 psi pressure for 20 to 40 minutes, depending upon the composite. After hot pressing, bulk densities of all the compacts were measured in water by the Archimedes technique was found to be between 95 to 98% of the

* Product of Union Carbide.
theoretical density. Compacts containing more than 40 v/o alumina could not be pressed to full density, probably due to particle-particle contact.

The compacts were then glued to high-alumina porcelain ceramic plates and cut into strength specimens approximately 0.05 inch thick with a precision diamond saw using kerosene as a coolant. After cutting, the specimens were removed from the plate by acetone which dissolved the glue. From the same disc, some of the pieces were used for the uniaxial bend tests and the rest were used for the double-cantilever test. The DCB specimens were prepared by cutting the pieces into proper length and breadth by a thin diamond blade and carefully drilling two holes at one end of the specimen with a diamond core drill. A thin slot approximately 0.016 inch wide was cut in the specimen with a thin diamond blade. The approximate dimensions for the DCB test specimen are shown in Fig. 2.

The dimensions were measured by an optical microscope. The ratio of L/t was kept between 1.9 to 2.3 for all specimens. Although 16 specimens were cut from each hot-pressed disc at the start, only 8 to 14 were useful for the double-cantilever test.

C. Uniaxial Bend Strength Measurement

Uniaxial bend strength was determined by loading the specimens to failure by using a 4-point bending machine with a 0.75" overall span. The diamond sawed surface was stressed. The bending strength was calculated from the following equation:

$$\sigma = \frac{PL}{bd^2}$$

(5)
Fig. 2. Specimen configuration for the double-cantilever-beam test.

Thickness $\approx 0.05''$

$1.9 \leq L/t \geq 2.3$
where \( P \) = load necessary to fail the specimen
\( L \) = overall span
\( b \) = width and
\( d \) = thickness of the specimen.

The equation was derived as follows:

Bending moment \( M = \frac{PL}{6} \)

Moment of inertia \( I = \frac{1}{12} bd^3 \)

Bending strength, \( \sigma = \frac{M}{I} \cdot \frac{d}{2} = \frac{PL}{6} \cdot \frac{12}{bd^3} \cdot \frac{d}{2} = \frac{PL}{bd^2} \)

The width and the thickness were measured at the point of failure of the specimen. An average of 6 to 16 specimens were used for this bending test.

D. Fracture Initiation Energy Measurement

An attempt was made to measure the fracture initiation energy of the glass-alumina composite system using the double-cantilever-beam technique. The fracture initiation energy was calculated using Eq. (3). This glass-alumina composite system is non-homogeneous.

The specimen (Fig. 2) was loaded in tension as a double-cantilever in an Instron machine with a constant cross-head speed. The sample failed catastrophically at a critical load. A typical load deflection curve is shown in Fig. 3a. The load (at \( P_f \)) obtained was for the initiation of the crack. When this load is used in Eq. (3), the \( \gamma_f \)-fracture initiation energy is obtained. The effect of hardness of the testing machine for different materials was shown by Tattersall and Tappin. An extremely hard machine with controlled crack growth will give a load-deflection curve like that shown in Fig. 3b. The work of fracture \( \gamma_p \).
Fig. 3. Typical load-deflection curves.
can be determined by measuring the energy required for failure by integration of the load-deflection curve and dividing the energy by the area of the fracture face produced. The tail of the curve can be used to calculate the energy necessary to move a crack.

E. Microstructure and Scanning Electron Micrograph of the Specimen

Microstructure of the specimens was obtained by grinding the specimen on glass plate containing different sizes of SiC powders and finally polishing using 6μ and 1μ diamond on a Syntron Vibratory polisher. A scanning electron microscope was used to view the fracture surfaces of the fractured specimens from the double-cantilever test.
III. RESULTS AND DISCUSSION

The properties of the glass-alumina composites are summarized in Table I and the values of uniaxial bend strength and fracture initiation energy with their standard deviations for various compositions and particle sizes are given in Table II. The mean strength of the glass was found to be 14,400 psi determined experimentally from a hot-pressed glass specimen. Figure 4 shows the strength plotted against the volume fraction of alumina. It was found that the strength of the composite is increased with increase in volume fraction of alumina. The composite containing larger particle size and higher volume fraction of alumina (30-37 μm and 47 v/o alumina) shows lower strength because of the presence of greater porosity. The coefficient of variation of the uniaxial bending strength of the composite is less than that of the glass itself. With few exceptions, the coefficient of variation generally decreases with increasing volume fraction of the dispersed alumina similar to that reported by Hasselman and Fulrath.

Figure 5 shows the fracture initiation energy values calculated from Eq. (3) plotted against the volume fraction of alumina. The for borosilicate glass is found to be 6400 ergs/cm from the DCB test at room temperature (close to that measured in liquid N2 by Lange). Figure 5 shows a definite increase in values with increase in volume fraction of alumina. The coefficient of variation is quite scattered and does not follow a definite trend. The question arises whether the DCB method can be applied for this non-homogeneous system with catastrophic failure of the DCB specimens. Typical microstructures of 24.4 v/o and 42 v/o alumina-glass composites are shown in Fig. 6(a,b,c, and d).
Table I. Properties of the Glass-Alumina Composites

<table>
<thead>
<tr>
<th>Reciprocal square root of mean free path (d^{-1/2})</th>
<th>Glass(^1)</th>
<th>0.18</th>
<th>0.20</th>
<th>0.22</th>
<th>Alumina(^2)</th>
<th>8.0 \times 10^{-6} \text{ in/in per } ^\circ\text{C}</th>
<th>3.98</th>
</tr>
</thead>
<tbody>
<tr>
<td>Glass(^1)</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>98</td>
<td>11.7</td>
<td></td>
</tr>
<tr>
<td>0.18</td>
<td>10-20(\mu)m</td>
<td>0.244</td>
<td>&gt;97</td>
<td>16.5</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>20-30(\mu)m</td>
<td>0.350</td>
<td>98</td>
<td>19.3</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>30-37(\mu)m</td>
<td>0.420</td>
<td>96</td>
<td>21.6</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>0.20</td>
<td>10-20(\mu)m</td>
<td>0.30</td>
<td>98</td>
<td>17.8</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>20-30(\mu)m</td>
<td>0.40</td>
<td>&gt;97</td>
<td>21.0</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>30-37(\mu)m</td>
<td>0.47</td>
<td>95</td>
<td>23.3</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>0.22</td>
<td>10-20(\mu)m</td>
<td>0.326</td>
<td>98</td>
<td>18.8</td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>20-30(\mu)m</td>
<td>0.446</td>
<td>97</td>
<td>22.8</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Average particle size of alumina</th>
<th>Volume fraction of alumina</th>
<th>Density of the composites after hot pressing(^3)</th>
<th>Elastic modulus(^4) psi(\times)10^{-6}</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>&gt;97</td>
<td>21.0</td>
</tr>
</tbody>
</table>

- Coefficient of thermal expansion (room temp. to 450\(^\circ\)C) gms/cc

- (1) Glass 7.8 \times 10^{-6} \text{ in/in per } ^\circ\text{C} 2.48

- (2) Alumina 8.0 \times 10^{-6} \text{ in/in per } ^\circ\text{C} 3.98

- (3) Hot pressing temp = 725\(^\circ\)C

- and pressure = 1000-2000 psi

- (4) From Ref. 4
Table II. Uniaxial Bend Strength and Fracture Initiation Energy of Sodium Borosilicate Glass Matrices Containing Dispersed Alumina Particles

<table>
<thead>
<tr>
<th>Alumina particle size</th>
<th>Volume fraction of alumina</th>
<th>Elastic modulus (psi x 10^{-6})</th>
<th>Uniaxial bend strength (^a) (psi)</th>
<th>Fracture energy (^b) (ergs/cm(^2))</th>
</tr>
</thead>
<tbody>
<tr>
<td>Glass</td>
<td>-</td>
<td>11.7</td>
<td>14,400 ± 10.8%</td>
<td>6,400 ± 15%</td>
</tr>
<tr>
<td>10-20(\mu)m</td>
<td>0.244</td>
<td>16.5</td>
<td>18,500 ± 9.6%</td>
<td>13,800 ± 16.5%</td>
</tr>
<tr>
<td></td>
<td>0.300</td>
<td>17.8</td>
<td>20,800 ± 7.6%</td>
<td>17,100 ± 12%</td>
</tr>
<tr>
<td></td>
<td>0.326</td>
<td>18.8</td>
<td>20,700 ± 6.1%</td>
<td>17,000 ± 7.4%</td>
</tr>
<tr>
<td>20-30(\mu)m</td>
<td>0.350</td>
<td>19.3</td>
<td>21,500 ± 7.1%</td>
<td>16,200 ± 11.5%</td>
</tr>
<tr>
<td></td>
<td>0.400</td>
<td>21.0</td>
<td>23,450 ± 9.8%</td>
<td>19,000 ± 6.4%</td>
</tr>
<tr>
<td></td>
<td>0.446</td>
<td>22.8</td>
<td>24,300 ± 5.4%</td>
<td>20,200 ± 7.2%</td>
</tr>
<tr>
<td>30-37(\mu)m</td>
<td>0.420</td>
<td>21.6</td>
<td>23,300 ± 6.3%</td>
<td>21,000 ± 18%</td>
</tr>
<tr>
<td></td>
<td>0.470</td>
<td>23.3</td>
<td>23,000 ± 8.7%</td>
<td>18,600 ± 13.9%</td>
</tr>
</tbody>
</table>

\(^a\) - Average of 6 to 16 specimens.
\(^b\) - Average of 8 to 14 specimens.
Fig. 4. Experimental data for uniaxial bend strength as a function of volume fraction of alumina.
Fig. 5. Experimental data for fracture initiation energy plotted against volume fraction of alumina.
Fig. 6. Typical microstructures of glass containing
(a,c) 24.4 v/o of 10-20 μm, and
(b,d) 42.0 v/o of 30-37 μm alumina.
Figure 6a and 6b shows the 24.4 v/o of the 10-20 μm alumina and 42 v/o of 30-37 μm alumina in glass composites where the particle sizes vary over a wide range. The tendency of alumina particles to agglomerate is shown in Fig. 6c and 6d; this effect makes the system more non-homogeneous. The Eq. (2) derived by Gilman\textsuperscript{19} for single crystals was based on the isotropic elasticity theory and the $\gamma$ was calculated by measuring the forces when the cracks at the cleavage began to increase in length, i.e. the critical forces for the crack propagation of a slow moving crack. Wiederhorn used the DCB test for isotropic, homogeneous glass specimen and calculated the $\gamma$ from Eq. (3). But for the glass with second phase alumina particles dispersed in it, the system is non-homogeneous as shown in microstructures and moreover, the failure of the DCB specimens was catastrophic, so how is it possible to use Eq. (3) to calculate the $\gamma$ for the composites?

Again in specimens like glass or glass-alumina composites with machined slots or notches, numerous microflaws are left along the root of the notch. Fracture can initiate at a number of these microcracks simultaneously. The double-cantilever test specimen used by Lange\textsuperscript{23} consisted of a crack introduced by slotting the specimen with a thin diamond blade followed by extending a crack by wedging the slot open. The length $L$ in Eq. (3) was taken from the end of the wedged crack. The specimens used in the present work consisted of a crack introduced by a thin diamond blade alone. The $\gamma$ values obtained were almost identical to those reported by Lange.\textsuperscript{23} The microflaws present at the crack tip for the above two specimens also act as stress concentrators. The load necessary to break these specimens is determined by these microflaws.
Again, from the scanning electron micrograph of the fractured specimen, Fig. 7(a,b,c,d and e) it was found that the crack always stays in the glass matrix. The load necessary to fracture the specimen appears to depend on the microflaws formed by sawing, so the increase in $\gamma$ values for the glass-alumina composite system can be dependent on these microflaw lengths. For the 4-point bending specimens, the load was recorded at the point of fracture. At the surface of these specimens there are a number of microflaws introduced by diamond sawing. Therefore, when the applied load reaches a certain critical load at the tip of the microflaws, the fracture occurs. This leads to the appearance of the similar information from the double-cantilever test where the load necessary to fracture the specimen is controlled by the same microflaws introduced by the diamond blade at the slot tip. If we plot the bending strength against the fracture initiation energy obtained from the double-cantilever test, a straight line relationship is obtained as shown in Fig. 8. Because the dimensions of the specimens are almost identical for all the specimens, it can be interpreted that the load necessary to break the specimen in 4-point bending test is proportional to that of the double-cantilever-beam test. Therefore it is proposed that the increase in $\gamma_I$ value obtained (Fig. 5) is mainly due to the increase in load necessary to fracture the specimens. Therefore if the load at fracture is increased, then $\gamma_I$ will be increased from Eq. (3).

For this composite system, the increase in strength of the glass matrix by the alumina particles can be easily explained by the proposed fracture theory of Hasselman and Fulrath. The dispersed alumina particles within the matrix limit the flaw size and strengthen the composite.
Fig. 7. Scanning electron micrographs of the fracture surface of glass matrix containing:
(a) 0; (b) 24.4 v/o of 10-20 μm; (c) 35.0 v/o of 20-30 μm;
(d) 40.0 v/o 20-30 μm; and (e) 42.0 v/o of 30-37 μm alumina.
Fig. 8. Plot of uniaxial bend strength against fracture initiation energy of the glass-alumina composites.
The present work was done with the specimens in region II of Fig. 1. The constant mean free path should give the constant strength but the strength of the composite so obtained shows some variation. This can be explained as follows:

(1) The mean path was calculated on the basis of the average alumina particle size, but the Eq. (4) is based on particles of radius R. With standard deviations, the strength value so obtained is close to the identical strength for the constant mean free path composites.

(2) The tendency of alumina particles to agglomerate affects the strength of this kind of composite system.

This agglomeration tendency of alumina particles has been avoided by improving some of the processing techniques during fabrication of the composites. It was found that the strength was approximately 15% higher for a composite without the agglomeration of alumina.

Therefore, uniaxial bend strength of this glass-alumina composite system depends on the second phase dispersion. Strengthening of the glass by the alumina can be explained by the limitation of the flaw size present in glass by the alumina particles. The application of DCB method for calculating \( \gamma_I \) for this kind of non-homogeneous system is still in question.
IV. SUMMARY

The uniaxial bend test for the measurement of strength and the double-cantilever-beam test designed for the measurement of the fracture initiation energy $\gamma_I$ of a glass-alumina composite system show that the alumina second phase dispersion increases the strength of the composites and increases the load necessary for failure in the DCB test. This load when used to calculate the $\gamma_I$ from Eq. (3), gives increased $\gamma_I$ with an increase in the volume fraction of alumina. The scanning electron micrograph of the fractured specimens from the DCB tests show that the crack always remains in the glass matrix. The presence of microflaws around the crack tip of the DCB specimens introduced by the diamond blade as observed by the scanning electron microscope leads to similar information about the load necessary to break the specimen to that of 4-point bending test. The tendency of alumina particles to agglomerate reduces the strength of the composite.
ACKNOWLEDGMENT

I wish to express my sincere thanks to my teacher and research advisor, Richard M. Fulrath, for his continuous guidance in this work. Thanks are extended to Prof. Joseph A. Pask and Prof. Iain Finnie for their helpful comments and suggestions. The technical assistance of George Georgakopoulos, Lee Johnson, Walt Toutolmin, Doug Kreitz and Kelly Radmilovic was greatly appreciated. Discussions with Leo Froschauer, Gautam Bandyopadhyay, David Wang, John Sherohman, Truett Sweeting and other friends were useful and also acknowledged.

Finally, I want to especially thank my parents Mr. and Mrs. Debendra Nath Biswas and to my whole family for their continuous encouragement at all times.

This work was done under the auspices of the United States Atomic Energy Commission.
REFERENCES

This report was prepared as an account of work sponsored by the United States Government. Neither the United States nor the United States Atomic Energy Commission, nor any of their employees, nor any of their contractors, subcontractors, or their employees, makes any warranty, express or implied, or assumes any legal liability or responsibility for the accuracy, completeness or usefulness of any information, apparatus, product or process disclosed, or represents that its use would not infringe privately owned rights.