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EFFECT OF INTERFACIAL BONDING ON STRENGTH OF A MODEL TWO-PHASE SYSTEM

Berkeley, California

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Donald O. Nason
(M. S. Thesis)

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Lawrence Radiation Laboratory
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ABSTRACT

Model systems consisting of tungsten spheres in a dense glassy matrix were fabricated by vacuum hot-pressing. Metal-glass interfacial adherence was achieved while development of new crystalline phases in the glass was avoided.

Interpretation rested on the existence of a model surface on which the introduction of flaws could be controlled. A new criterion based on the limitation of flaw size is advanced to explain the effect of second-phase inclusions on strength in brittle systems. The roles of surface treatment and internal stresses for bonded and nonbonded inclusions are outlined in the context of this hypothesis.

INTRODUCTION

In recent years, increased attention has been directed toward understanding the strength of brittle two-phase systems. Many variables must be controlled if an understanding of the mechanism of fracture is to be gained. Among the critically important parameters are the type and extent of porosity; the mechanical properties of each phase; the size, shape, and distribution of the minor phase; the character of internal stresses; and chemical bonding or interaction between the two phases.

The study of models has provided much fruitful information. Models are invaluable because they allow microstructure to be closely controlled. Many complicating variables that interplay in conventional systems can be eliminated so that basic parameters can be evaluated singly. In each of the following investigations, crystalline particles are dispersed in a matrix of glass in an attempt to approach model conditions.

Fulrath has demonstrated the existence of internal stresses in alumina-glass compacted composite systems.[1] The magnitude of induced internal stress has been correlated with the level of interphase chemical reaction and the difference in thermal expansion of the materials. Grossman applied the same techniques to thoria-glass compacts.[2] Studt explained strength trends in mullite-glass compacts on the basis of interfacial chemical reaction.[3] Binns studied mechanical strength and elasticity in alumina-glass and zircon-glass compacts as a function of internal stresses.[4] His interpretation was based on observations of interlinking microstructural cracks surrounding the dispersed crystal particles. Some of these studies are compared in Fig. 1.

The above investigations were hampered by inherent difficulties with the model. The shape and orientation of the included phase, the configuration of the internal stress field, and the surface conditions are unknown or unspecified. Jacobson further refined the systems, substituting nonreactive nickel spheres as the dispersed crystalline phase.[5] His results are shown in Fig. 2.

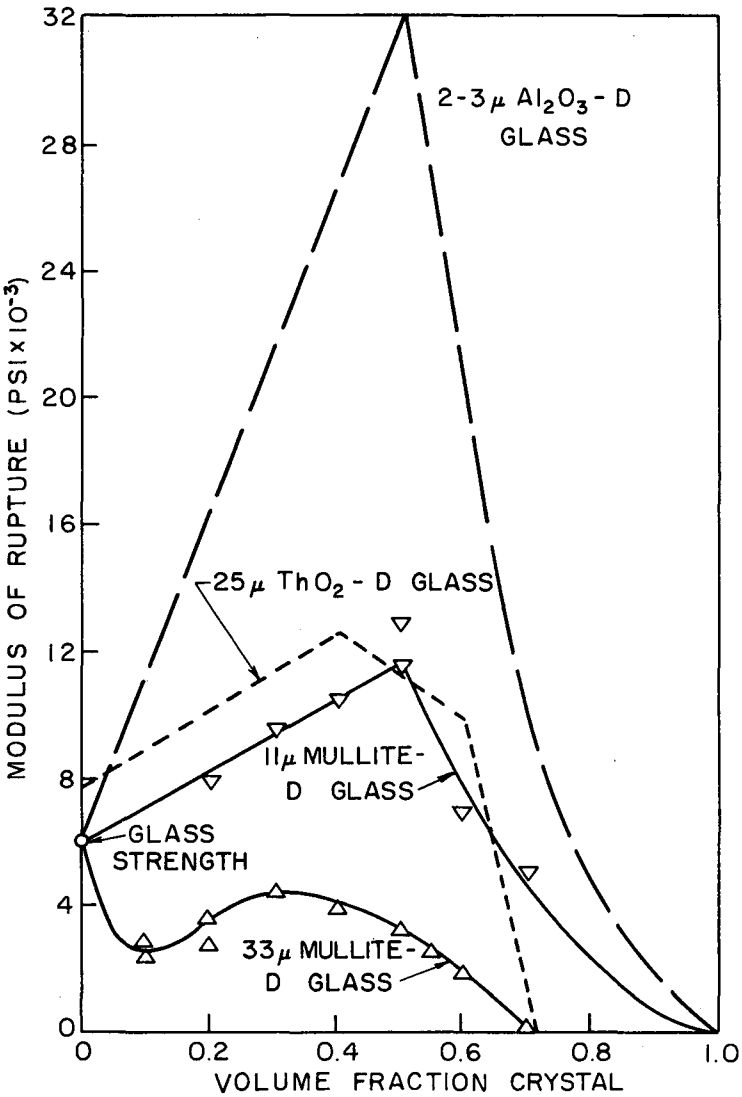

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Fig. 1. Rupture moduli of various systems of crystals dispersed in glass D as a function of volume fraction crystal.

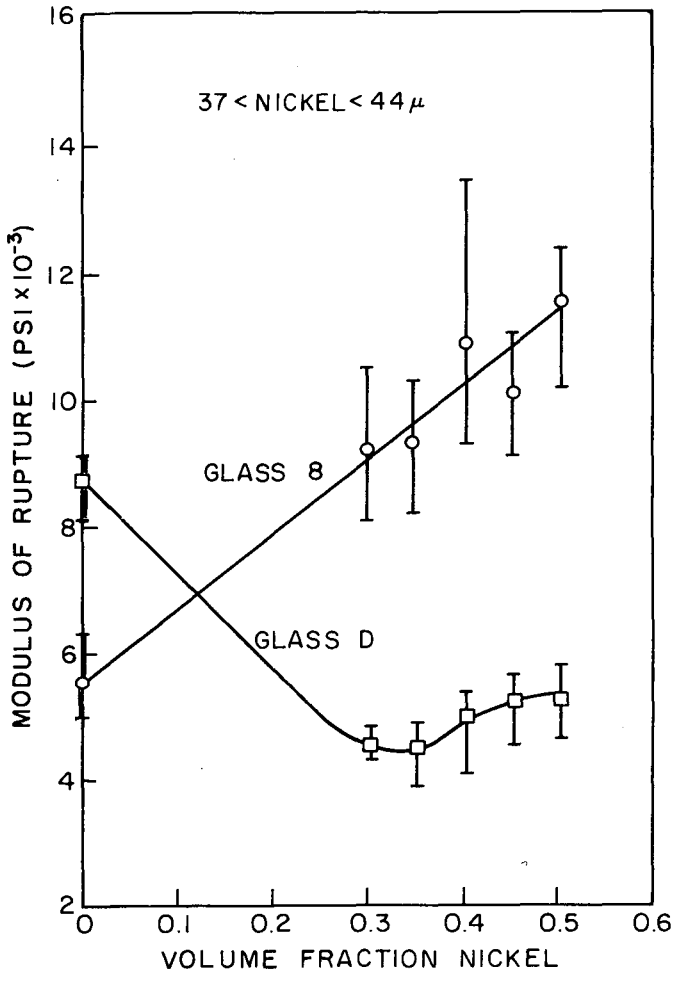

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Fig. 2. Rupture moduli of nickel-glass systems as a function of volume fraction nickel.

A primary aim of this present study was to further idealize this type of model system in order to guarantee a model surface. It was desired to develop interfacial bonding without sacrificing the simplicity of a two-phase system. A criterion was sought that would explain the dependence of strength on the amount and nature of the dispersed particles when interphase bonding is realized. A secondary aim was to investigate the role of internal stresses in determining strength in systems of both bonded and nonbonded included phases.

EXPERIMENTAL PROCEDURE

Materials Preparation

The tungsten spheres used in this study were obtained from the Linde Company.[6] The spheres were separated into size ranges by screening and air-elutriation by means of a Roller Particle Size Analyzer.[7] Subsequent microscopic examination verified the sphericity of the particles and showed the size fractions to be quite reproducible.

Glass compositions shown in Table I were selected from the soda-borosilicate system on the basis of predicted thermal expansions relative to tungsten. The probable adherence of tungsten and these glasses was anticipated from a previous study.[8] The glasses were prepared from their component oxides by air-smelting in a rotating, inclined, platinum crucible. The batches were quenched in graphite molds, and reduced to 325 mesh by dry comminution in an Al_2O_3 mortar. Post-smelted compositions were determined by analytical chemical methods.

Thermal-expansion measurements of the glasses in the as-cast state were made on a silica-rod dilatometer over the indicated temperature range. The softening point of each glass was checked independently by observing the stress-relief temperature by means of the crossed-polaroid technique.

Sample Fabrication

Specimen compositions were determined by assuming theoretical density in a compact 1/8 in. thick and 1 in. in diameter, and back-calculating the proper weight of the glass and crystal powders to assure the desired volume ratio. It was assumed that postfabrication annealing was unnecessary for samples of this size. The materials were dry-mixed manually by tumbling in vials.

Compaction was achieved by hot-pressing in a graphite die under a vacuum of 10^{-4} of mercury. Controlled fabrication schedules for heating (8° C/min) and cooling (6° C/min) were uniform from sample to sample for each type of glass. In particular, the materials were heated 100 C° above the temperature at which yielding was first noticed under

Table I. Properties of materials

Material	Composition (wt%)	Thermal expansion (in./in° C × 10^6)	Fabrication temperature (° C)	Density (g/cm^3)
N-2 glass	SiO_2: 66.1	6.10 (R. T[a] – 550° C)	700	2.37
	Na_2O: 12.7			
	B_2O_3: 21.1			
N-3 glass	SiO_2: 61.2	7.95 (R. T. – 550° C	700	2.57
	Na_2O: 18.2			
	B_2O_3: 20.5			
N-4 glass	SiO_2: 67.0	3.25 (R. T. – 450° C)	650	2.16
	Na_2O: 4.5			
	B_2O_3: 28.4			
D glass[b]	SiO_2: 68.4	7.7 (R. T. – 450° C)	650	2.47
	Na_2O: 11.6			
	B_2O_3: 14.9			
8 glass[c]	SiO_2: 50			
	B_2O_3: 37			
	Al_2O_3: 13			
Tungsten		5.6 (R. T. – 700° C)		19.3
Nickel		13.9 (R. T. – 300° C)		8.9

[a] R. T. – room temperature

[b] From Jacobson[5]

[c] Presmelted composition

the initial load of 2000 psi. They were soaked at this maximum temperature for 20 minutes under 3500 psi, and the load was maintained at 2000 psi during cooling.

Glass powders were hot-pressed adjacent to polished tungsten and nickel disks in an additional series of experiments referred to as pressure-wetting studies. Care was taken to assure that these large interfaces duplicated conditions at typical metal-glass boundaries in the compacts. A cast piece of N-4 glass was also placed on a tungsten disk and carried through the heating cycle, but without the application of pressure.

Specimen Testing

Both upper and lower planar surfaces of the fabricated glass-crystal compacts were ground successively with 240- and 600-grit SiC with a water lubricant and by a uniform technique. Certain specimens were purposely surface-treated differently, as outlined in Table II. Rectangular bars about 1/8 in. wide were fractured at room temperature by means of a four-point loading apparatus. Three or four strength measurements were made from each compact. In selecting the tension-bearing face, no effort was made to distinguish between what had been upper and lower faces of the compact during fabrication.

The interfaces in the pressure-wetting studies were tested manually for shearing strength.

Compact densities were measured with an air pycnometer and by a mercury-displacement method. Each technique provides densities accurate to about ± 2% with samples of this size.

Sample Examination

Representative microstructures of the load-bearing surfaces were photographed without further preparation. Replicas were made of the nonplanar fracture faces and gold-plated to aid examination. These were flattened for direct observation of surface detail with the metallographic microscope. The metal disks in the pressure-wetting studies were also photographed after being parted from their mating glasses.

Table II. Effect of surface treatment

Procedure	Strength[a]
Normal	σ
Normal, but ground only with 120-grit SiC	0.92 σ
Normal, but ground only with 600-grit SiC	1.05 σ
Sawed but not ground	0.85 σ
Normal, but kerosene used as grinding and sawing lubricant in place of water	1.2 σ

[a] With normal spreads

EXPERIMENTAL RESULTS

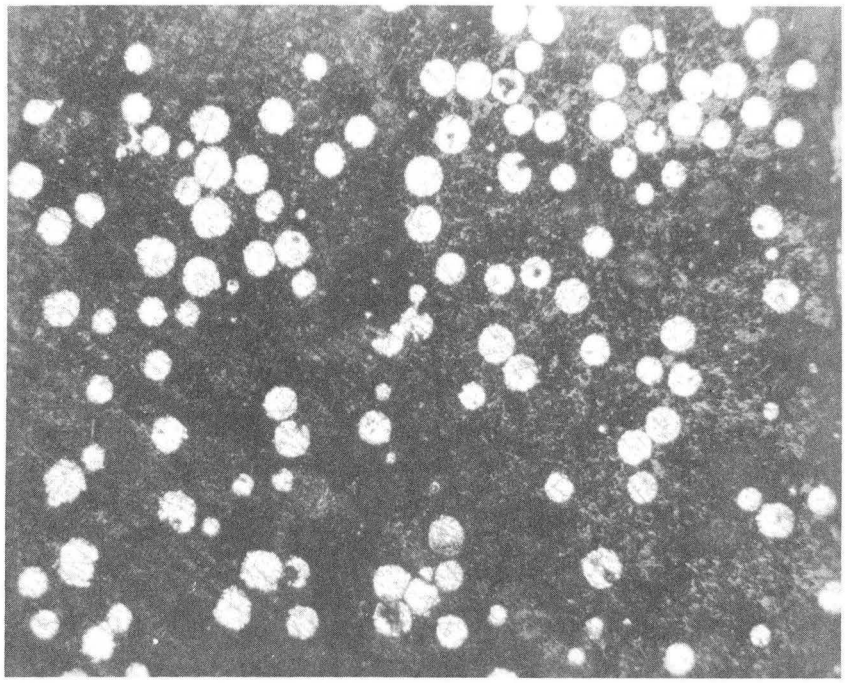
Nearly ideal density was achieved in all compacts. Deviations from the particular theoretical value clustered at random within the stated accuracy of the pycnometer. The dispersion of the included phase is illustrated in Fig. 3, a micrograph of the as-ground, load-bearing surface. From repeated examinations of similar areas, it was concluded that the dry-mixing technique was fully adequate to disperse the heavy spheres randomly in the matrix.

The variation of strength with volume fraction of included tungsten is shown in Fig. 4 for the case in which the glass had a lower thermal expansion than tungsten. The spread in any one set of points is remarkably small for brittle-fracture measurements and is quite uniform over the whole range, considering that only three or four specimens were broken at each composition. For the same basic system, reversal of the relative thermal expansion differences yielded the data in Fig. 5. The implied slopes in Fig. 4 and 5 are quite similar for equivalent particle-size ranges. Although the data are reliable and reproducible, it is felt that analytical representation at this stage is premature. X-ray diffraction patterns from these compositions showed no evidence of any new phases.

The nickel-glass-D compositions studied by Jacobson (see Fig. 2) were reexamined by means of the experimental techniques of this study. Figure 6 shows that the results nearly duplicate the earlier work.

The effect of variation of particle size at constant composition (10% tungsten, 90% glass by volume) is displayed in Fig. 7. The upper graph is included to indicate the spread in values; the lower is a log-log plot. There appears to be an abrupt change in the behavior at approximately 60 μ. At larger sizes, strength seems to be independent of crystal size for this volume fraction. The plotted abscissas are median values for each particle-size range.

The interesting result of various prefracture-surface treatments is shown in Table II. There is surprisingly little correlation between the severity of surface abrasion and strength. As-sawed surfaces bear an applied load nearly as well as those treated in the normal way.



ZN-4004

Fig. 3. A typical microstructure of a tungsten-glass compact (30 vol % tungsten, $63 < d < 74$ μ; 70 vol % N-2 glass) in the normal as - ground condition preceding fracture. 106×.

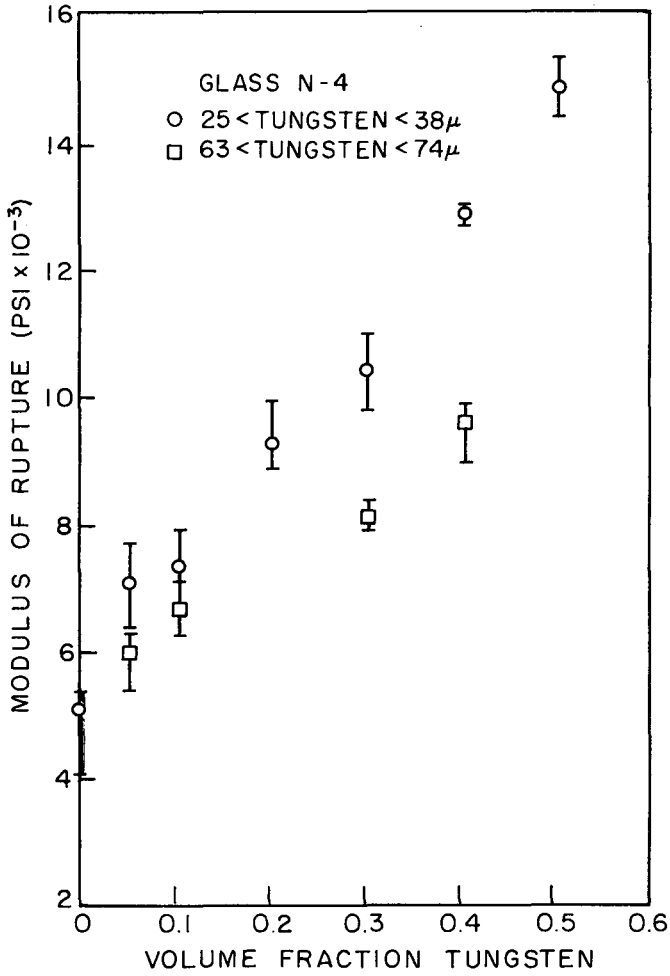


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Fig. 4. Rupture moduli of tungsten–glass N-4 systems as a function of volume fraction tungsten for two particle-size intervals.

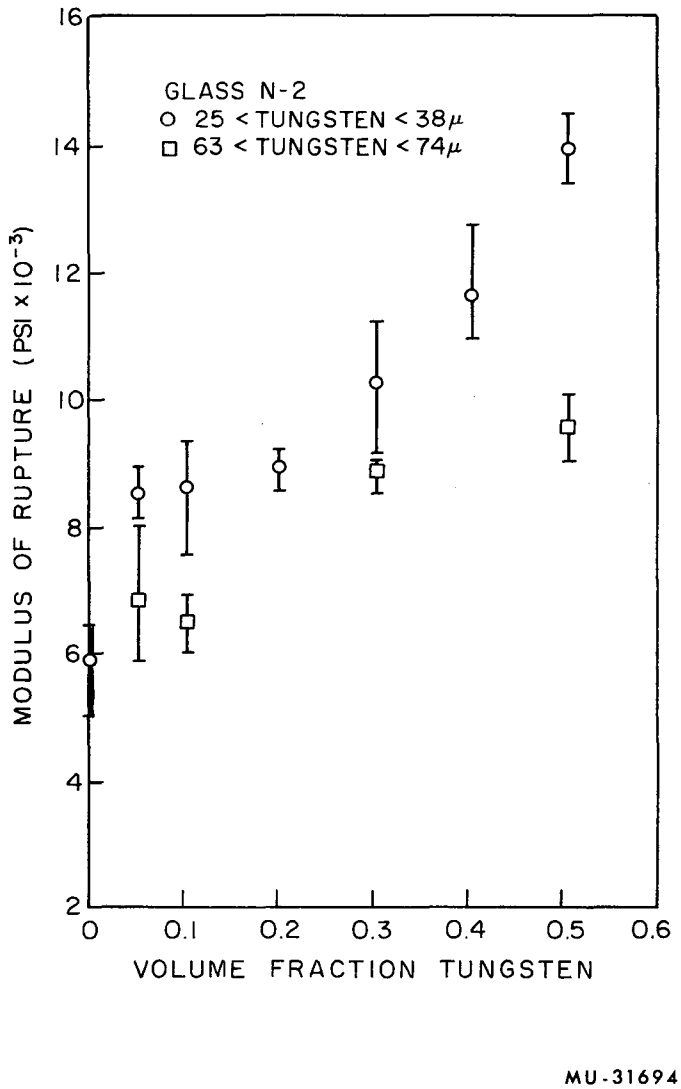

Fig. 5. Rupture moduli of tungsten-glass N-2 systems as a function of volume fraction tungsten for two particle-size intervals.

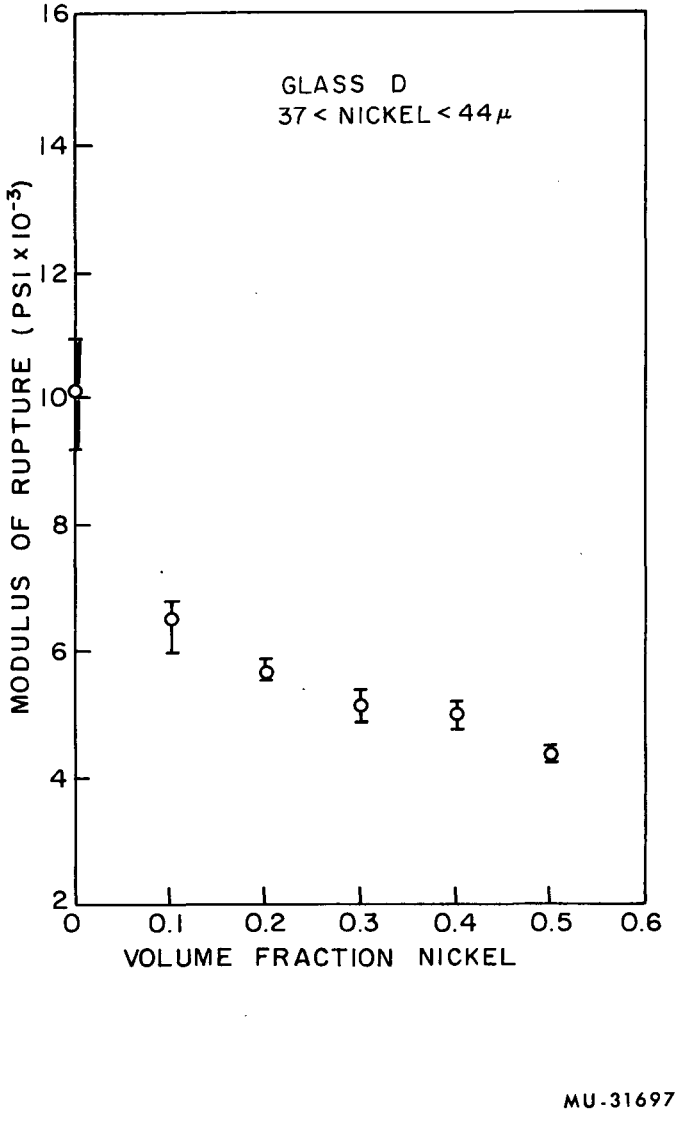

Fig. 6. Rupture moduli of nickel-glass-D systems as a function of volume fraction nickel.

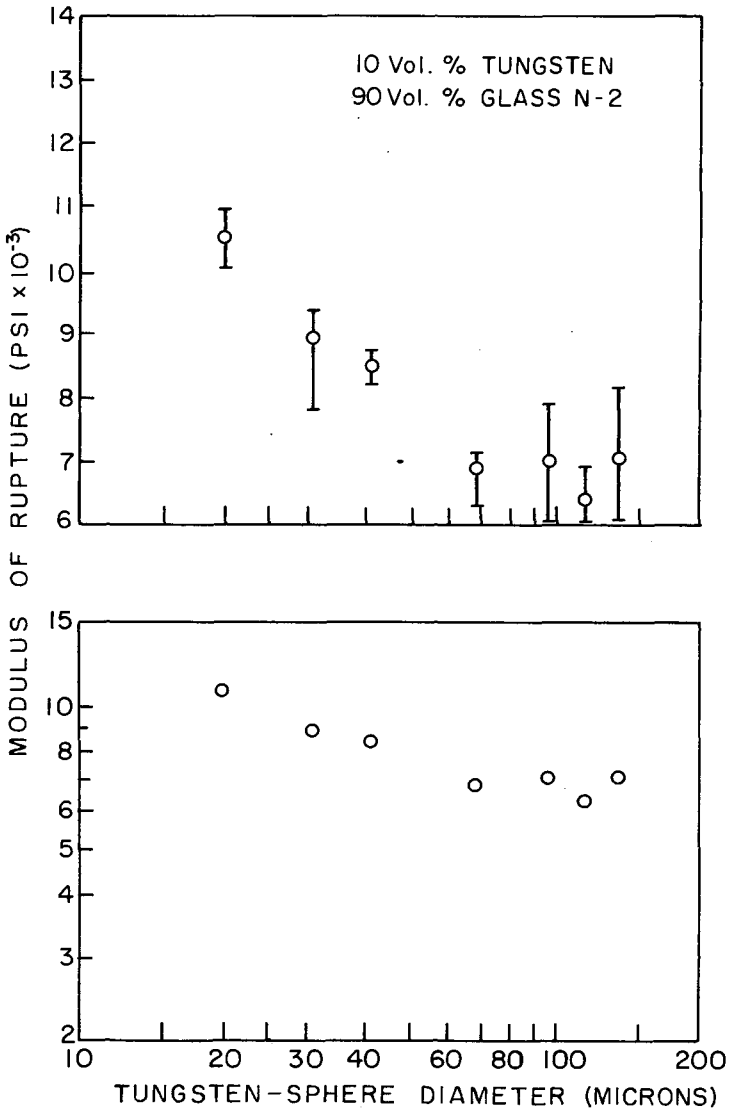

Fig. 7. Rupture moduli of tungsten-glass N-2 systems as a function of tungsten-particle size, plotted both linearly (top) and logarithmically.

The compositions of Fig. 8 showed evidence of unwanted chemical reaction during hot-pressing. X-ray and microscopic analysis revealed devitrification products in the glass, visible in a polished section as isolated grains in the matrix. No evidence of tungsten chemical participation in the devitrification could be found.

The pressure-wetting experiments supplied key information. Macropictures of the sheared interfaces are shown in Fig. 9. Both glass D and glass 8, which span the value of the thermal expansion of nickel, sheared from it very easily. No roughening of the polished disk had occurred; only a slight clouding was visible. Tungsten and glass N-4 were mutually adherent, resisting all manual attempts to break the interface apart. No quantitative measure of the ultimate cleavage force was made because of the experimental difficulty of applying a shearing load without an accompanying bending moment at the interface. When the disks were eventually broken by machine, fracture occured not at the bonded interface, but in the adjacent glass (see Fig. 9). It is clear that tungsten bonded to a soda-borosilicate glass whereas nickel did not. The piece of glass N-4, heated with tungsten according to the appropriate schedule but with no applied pressure, failed to adhere. Pressure-induced flow of the glass seems necessary for bond formation at these intermediate temperatures.

The experimental procedures used in this study yield a reasonable approximation to a model surface. Figure 3 shows that a certain unavoidable number of sphere pull-outs occurred, as might be expected. Observations of polished sections and the achievement of ideal density verify that open porosity was absent; the glass surface was planar and had been abraded uniformly.

Photographs of replicas of the surface of the fracture are shown in Fig. 10. In the nonbonding system of nickel and glass D, the fracture path intersects many particle-containing cavities. The spheres themselves were not fractured. In the tungsten compacts, the path seemed to avoid the particles, bypassing them so that they appear obscure in the replica. This would be normal for well-bonded particles.

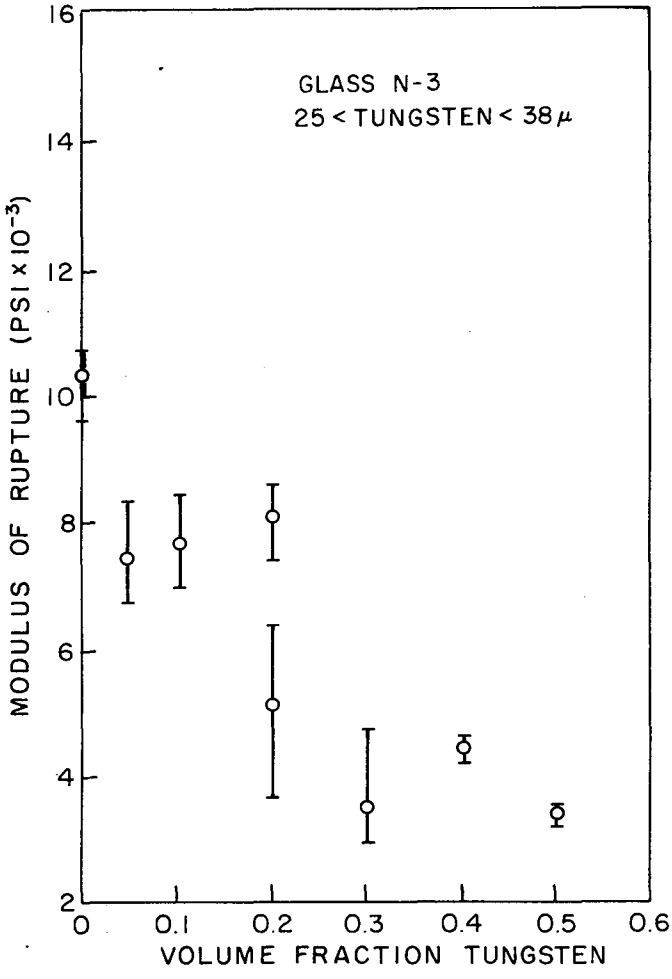

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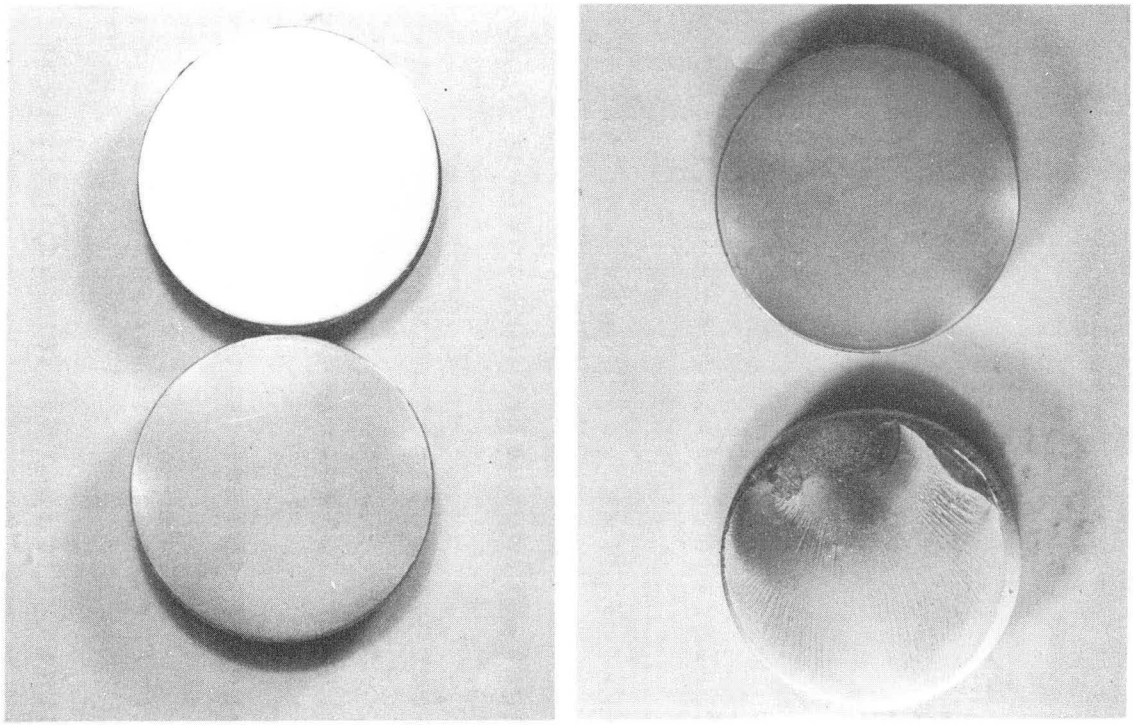
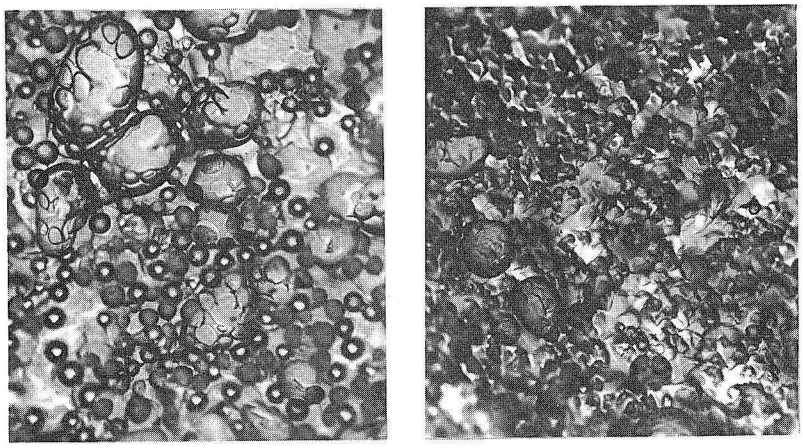
Fig. 8. Rupture moduli of tungsten–glass N-3 systems as a function of tungsten particle size.



(a) (b)

ZN-4003

Fig. 9. Photomacrographs from the pressure-wetting studies of the metal disks at the parted interfaces. 1.9×. (a) nickel disk before (top) and after experiment; (b) tungsten disk before (top) and after experiment.



(a) (b)

ZN-4002

Fig. 10. Photomicrographs of replicas of fracture faces, using direct light at 70×.
(a) 40 vol % nickel, $37 < d < 44\mu$; 60 vol % glass D
(b) 40 vol % tungsten, $25 < d < 38\mu$; 60 vol % glass N-4.

DISCUSSION AND INTERPRETATION

According to the well-known criterion introduced by Griffith, preexistent flaws in a brittle material propagate under a critical stress.[9] When the load is maintained, failure ensues. Others have modified the theory to account for stress concentrations, plastic deformation, and other significant factors.[10] The normal scatter in strength values is due largely to the random size and orientation of these flaws. According to Shand, there is no reason to believe flaws must inherently occur internally in glass.[11] They are introduced accidentally and erratically. In order to improve confidence in strength values for brittle materials, either large numbers of identical measurements must be made and statistical reliability invoked, or deliberate steps must be taken to control the flaw character. Fisher and Holloman[12] and Epstein[13] have developed expressions for strength distributions from a statistical approach. Such analytical formulations always require the assumption of a particular flaw distribution.

In this study, systems have been fabricated that consist of two-phase compositions of spherical inclusions in a pore-free brittle matrix. As the samples are tested in bending, attention can be focused on the outermost volume of the tension-bearing surface where fracture initiates. Microexamination shows that the fracture path remained in the glass and did not pass through spheres. In beams loaded in flexure, the tensile stress is greatest at the surface and decreases linearly with depth. At typical breaking loads in this study, the stress gradient was 4 to 10 psi per micron of depth below the surface.

In an attempt to circumvent the usual uncertainty about flaw distribution, the samples were subjected deliberately to controlled surface abrasion. Calculations based on Griffith's formula[9] indicate that flaws on the order of 100 μ in length are responsible for failure in glass. Edge flaws can be somewhat smaller but be equally effective. The treatment in this study was designed to be severe enough to inflict flaws of this size uniformly over the surface.

This approach leads to a general explanation of the strengthening effect of scattered inclusions in a brittle matrix. According to Hasselman, a likely rationale is that the particle spacing limits the size of flaws that can be introduced in the glass surface.[14] Comparison can be made between the probable flaw size just mentioned and typical interparticle distances shown in Figs. 3 and 10.

The picture is then one of a model two-phase surface of sectioned spheres in a matrix of glass that has been saturated with flaws that are restricted in size by the dispersed particles.

Returning to the experimental data, this idea qualitatively explains the strengthening effect of included spheres of tungsten in glasses N-2 and N-4, as exhibited in Figs. 4 and 5. Interspersing crystals in the matrix decreases the average unobstructed glass path. Of course, because of random placement of particles, the largest available path through the glass is larger than the average interparticle spacing. A small spread in strengths at any composition reflects proper mixing and effective averaging of these distances. The effect should also be somewhat proportional to the number of spheres. This is in agreement with Fig. 7. Below 60 μ the familiar slope of -1/2 is seen in the log-log plot. At larger sizes, strength appears to be independent of particle size and therefore of interparticle distance.

The role of internal stresses must be clarified. The importance of this factor in model systems with a nonbonded crystalline phase has been demonstrated by Jacobson (Fig. 2).[5] With tungsten spheres, stress fields surrounding the bonded particles of several thousand psi may be calculated from the equation of Selsing[15] and the proper elastic constants. However, it is believed that internal stresses played only a minor role in the tungsten-glass systems. Comparison of Fig. 4 with Fig. 5 shows that reversal of the type of induced stress had little effect on the strength.

It is recognized that a propagating flaw will travel unchecked through a brittle volume if the critical load is maintained. As mentioned previously, there is a stress gradient in the tungsten-glass systems of 4 to 10 psi/μ. This gradient is small enough that the fracture path may easily bypass surface obstructions by traveling beneath them. Thus, it

is felt that the major influence of thermal expansion differences in these systems is not in any impedance offered by local regions of internal stress.

The principal distinction between tungsten and nickel systems has been made previously. Spheres of nickel dispersed in glass D constitute a model system physically similar to tungsten spheres in N-4 glass. Comparison of Fig. 4 with Fig. 5 shows the drastically different dependence of strength on included-crystal content in the two cases. The major difference is a chemical one. Tungsten is well bonded to its matrix glass envelope, whereas nickel lacks adherence altogether. The contrast in bond development is demonstrated in Fig. 9. That the tungsten-glass interface may even be stronger than the glass itself is made plausible by the replica micrographs in Fig. 10.

The contrast in behavior of bonded and nonbonded systems may be interpreted by the restricted-flaw idea in the following way. (A reasonable hypothesis for the role of internal stresses becomes clear also.) Bonded tungsten inclusions act to limit the length and depth of flaws existing on the surface because there is adherence at the glass-crystal contact. If bonding is maintained, the presence of differing internal stresses is insignificant by this hypothesis. On the other hand, a flaw intersected by a nickel-glass-D interface is thereby extended by at least the diameter of the cavity, as the spheres are unbonded. Relative thermal contraction of the nickel has created an ill-defined surface of sectioned spherical voids and holes containing loose particles. Model two-phase surface conditions are no longer even approximated. In the nickel-glass-8, the sign of the expansion difference is reversed. The sphere is clamped in place mechanically by the contraction of the matrix. Though chemical bonding is lacking, there is no gap at the interface. Internal stresses have maintained a model surface, and strengthening should occur according to the restricted-flaw postulate.

The slope of nickel-glass-8 in Fig. 2 differs by a factor of two from the slopes of the tungsten-glass systems. Though the average sphere diameters of the two metals are different, Fig. 7 shows that

the size effect alone cannot be responsible. Another explanation depends on the relative load-bearing capacities of chemically and mechanically bonded interfaces.

When inclusions with a higher elastic modulus are introduced into a medium, the stiffness of the composite is increased, providing that a stress-bearing interface has developed between the two phases. The greater share of any imposed load is borne by the stiffer phase. Tungsten and a soda-borosilicate glass form such a system. The effective transfer of stress from glass to metal allows failure of the glass to be postponed until external loads are higher. Nickel-glass-8 interfaces were not bonded and may not shift an applied load from the glass as effectively as does tungsten. This would be reflected in reduced stiffness of the system and failure under smaller loads.

The actual distribution in the matrix of coexisting internal and applied stresses is complex. The elastic properties as well as the chemical properties of tungsten and nickel are different. These other factors may also influence the slopes under discussion.

A further test of the restricted-flaw idea was made by varying the severity of surface abrasion. The effect on fracture strength is shown in Table II. The mildest form of treatment (grinding with 600-grit SiC) was judged to be harsh enough to produce flaws of the maximum size possible at this crystal-glass composition. All coarser abrasion was intended to test the postulate that the spacing between the exposed particles was limiting the length of flaw inflictable at the surface. In support of this idea, strengths were not seriously affected by the amount of overabrasion. This is further demonstration of the applicability of the restricted-flaw postulate as the interpretative tool for the data of this study.

SUMMARY

In model two-phase systems, strengths were found to depend in a predictable way on the volume fraction and particle size of the included phase. Pressure-wetting studies showed that tungsten bonds to soda-borosilicate glasses, but nickel does not. In tungsten-glass compacts, reversal of the sign of the thermal expansion difference between the phases had little effect. Both increasing the sphere content and decreasing the sphere size enhance strength.

With nonbonding nickel spheres, strengths increased with crystal content only when the induced internal stresses maintained a mechanical bond at the glass-metal interface. When expansion coefficients were such that contraction of the sphere relative to the matrix created a gap at the boundary, strengths were sharply reduced.

The restricted-flaw postulate was advanced to explain the strengthening effect of introducing inclusions into a brittle matrix. Dispersed bonded particles in a model surface act to limit the size of flaw inflictable in the surrounding glass, i. e., interparticle spacing controls strength. Internal stresses are important in their effect on maintaining or disrupting metal-glass contact for nonbonded particles.

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REFERENCES

1. R. M. Fulrath, Internal Stresses in Model Ceramic Systems, J. Am. Ceram. Soc. 42, 423-29 (1959).
2. L. N. Grossman, Fracture in Model Ceramic Systems (M. S. Thesis), University of California, Berkeley, 1961.
3. P. Studt and R. M. Fulrath, Mechanical Properties and Chemical Reactivity in Mullite-Glass Systems, J. Am. Ceram. Soc. 45, 182-88 (1962).
4. D. B. Binns, Some Physical Properties of Two-Phase Crystal-Glass Solids, in Science of Ceramics, edited by G. H. Stewart, (Academic Press, Inc., New York, 1962), vol. 1, p. 315-334.
5. L. A. Jacobson and R. M. Fulrath, Strength of a Two-Phase Model System (M. S. Thesis), University of California, Berkeley, 1962.
6. Linde Co., Crystal Products Division, Bulletin F-1402.
7. American Instrument Co., Inc., Catalog No. 5-445, Instruction No. 779-A.
8. R. M. Fulrath, S. P. Mitoff, and J. A. Pask, Fundamentals of of Glass-to-Metal Bonding: III. Temperature and Pressure Dependence of Wettability of Metals by Glass, J. Am. Ceram. Soc. 40, 269-74 (1957).
9. A. A. Griffith, Phil. Trans. Roy. Soc. London A221, 163 (1920-1921).
10. Review by T. L. Johnston, Fracture Mechanisms in Crystalline Ceramics, in Mechanical Behaviour of Crystalline Solids: Natl. Bur. Std. (U. S.) Monograph 59 (1963), pp. 63-74.
11. E. B. Shand, Experimental Study of Fracture of Glass: I. The Fracture Process, J. Am. Ceram. Soc. 37, 52-56 (1954).
12. J. C. Fisher and J. H. Holloman, Metals Technol. 14, (1947).
13. B. Epstein, "Statistical Aspects of Fracture Problems," J. Appl. Phys. 19, 140-147 (1948).
14. D. P. H. Hasselman, Lawrence Radiation Laboratory (private communication).
15. J. Selsing, "Internal Stresses in Ceramics," J. Am. Ceram. Soc. 44, 419 (1961).

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