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Some tensile properties of metal–metallic glass laminates

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The tensile properties of brass (Cu–30% Zn)–nickel base metallic glass (MBF-35 Metglas) laminates have been investigated. Laminates were prepared by soldering these constituents together with a Pb–Sn alloy. The metallic glass exhibited an enhanced tensile ductility in the metal matrix environments, and such enhanced ductility depended on the metal matrix strength and elongation. Multiple shear bands have been observed in the metallic glass ribbon with enhanced tensile ductility. The failure modes of the laminates have been analysed based on stress–strain concentration concepts, following failure of the glass. The results were consistent with the experimental observations.

1. Introduction

Tensile ductilities of metallic glasses are limited. This is a result of the highly localized plastic flow, manifested by shear banding, in metallic glasses. Under conditions of uniaxial tensile loading the formation and propagation of one shear band is a precursor to sample failure. Even though the localized shear strain in the deformation band is extensive [1] macroscopic strain is limited as a result of the small volume of the material experiencing banding. Multiple shear banding in metallic glasses can be effected by subjecting the material to other stress states. For example, extensive banding is observed in metallic glasses subjected to rolling deformation [2].

On the basis of the extensive localized deformation which precedes tensile failure, metallic glasses can be considered "tough" engineering solids. On the basis of the low tensile ductilities which characterize them, they can be considered "brittle" from a macroscopic viewpoint. Clearly, if extensive banding precedes fracture, the material can be considered fracture tough from both points of view.

In this paper we report on the enhancement of tensile ductility in metallic glasses when they are used as a constituent in laminated metallic composites. The original idea behind this use of them was based on the phenomenological similarity of this type of composite to composites comprised of a ductile matrix in which brittle reinforcing fibres are embedded. In the latter, failure of an individual fibre need not be concurrent with sample failure. Instead the brittle constituent can be reloaded along most of its length, provided the fibre ineffective length is small in comparison to the total fibre length. We supposed that a shear band in a tensile loaded composite could be considered somewhat similar to a fibre fracture in a composite of the kind just described. On this basis, the metallic glass could be reloaded over much of its length, and subsequent additional shear bands could be formed provided the analogous ineffective length in the metal–metallic glass composite was not too large. While this presumed scenario is not followed, we have nonetheless found some interesting tensile behaviour in metal–metallic glass laminates, and these are described in this paper.

2. Experimental procedures

Brass (Cu–30% Zn) was used as the metallic constituent of the composites we investigated. This is because the mechanical properties of brass can be varied substantially. The starting material in our case was 0.8 mm thickness sheet which was cold rolled to 0.2 mm in thickness. This sheet was then annealed for 1 h at several temperatures between 473 and 773 K, resulting in various strength–ductility combinations. MBF-3 metglas, supplied by the Allied Corporation in the form of ribbon having a thickness of 0.038 mm, was the composite glass constituent. This is a nickel base (7.31% Si, 2.17% B, balance nickel) alloy which exhibits relative higher embrittlement temperatures than iron base amorphous alloys.

We have tried a number of materials, including several organic based "glues", as a binder phase for the metal and the metallic glass. However, best adhesion was achieved by using a Sn–40% Pb alloy for this purpose. The brass and the metallic glass were first soldered together, in the form of three-layer (metal–glass–metal) and five layer (metal–glass–metal–glass–metal) laminates. These were then pressed at 25 MN m^{-2} at a temperature of 473 K for 5 to 10 min (the melting point of the solder is 456 K). The "hot" pressing resulted in thin and relatively uniform solder thicknesses (about 10 to 20 μm).

Tensile tests were carried out on an Instron machine on samples having dimensions of 8 mm in width and 50 mm in length. The strain rate was 0.0025 sec^{-1}. Strain gauges EA-06-250-120 supplied by Measurements Group, Inc. were used to measure the strain to ultimate loading. (As often as not final fracture took place outside of the region over which strain was

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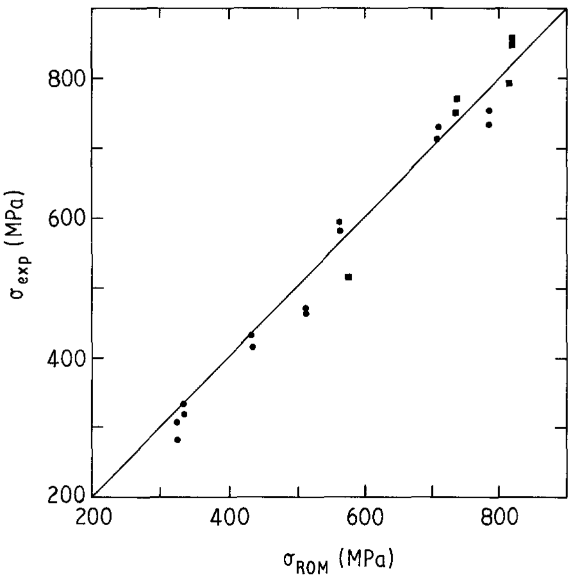

Figure 1 Comparison of laminate tensile strengths with the strengths predicted from the rule-of-mixtures. (■) five layer, ● three layer, $\sigma_g^f = 1820\,\text{MPa}$, $\sigma_{ROM} = V_g\sigma_g^f + V_m\sigma_m^c$).

measured; thus uniform strain is the "ductility" parameter used in this study.)

3. Results and discussion

3.1. Composite tensile strength

As indicated in Fig. 1, composite tensile strengths follow the rule-of-mixtures as given by Equation 1

$$\sigma_{ROM} = V_g\,\sigma_g^f + V_m\,\sigma_m^c \qquad (1)$$

In Equation 1 σ_g^f is the glass tensile strength and σ_m^c is the matrix stress at the composite strain corresponding to ultimate loading. The solder contribution to composite strength was neglected as its strength is low (about $52\,\text{MN}\,\text{m}^{-2}$) and thickness small; thus the solder volume fraction in the composite was low. While the results of Fig. 1 are unremarkable, certain other facets of the tensile behaviour of the composite were of interest. They are discussed in the following sections.

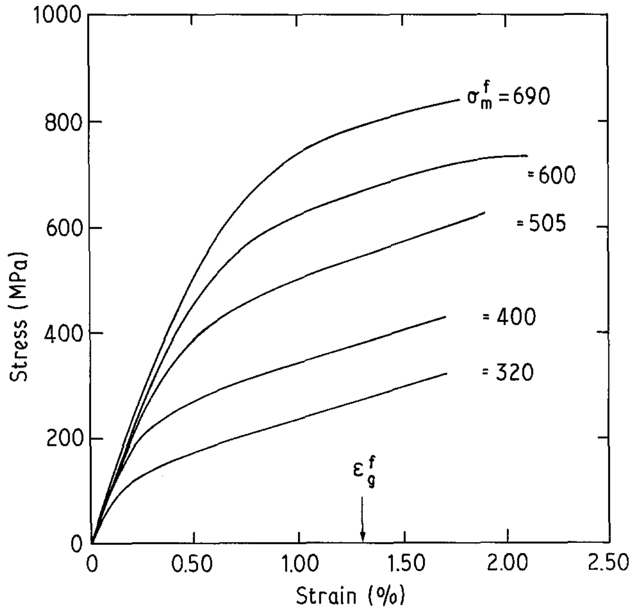

Figure 2 Tensile stress–strain curves of laminated composites up to the point of ultimate loading. The tensile fracture strain of the glass (ε_g^f) is noted in the figure; laminate strain at ultimate loading exceeds this strain.

3.2. Enhanced ductility

Fig. 2 represents the stress–strain behaviour of three layer composites up to the point of ultimate loading, which corresponds to fracture of the metallic glass. The brass tensile strengths (σ_m^f), which varied from 320 to $690\,\text{MN}\,\text{m}^{-2}$ are indicated in this figure (corresponding tensile ductilities of the brass varied from about 35% to 1.6%). Also indicated on the figure is the measured tensile strain to failure (1.3%) of the metallic glass constituent. Clearly the tensile strain to failure of the glass in the composites is greater than that of the metallic glass in monolithic form, and the degree of enhanced ductility depends on the properties of the matrix.

The degree of enhanced strain is plotted against matrix tensile strength in Fig. 3a. The figure has been divided into three regions. In region I, characterized

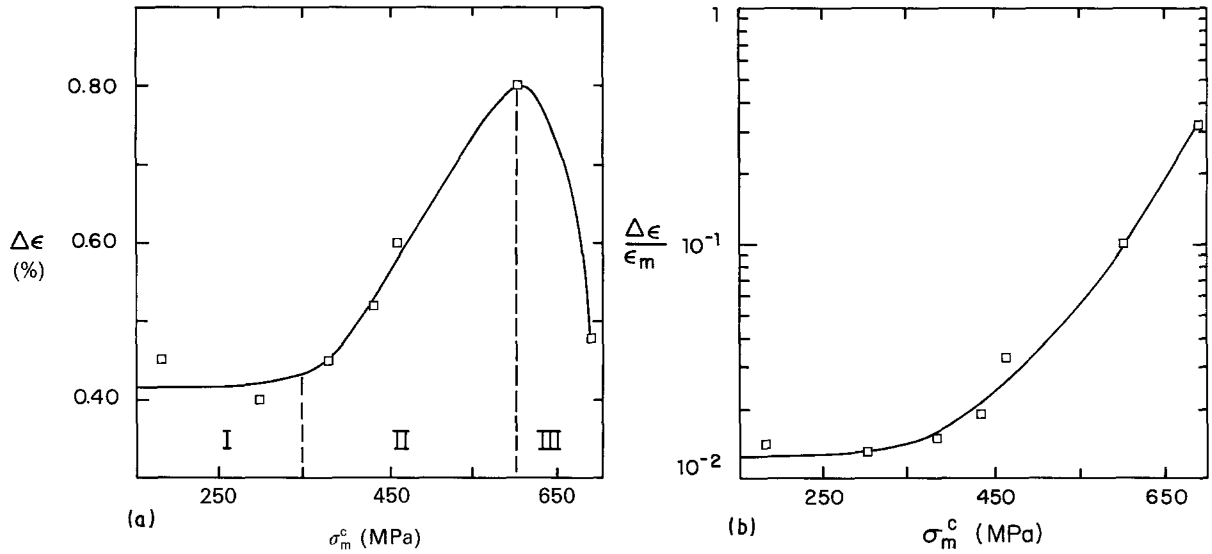

Figure 3 (a) The metallic glass enhanced tensile strain plotted against the brass flow stress at which the metallic glass fails. Maximum enhanced strains are observed at intermediate brass strengths. (b) The enhanced strain divided by the matrix uniform strain. The relative increase in strain increases monotonically with matrix strength.

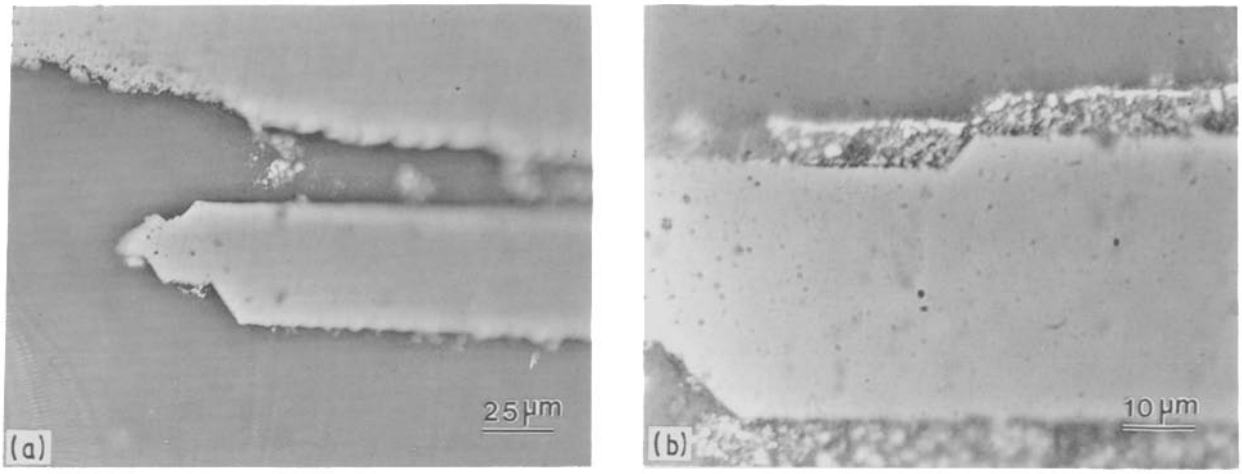

Figure 4 Optical micrographs of the metallic glass fracture regions of the composite with matrix strength $\sigma_m^f = 600$ MPa. (a) Multiple shear bands developed at the metallic glass fracture tip region. (b) Higher magnification view of the multiple shear bands at the fracture tip.

by relatively low brass strengths, the enhanced strain is relatively small and approximately independent of the strength of the brass. In region II, where the brass strength is somewhat greater, the enhanced strain increases rapidly, and in almost a linear manner, with the matrix strength. Finally, at the highest of the brass strength levels investigated, the enhanced strain is again relatively low. In part this is a result of the low matrix ultimate strains (about 1.6%) which characterize the highest strength brasses used in this study. The point is clarified in Fig. 3b in which the enhanced strain divided by the brass ultimate strain is plotted against the brass strength. As indicated therein, the relative increase in composite tensile strain increases monotonically with the brass strength.

The enhanced composite ductility is related, in as yet an undefined way, with fracture surface features of the metallic glass in the laminate. Multiple shear banding of the glass (Fig. 4) was often observed near the fracture surface. In addition, multiple shear banding was occasionally found in regions of the composite well removed from the fracture surface (Fig. 5). The fracture surface of the metallic glass was often "roof like" in appearance (Fig. 6a), a feature caused perhaps by cross shearing at the fracture point. The latter interpretation is given some credence by virtue of the cracking that was occasionally found along the roof tip (Fig. 6b).

In summary, the enhanced ductility of the metallic glass in the composite is related to the constraint provided by the adjoining ductile metal. The constraint is manifested by alterations in the fracture behaviour of the glass, as indicated by fracture surface observation, as well as by changes in macroscopic ultimate strains.

3.3. Failure modes

While uniform strains were monitored via strain guage measurements, the failure mode could be investigated by monitoring stress-crosshead displacement (time) for strains which exceed the ultimate one. Schematics of the kinds of failure observed in the metal–metallic glass laminates are shown in Fig. 7. An "instantaneous" failure is said to occur when composite failure takes place immediately following the glass failure and at the same macroscopic strain. Conversely, a "stepwise" failure is associated with further deformation of the matrix following glass fracture.

The situation immediately following the breaking of the glass is schematized in Fig. 8. Region A in this figure represents the portion of the sample removed from the glass fracture and in which the metallic glass

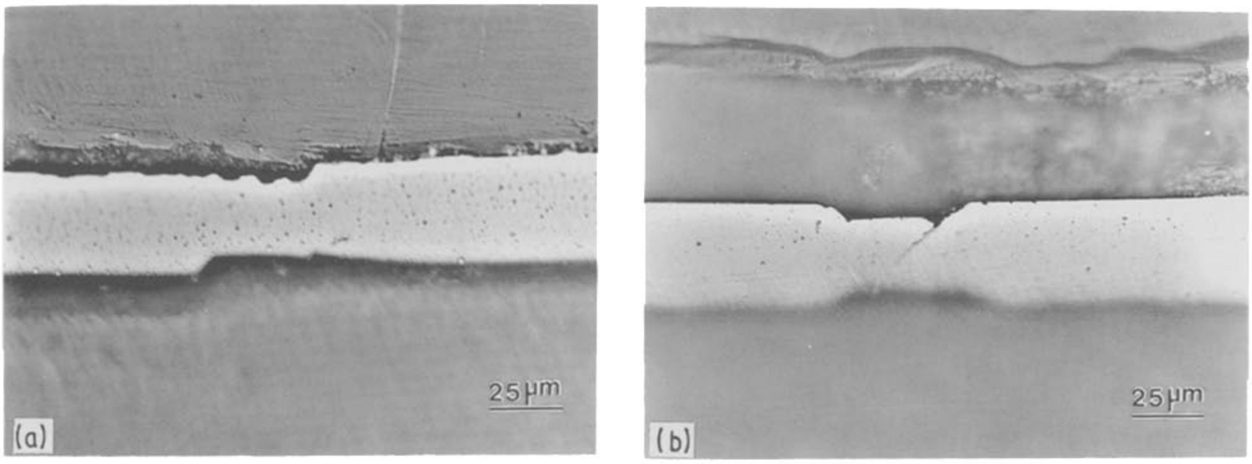

Figure 5 Optical micrographs of the metallic glass in a composite with matrix strength $\sigma_m^f = 600$ MPa. (a) and (b) indicate the existence of multiple shear bands in the metallic glass at regions away from the fracture surface.

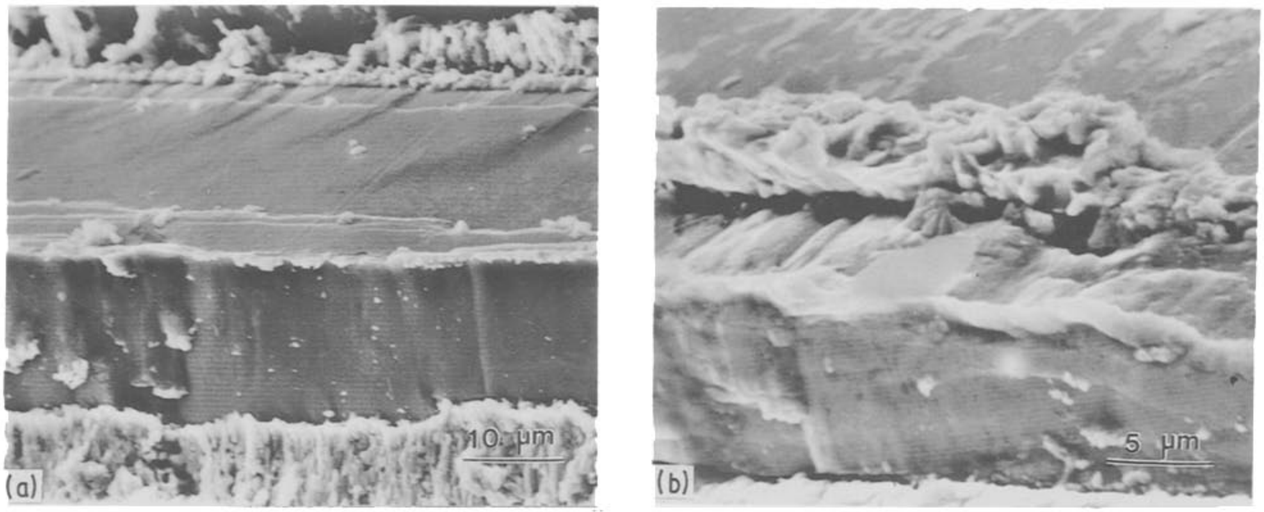

Figure 6 SEM micrographs of the metallic glass fracture surface in the composite with matrix strength $\sigma_m^f = 505$ MPa. (a) A roof-like fracture surface is often found. (b) Cracks are also often observed at the tip of the "roof".

continues to carry tensile load. This area experiences an elastic contraction as a result of the dimished load concommitant with the glass failure. Since the tensile machine operates under the restriction of constant extension, the contraction in Region A is accompanied by an extension in Region B. The axial extent of Region B is, at a minimum, equal to the ineffective length of the glass laminate. Its actual extent is equal to this length plus the length associated with any debonding accompanying initial fracture. The physics of the process have been analysed previously by Craig and Courtney [3] for the case of fibrous composites. Based on a force balance analysis they showed that the strain increment in region B can be expressed as

$$\Delta\varepsilon_B = \frac{\sigma_g V_g}{V_m K_m + E_c}\left(\frac{\delta}{L-\delta}\right) \qquad (2)$$

In Equation 2 the subscripts g and m represent the glass and matrix, respectively, L is the total laminate length, Vs are the respective volume fractions, K is the slope of the brass plastic stress–strain curve at the strain $\varepsilon_0 + \Delta\varepsilon_B$ where ε_0 is the matrix strain at glass failure and E_c is the composite modulus. The increased strain in region B is associated with an increase in stress ($= K_m \Delta\varepsilon_B$). Calling this stress increment $\Delta\sigma_B$ we have

$$\Delta\sigma_B = \sigma_g V_g \Big/ \left[\left(V_m + \frac{E_c}{K_m}\left(\frac{\delta}{L-\delta}\right)\right)\right] \qquad (3)$$

If the increase in stress accompanying the strain extension in region B is sufficient to load the brass in excess of its tensile strength, instantaneous failure results. If not, the failure mode is stepwise.

Knowledge of the flow properties of the brass allows us to calculate the critical length at which the value of $\Delta\sigma_B$ is sufficient to cause instantaneous failure. The results are shown in Table I. If the calculated critical ineffective length is greater than the actual one, instantaneous failure is predicted. We have estimated minimum critical ineffective lengths as $2t\sigma_g/\tau_m$, where t is the glass thickness, σ_g is the ultimate strength of the metallic glass and τ_m is the shear stress at the matrix–metallic glass interface. In calculating we have used the shear strength of the lead–tin solder ($= 38.6$ MN m^{-2}) [4]. On this basis, $\sigma_g/\tau_m = 50$ and with $t = 0.038$ mm, the ineffective length is calculated as 1.9 mm. As noted this is a lower limit on δ, which is equal to the above calculated value plus the debond length. Nonetheless we have used this value of δ for predicting composite failure modes, as indicated in Table I. Composite failure is predicted to be stepwise

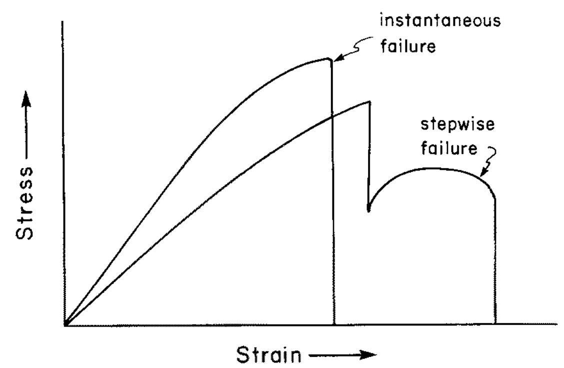

Figure 7 Schematic representation of failure modes of the metal–metallic glass composites. A failure is considered "instantaneous" when composite failure is concurrent with glass failure. Some composites fail in a "stepwise" manner, i.e. composite failure is not coincident with the failure of the glass.

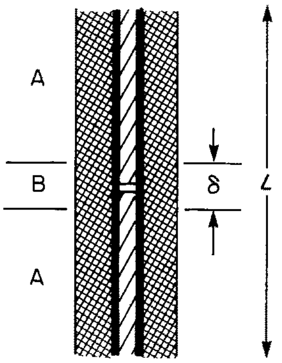

Figure 8 Model of laminate failure in uniaxial tension. When the glass fails in region B, it can not carry stress over a length, δ, the ineffective lamellar length. This leads to a strain and stress concentration in B as a result of the contraction in Region A and the constraint of constant extension in the tensile test. The degree of stress concentration determines whether stepwise or instantaneous failure is observed, as discussed in the text.

TABLE I Observed and predicted failure modes in composites

σ_m^f (MN m^{-2})	$\Delta\sigma_m^c$ (MN m^{-2})	$\left(\frac{\delta}{L-\delta}\right)^c$	δ^c (mm)	Predicted failure mode	Observed failure mode
320	319	0.054	1.03	stepwise	stepwise
400	105	0.064	1.21	stepwise	stepwise
500	35	0.162	2.79	instant.	stepwise or instant.
600	4.8	0.80	8.89	instant.	instant.
700	$\rightarrow 0$	$\rightarrow \infty$	20.0	instant.	instant.

Superscript c represents calculated critical value.

in nature for composites containing brasses having strengths less than 500 MN m^{-2}, whereas the failure should be of the instantaneous kind for composites containing higher strength brasses. The predictions are qualitatively substantiated, and better agreement would be had were an accurate way of evaluating the debond length available.

4. Conclusions

1. Metal–metallic glass composite laminates can be prepared by soldering the composite constituents together, and processing them in a way to assure uniform joint thickness. The strengths of these composites follow the rule-of-mixtures.

2. Although ultimate composite strengths obey the rule-of-mixtures, the glass constituent in the composite fails at a greater strain than the glass constituent when tested alone. Composite failure is accompanied often by the development of multiple shear bands in the glass, indicative of the constraint on deformation placed by the metal adjacent to the glass. The degree to which the strain to fracture of the glass in the composite is increased depends on matrix strength. Higher fractional increments in strain at ultimate load correlate with higher brass strengths.

3. Composite failure mode is either stepwise or instantaneous in nature. That is, the composite experiences additional strain prior to composite failure following glass fracture or it does not. Which of these two scenarios is followed depends on the extent of additional stress carried by the matrix in the region of the material where the glass has failed. If this is high, the composite fails instantaneously and vice-versa.

Acknowledgements

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