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PREDICTION OF THE TRANSVERSE STRENGTH OF GRAPHITE/ALUMINUM COMPOSITES

DENNIS RIGGS
METALS RESEARCH DIVISION

May 1978

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ABSTRACT

The theoretical transverse tensile strength of graphite/aluminum composites was calculated using an elastic analysis technique. It was found that the theoretical behavior of the composite is quite similar to that of a multi-holed flat plate loaded in the transverse direction. Large stress concentrations build up at the fiber-matrix interface resulting in failure at low levels of applied stress. The applied failure stress decreases markedly as the volume fraction of fibers increases. It was concluded that the transverse strength for a 35 volume percent graphite/aluminum composite would probably not exceed 10 ksi (unless interleaving was used, for example) mainly because the transverse properties of the reinforcing fibers are inherently poor and the fiber-matrix bond is very weak.

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INTRODUCTION

The transverse strength of graphite/aluminum composites is quite low in comparison to the ultimate tensile strengths of various aluminum alloy matrices. Typical graphite/aluminum composites exhibit average transverse tensile strengths on the order of 5 to 10 ksi.^{1,2} By comparison, the ultimate tensile strengths of annealed commercially pure aluminum and aluminum alloy 6061 in the T6 condition are 11 ksi and 40 ksi, respectively. The reason for this very low transverse tensile strength is thought to be related to the weak interfacial bond between the graphite fibers and the matrix material. Numerous attempts³⁻⁶ have been made to increase the strength of the interfacial bond by the application of various coatings to the fibers, such as B, TiC, TiB₂, etc., but while enhancing the wettability of the matrix with the fibers, none of the coatings have resulted in significantly improved transverse strengths in the composites.

In this report, reasons for the very low transverse tensile strengths of graphite/aluminum composites will be explored. Essentially, elasticity theory will be used to show that the composites behave similarly to a flat plate with many cylindrical holes, and that the transverse strength is governed not so much by the fiber-matrix bond, but rather by the inherent properties of the carbon fibers.

STRESS CONCENTRATIONS IN PLATES WITH CYLINDRICAL HOLES

Before attempting to estimate the transverse strength of a graphite/aluminum composite, it is helpful to first examine the factors affecting the transverse strength of a limiting case, namely that of a material "reinforced" with cylindrical holes. When these materials are pulled in a direction perpendicular to the long axis of the cylindrical holes, a localized nonuniform stress distribution near the holes will occur. Thus, stress concentrations appear in the material. Shown in Figure 1 is a segment of an infinitely large flat plate with a single cylindrical hole. The plate is being subjected to a uniaxial tensile load of σ_0 in a direction perpendicular to the cylinder axis. From an elastic analysis performed by Timoshenko and Goodier,^{7,8} it was determined that the stresses produced around this hole can be given by the equations:

$$\sigma_r = \sigma_0/2 (1 - a^2/r^2) + \sigma_0/2 [1 + 3(a^4/r^4) - 4(a^2/r^2)] \cos 2\theta$$

$$\sigma_\theta = \sigma_0/2 (1 + a^2/r^2) - \sigma_0/2 [1 + 3(a^4/r^4)] \cos 2\theta$$

1. Launch Vehicle Materials Technology Program. *The Engineering Development of Graphite Fiber-Reinforced Aluminum Composites*. Naval Sea Systems Command, Technical Summary Report, BCL 74-4312-A, December 1976.
2. KREIDER, K. G., ed. *Composite Materials: Metallic Matrix Composites*. Academic Press, New York, v. 4, 1974.
3. LACHMAN, W. L., et al. U.S. Patent No. 3,860,443, January 1975.
4. FITZER, E., et al. *Chemical Vapor Deposition of Pyro-Carbon, SiC, TiC, TiN, Si, and Ta on Different Types of Carbon Fibers*. Carbon, v. 12, 1974, p. 358.
5. FITZER, E., and IGNATOWITZ, E. *The Wetting and Infiltration of Tantalum-Coated Carbon Fibers by Aluminum*. High Temperatures - High Pressures, v. 7, 1975, p. 299.
6. GEBHARDT, J. J. *Development of Carbon-Aluminum Composite Materials*. General Electric Company, RESD, Contract DAAG46-72-C-0185, Final Report, AMMRC CTR 73-36, September 1973.
7. TIMOSHENKO, S., and GOODIER, V. N. *Theory of Elasticity*. McGraw-Hill, New York, 1951.
8. DIETER, G. E. *Mechanical Metallurgy*. McGraw-Hill, New York, 1961.

$$\tau = \sigma_0/2 [1 - 3(a^4/r^4) + 2(a^2/r^2)] \sin 2\theta$$

where σ_r = radial stress

σ_0 = applied stress

a = radius of the hole

σ_θ = tangential stress

τ = shear stress.

When these equations are examined in detail, they show that the maximum stress occurs at point A, when $\theta = \pi/2$ and $r = a$. For these conditions, $\sigma_\theta = 3\sigma_0 = \sigma_{\max}$. The theoretical stress concentration factor for an infinite elastic plate with a single cylindrical hole is therefore equal to 3.

$$\text{Stress Concentration Factor } K_{tg} = \sigma_{\max}/\sigma_0 = 3$$

$\sigma_{\max}$ = maximum stress in the composite

σ_0 = applied stress distant from the hole.

The stress distribution about the hole is shown in Figure 2.

For finite plates, it has been found using photoelastic techniques,^{9,10} that the stress concentration factor K_{tg} is dependent upon the ratio of the hole diameter $d(=2a)$ to the plate width w . Shown in Figure 3 is a plot of K_{tg} versus d/w .

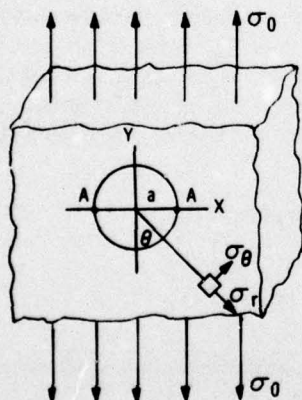


Figure 1. Infinitely large flat plate with a cylindrical hole.

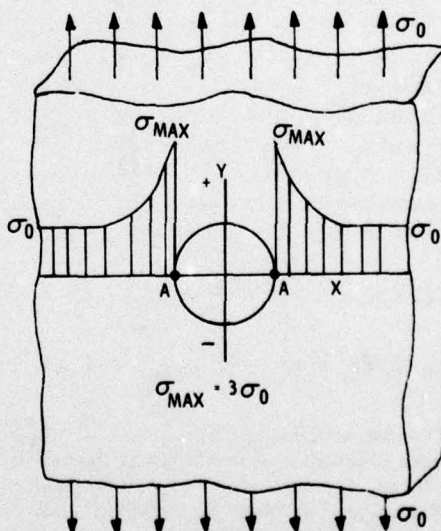


Figure 2. Stress distribution about a cylindrical hole in an infinitely large flat plate.

9. PETERSON, R. E. *Stress Concentration Factors*. John Wiley and Sons, New York, 1974.

10. GRIFFEL, W. *Handbook for Formulas for Stress and Strain*. Frederick Ungar Publishing Company, New York, 1966.

As the ratio d/w increases, the stress concentration factor goes up. For equivalent sized holes, the smaller the plate becomes, the greater the stress concentration about the hole. Thus, a 1/4"-wide aluminum plate stressed to 7,000 psi and containing a single hole the size of a typical carbon fiber ($8\ \mu$) would see a stress of about 21,000 psi immediately adjacent to the hole, assuming an elastic matrix, (see Figure 4). If the matrix behaved plastically, the stress concentration would be reduced.

For finite elastic plates containing large numbers of holes, a similar analysis can be accomplished. For example, consider a flat, multi-holed plate where holes are arranged in a diamond-like pattern (Figure 5). When this plate is pulled in the y-direction, it can be shown¹⁰ that the stress concentrations about points A, B, and C could appear as in Figure 6. In addition, the maximum stress about each hole could be calculated using Figure 7 and the equation:

$$\sigma_A = K_{tg, A} \sigma_0$$

$$\sigma_B = K_{tg, B} \sigma_0$$

$$\sigma_C = K_{tg, C} \sigma_0$$

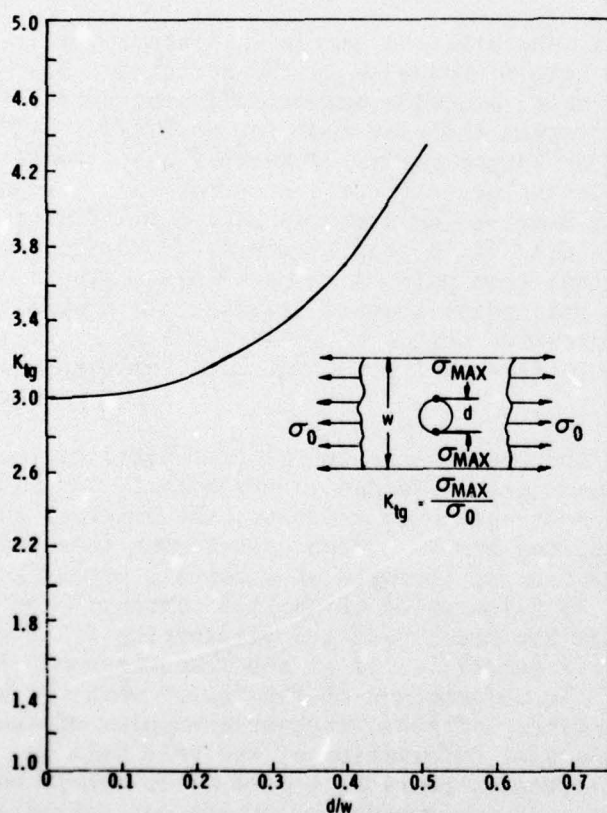


Figure 3. Stress concentration factor for finite plate with cylindrical hole.

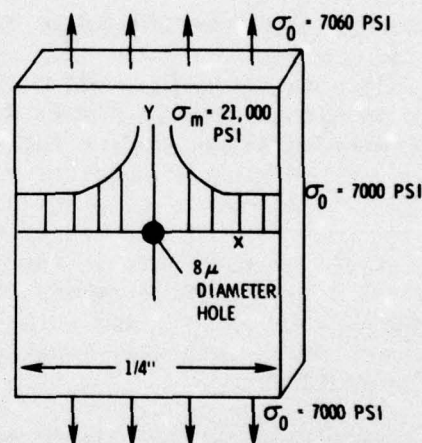


Figure 4. Stress distribution about an $8\ \mu$ diameter hole in a 1/4" aluminum plate.

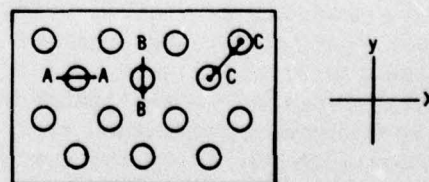


Figure 5. Flat plate with diamond array of hole.

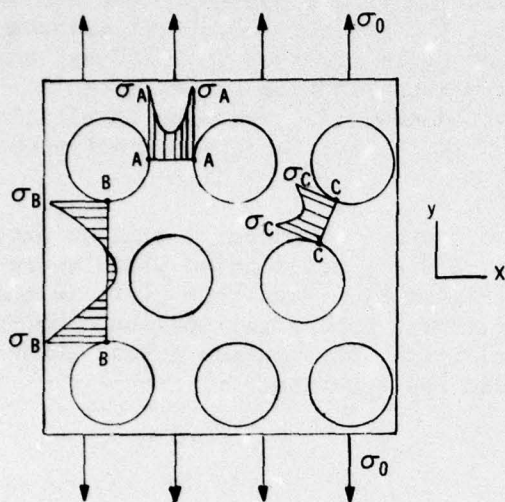


Figure 6. Stress distribution in a multi-holed flat plate under tension.

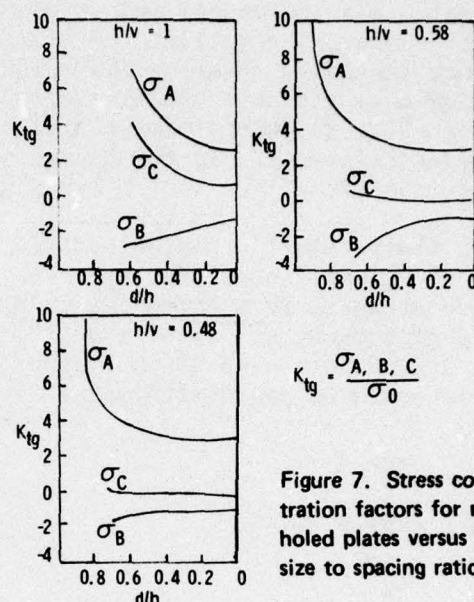


Figure 7. Stress concentration factors for multi-holed plates versus hole size to spacing ratio - d/h .

As can be seen from Figure 7, the stress concentration factor is dependent on the ratios d/h and h/v where d is again the hole diameter, h is the horizontal distance between adjacent holes, and v is the vertical distance between adjacent holes. This is illustrated in Figure 8. These curves indicate that for equivalent center-to-center spacings, larger holes result in larger stress concentrations. Small holes spaced closely together also result in large stress concentrations. As an example, consider an aluminum plate with 3-mil holes arranged in a diamond pattern. Assume the ratio of $h/v = 0.48$ and $d/h = 0.4$. If a tensile stress of 10,000 psi was placed on the plate in the y -direction, then point A on each hole would be subjected to a tensile stress of 32,000 psi, point B would experience a compressive stress of 13,000 psi, and point C a compressive stress of about 1,000 psi. These stresses would initially result in plastic flow, and with continued loading, the plate would fail.

To summarize, in elastic materials the presence of cylindrical holes can lead to significant stress concentrations immediately adjacent to the holes. If plastic deformation does not occur to relieve the stress concentrations, the material can fail at rather low apparent levels of applied stress. When considering this model as a limiting case for explaining the transverse strength of composite materials, it can be seen that failure would occur at a low value of applied stress if the matrix material was fairly brittle and if the modulus of the reinforcing fiber was significantly less than that of the matrix material. It is the transverse modulus of the fibers which will tend to resist the deformation of the "hole" when the composite is pulled in the transverse direction. If this transverse modulus of the fiber is much greater than that of the matrix, deformation of the hole will be impeded when pulled in the transverse direction, provided a good fiber-matrix bond exists, thereby partially (or possibly totally) negating the stress concentration at the interface.* If on the other hand, the modulus of the fiber is significantly

*J. Bluhm, AMMRC, personal communication.

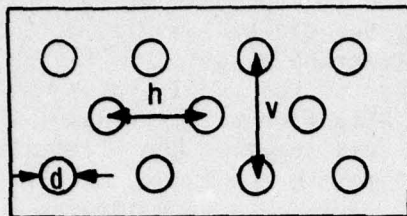


Figure 8. Definition of the parameters v , h , and d .

less than that of the matrix, deformation of the hole will be more or less unimpeded and thus stress concentrations approximately equal to those for flat plates with cylindrical holes will be built up. In a case such as this, the bond between the fiber and matrix should be relatively insignificant. In the following sections, the significance of graphite fiber as a reinforcing agent in the transverse direction will be investigated and the application of the theory to graphite/aluminum composites will be discussed.

MOLECULAR STRUCTURE AND PROPERTIES OF GRAPHITE FIBERS

When graphite fiber is used in a composite material, a significant amount of reinforcement can usually be expected in the direction parallel to the fiber axis. Carbon fibers range in modulus from 30 million psi to 120 million psi with strengths from 200 ksi to more than 500 ksi. The exact values of strength and modulus are highly dependent upon the precursor material, processing conditions, and a number of other factors.

As was discussed in the previous section of this report, the transverse properties of a composite with an elastic (or brittle) matrix will depend significantly on the properties of the fiber.

When graphite is used as the reinforcing material, it should be expected that the fibers will show anisotropies in properties depending upon the orientation of the fiber relative to the direction of applied stress. The crystallographic structure of a graphite crystal is very anisotropic and thus the properties of the crystal significantly depend upon its orientation during testing. For example, the modulus of a perfect single crystal of graphite in the direction parallel to the basal planes, i.e., the "a" direction, is governed by strong, covalent SP^2 bonding.

Thus, the elastic tensile modulus in this direction is very high, being on the order of 146 million psi.^{11,12} On the other hand, the elastic modulus of the graphite single crystal in a direction perpendicular to the basal planes, i.e., the "c" direction, is governed by relatively weak Van der Waals' bonding. As such, a predictably low value of the tensile modulus, being on the order of 5 million psi,^{11,12} is found in this direction. Graphically, the variation in modulus as

11. DIEFENDORF, R. J., and TOKARSKY, E. W. *The Relationships of Structure to Properties in Graphite Fibers*. AFML-TR-72-133, Part III, November 1975.
12. BACON, R. *Carbon Fibers from Rayon Precursors* in *Chemistry and Physics and Physics of Carbon*, P. L. Walker and P. A. Throver, ed., v. 9, Marcel-Dekker, Inc., New York, 1973.

a function of crystallographic orientation in a perfect single crystal of graphite is shown in Figure 9.¹¹ As can be seen, the elastic modulus will drop by a factor of ~ 4 in a direction just 5° from the direction parallel to the basal planes. For graphite fibers, the modulus perpendicular to the basal planes will be similar to that observed for the single crystals. Effective values ranging from 500,000 psi to 2 million psi have been estimated.* This range is due primarily to the more turbostratic nature of the "graphite" composing the fiber, as well as stacking defects, dislocations, microcracks, etc. This is a very important consideration when fabricating graphite-reinforced composites because the orientation of the basal planes is parallel to the longitudinal axis of the fiber. This is shown schematically in Figure 10. Thus, when the composite is subjected to a load in the transverse direction, it is the low modulus perpendicular to the basal planes (the "c" modulus) and *not* the modulus parallel to the basal planes (the "a" modulus) which will govern the response of the fiber to the load (at least in the elastic load regime). Effectively this means that when the composite is loaded in the transverse direction, deformation of the hole which the fiber is occupying will be impeded by a material with a modulus of only about 500 ksi to 2 million psi. In other words, the "reinforcing" fiber effectively has a stiffness approximately 5 to 20 times less than that of the matrix.

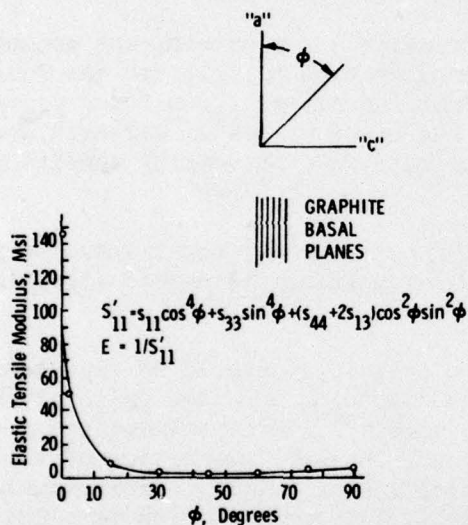


Figure 9. Calculated tensile moduli of a single crystal of graphite as a function of angular displacement ϕ from the "a" axis.

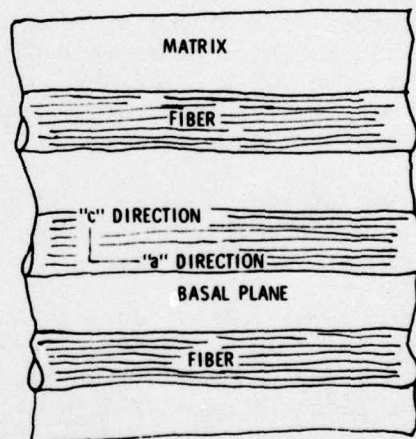


Figure 10. Orientation of basal planes in graphite fiber.

EFFECT OF STRESS ON THE TRANSVERSE STRENGTH OF GRAPHITE/ALUMINUM

In previous sections, the transverse strength of flat plates with cylindrical holes and the anisotropies in both crystal structure and physical properties of graphite fiber were discussed. In this section, the failure characteristics of the matrix material and the nature of the fiber-matrix interface in graphite/aluminum composites will be defined. An estimate of the theoretical transverse strength of the composite will then be made. When defining the characteristics

*R. J. Diefendorf, Rensselaer Polytechnic Institute, personal communication.

of the composites, it can be said that only a marginal degree of bonding exists between the coated graphite filaments and the aluminum alloy matrix. This is reasonable as is evidenced by Figure 11, where the degree of bonding between the fiber and the matrix appears to be quite small. A number of fiber "pull-outs" as well as completely debonded fibers are evident. Second, it can generally be assumed that the fibers possess an approximately circular cross section. This is true for a large percentage of PAN and pitch-base carbon fiber while rayon-base carbon fiber is generally crenulated in appearance. Third, it can be assumed that when the carbon fibers are dispersed in an aluminum matrix, they arrange themselves in roughly diamond shaped patterns. This is in contrast to a highly ordered square or rectangular array. Fourth, when the composites fail, it is assumed that failure will occur in those regions where there exists a high volume fraction of fibers and not in those regions depleted of fibers. Evidence of this can be found in numerous reports.^{13 14} Finally, the most important assumption regards the nature of the matrix material in the fiber-matrix interfacial region. It is reasonable to assume that the fiber-matrix interfacial region is fairly brittle and thus behaves elastically under load. First, the interfacial region should already be plastically deformed due to the stresses arising from the mismatch in thermal expansion between the graphite and the aluminum alloy; second, numerous embrittling carbides are found in the interfacial region as a result of Al-C interactions (Al_4C_3) and the presence of TiC as well as TiB_2 coatings given the fibers to enhance wettability with the matrix; and third, the matrix itself generally possesses a fairly brittle cast microstructure.

The implication of these assumptions, when combined with the fact that the modulus of the graphite fiber is very low perpendicular to the fiber axis, it that the composite is essentially equivalent to an elastic plate with cylindrical



Figure 11. Fracture surface of a Thornel 50/201 aluminum composite. Mag. 700X

13. Progress report submitted to AMMRC by Aerospace Corp.

14. HARRIGAN, W. C., and GODDARD, D. M. *The Effects of Processing Parameters on the Mechanical Properties of Aluminum-Graphite Composites*. Proceedings of the 1975 International Conference of Composite Materials, v. 2, 1975, p. 849.

holes. The fibers should not prove to be an impediment to deformation of the holes when a transverse stress is applied. Therefore, the same equations and curves for stress concentration factors discussed previously should be valid. Using Figure 7 and a model of the composite as shown in Figure 12, an estimate of the stress concentration at the fiber-matrix interface can be obtained. For instance, assume a graphite/6061-T6 aluminum alloy composite containing fibers approximately 8 microns in diameter. By considering Figure 8, and by examining numerous photomicrographs of the composites, the ratios d/h and h/v can be found to be approximately 0.5 and 1.0, respectively. Using these numbers on Figure 7, the maximum stress concentration factor can be obtained. It is found that this stress concentration factor is approximately equal to 6. The ultimate tensile strength of 6061-T6 alloy is about 40,000 psi. Failure of the composite should therefore be expected when this level of stress is achieved in the matrix whether due to stress concentration or uniform loading. Since the stress will be a maximum at the fiber-matrix interface, the applied failure stress can be calculated.

It is found that:

$$\sigma_0 = \sigma_A / K_{tg}$$

$$\sigma_0 = 40,000 \text{ psi} / 6$$

$$\sigma_0 \cong 6,700 \text{ psi failure stress.}$$

This calculated number is in the range of applied stress where graphite/aluminum composites are actually observed to fail in transverse tension. In addition, this theory predicts that failure should occur at the fiber-matrix interface and indeed this is where it is most often observed.

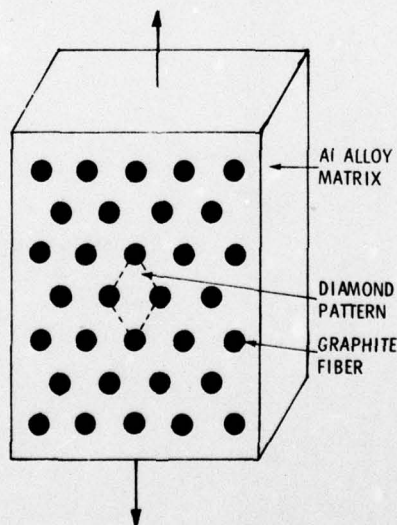


Figure 12. Working model of a graphite/aluminum composite being pulled in the transverse direction.

DISCUSSION AND CONCLUSIONS

The transverse strength of graphite/aluminum composites was modelled by an elastic plate with cylindrical holes. Such a model leads to reasonable strength predictions.

Previous attempts to increase the transverse strength of the composites have generally centered around the improvement of the fiber-matrix bond. It has been shown that the bond plays only a minor role in controlling the transverse properties of this system. In fact, the major reason for the low transverse strength appears to be the very low modulus of the fiber in the direction perpendicular to the fiber axis and the relatively brittle matrix. The longitudinal modulus of the matrix is roughly 5 to 20 times greater than that of the fiber in the transverse direction. No transverse reinforcement can therefore be expected because of the presence of the graphite fiber. When the composite is stressed in the transverse direction, the deformation of the holes is more or less unimpeded and thus the stress concentration factors are equivalent to those exhibited by elastic plates with holes. Composites such as boron/aluminum and FP/Al should, and do, exhibit much higher transverse strengths because the fibers have significantly greater moduli than the matrix in all directions and the bonding between fiber and matrix is much better. In view of these conclusions, it is highly unlikely that the transverse strength of graphite/aluminum composites can be improved without resorting to interleaving of the composites with higher strength materials, such as titanium, or by the addition of whiskers. As the volume fraction of fibers increases, the transverse strength should decrease even further. Two recommendations which may be worthwhile pursuing include: a) controlling the impurities, carbides, and microstructure of the matrix alloys in order to increase plasticity and therefore reduce the effect of the stress concentration; and b) varying the fiber to fiber spacing and arrangements as indicated by Figure 7.

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Watertown, Massachusetts 02172
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Dennis Riggs, ILT, CMIC

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