ORNL/TM-2023/3184

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Nuclear Energy and Fuel Cycle Division

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Prepared by
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managed by
UT-BATTELLE LLC
for the
US DEPARTMENT OF ENERGY
under contract DE-AC05-00OR22725

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ABSTRACT

As inhomogeneous orthotropic materials, composite mechanical properties can be difficult to predict. Even with tight fabrication process controls, there is always some microstructural variation that will affect these properties. As a highly defect-dependent phenomena, mechanical strength is even more subject to variation, especially considering that composites exhibit multiple potential failure modes. The most direct way to determine composite material strength is via experimental testing, maintaining loading and specimen morphology as close as possible to the in situ states. When multiple materials and designs are being considered, this process is often prohibitively costly and slow. However, some approximations and rules of thumb can consistently guide the design process and consideration of alternatives.

This paper discusses composite failure mechanics and presents analytical approaches to failure prediction for each of the failure modes that composites may exhibit. Longitudinal, transverse, tensile, compressive, and shear loadings all produce different ultimate strengths in a composite ply since the resulting failure modes are different. Given the nearly universal shortcomings in purely analytical approaches, an empirical approach is presented based on observations from published data for carbon-epoxy ply failure strengths. It is shown that when constituent fiber strength, constituent matrix strength, and fiber volume fraction are known, ply strength can be predicted consistently to enable evaluation of a range of material design choices and laminate analyses.

1. NOMENCLATURE

Lamina (also *ply* or simply *layer*) = a single layer of a composite material

Laminate (also *layup*) = a composite plate comprised of multiple bonded layers

Laminae = plural of lamina (i.e., multiple layers)

Matrix = the continuous phase of a composite material used to bind the fibers together, to give the fibers a nonzero compression strength, to allow for stress-bridging after fibers are broken

Neat (or *bulk*) = homogenous material without reinforcement (e.g., *neat matrix* without fibers)

Isotropic = a material for which the mechanical properties are the same in any direction

Orthotropic = a material for which the properties are different in each of the three orthogonal directions (but which also exhibits microstructural symmetry, such that when loaded orthogonally there is no shear-normal coupling)

Transversely isotropic = an orthotropic material for which the properties are equivalent in two of three orthogonal directions (e.g., a single ply of continuous fiber reinforced composite) [Note: plies are quite often referred to as orthotropic, although it is more exact to call them transversely isotropic]

Axial (also *longitudinal* or *fiber direction*) = direction aligned with or along the length of a continuous fiber reinforcement

Transverse = direction perpendicular to the length of a continuous fiber reinforcement

Fiber volume fraction (or simply *volume fraction*) = volume of fibers as a percentage (or ratio) of total composite volume

Matrix volume fraction = volume of matrix as a percentage (or ratio) of total composite volume

2. LIST OF SYMBOLS

σ_C^{max} = composite strength

τ_C^{max} = composite shear strength

$\sigma_{C,comp}^{max}$ = composite strength in compression

$\sigma_{fiber}^{max} = \sigma_f^{max}$ = fiber strength

$\sigma_{matrix}^{max} = \sigma_m^{max}$ = bulk matrix strength

V_{f} = fiber volume fraction

V_{m} = matrix volume fraction

E_{fiber} = fiber stiffness (Young's modulus)

E_{matrix} = bulk matrix stiffness (Young's modulus)

G_{ij} = shear modulus in plane "*i-j*"

$G_{\mathrm{fiber}} = G_{\mathrm{f}}$ = fiber shear modulus

$G_{\mathrm{matrix}} = G_{\mathrm{m}}$ = bulk matrix shear modulus

Φ = fiber initial misalignment

γ = strain at matrix yielding

S_L^+ = longitudinal ply strength in tension

S_L^- = longitudinal ply strength in compression

S_T^+ = transverse ply strength in tension

S_T^- = transverse ply strength in compression

S_{LT} = ply shear strength

3. INTRODUCTION

Understanding composite material strength and failure behavior is of paramount importance in structural design. By analyzing the mechanical properties of the individual plies, their orientations, and the applied loads, laminate strength can be predicted. Failure analysis of a laminate is accomplished on a ply-by-ply basis. First-ply failure analysis involves understanding the composite's response to different combinations of loadings while considering factors such as constituent properties and the interactions between adjacent plies. Thus, it is important to know the layer strength properties of any composite material system.

As with elastic constants, composite layer strengths are different in each material direction. Furthermore, analyzing the mechanics of strength and failure is more complicated than the analysis of orthotropic constitutive behavior. Consistent and accurate analytical prediction of lamina strengths based solely on constituent (fiber and matrix) properties is problematic because of two main convoluting factors: (1) composites have multiple potential failure modes, and (2) material failure is a highly defect-driven process, and the presence of defects is both highly variable and fabrication-process dependent.

4. OVERVIEW OF COMPOSITE FAILURE MECHANISMS AND MICROSTRUCTURAL DEFECTS

Even under simple loading conditions, composites exhibit multiple potential failure modes, including fiber breakage, matrix microcracking, fiber–matrix interface debonding, delamination between layers, and fiber microbuckling. The mode by which a laminate exhibits first-ply failure depends completely upon the layup, properties, and applied loading. In some cases of progressive failure, two or more failure modes may be concurrent.

Any microstructural defect becomes a point for failure initiation and subsequent progressive failure. Defects are always subject to significant sample-to-sample or point-to-point variation. Some level of process–microstructure–property relation is always present in materials. None of these relationships could be predicted based solely on constituent materials or on a simple characterization such as fiber volume fraction. Composites have a variety of potential defects or microstructural artifacts that affect strength properties. These features include the presence of matrix voids (either owing to the presence of trapped gases or to incomplete resin infiltration into fibers), potential lack of fiber–matrix interface bond strength, lack of exact fiber alignment, fiber waviness (kinking), or the presence of matrix-rich pockets (which could also be described as imperfect fiber packing).

Voids are always present at some level in the matrix material and are one of the biggest sources of deviation between theoretical strength predictions (based on constituent properties and microstructure) and actual realized experimental strength. Voids can be considered as a loss of cross-sectional area of effective load-bearing material; they also act as miniature internal stress concentrations. Their presence always decreases the realized experimental strength of a composite material. One of the benefits of curing composites under vacuum is to increase resin outgassing of cross-linking by-products (usually hydrogen) that are produced during this reaction. Another benefit is to draw out excess resin, thus increasing fiber volume fraction. During a cure cycle, the effectiveness of outgassing is progressively more limited as the matrix viscosity continually decreases during solidification and cure. Minimization of void size and content is a critical objective of any fabrication process.

The potential for manufacturing-induced cure stresses (pre-stresses) [1] is another source of variation that limits a direct prediction between constituent strengths and lamina strength. Manufacturing-induced stresses strongly depend on cure cycle or fabrication process, and the magnitude of stresses may or may not be a significant factor. Generally, some level of fiber pretension is introduced by thermal contraction of the binding matrix during cooling because the matrix coefficient of thermal expansion is generally much larger than that of a fiber. This pretension will tend to lower prepreg or layer strength compared with a constituent fiber strength.

Several ply strength properties depend on fiber–matrix interface strength. This bond strength is largely chemical in nature. Several mechanisms drive fiber–matrix bond strength, such as electrostatic forces or mechanical interlocking of surfaces, but chemical bonds dominate in magnitude. The interface strength tends to vary depending on the cure cycle used in the fabrication process, possible surface oxidation on fibers, and fiber sizings or coatings that may be present (i.e., it does not simply depend on the constituent fiber and matrix materials). Interface strength can be a limiting strength, especially under transverse loading, in which this failure mode may indeed precede matrix strength. Under axial loading, once the fiber–matrix interface fails, a load path to the embedded fibers no longer exists. Therefore, a weak interface may cause premature ply failure before any fiber breakage occurs. This type of failure would lead to a very large decrease in axial ply strength by a factor of 40 or more (i.e., down to the regime of a polymer matrix strength), but it is generally easily avoided by choosing appropriate material systems.

Axial compression strength is generally lower than axial tensile strength, largely because fiber waviness leads to eccentric loading and microbuckling at the individual fiber scale. This microbuckling is soon followed by matrix cracking and progressive buckling of surrounding fibers. True to the nature of buckling, it will occur at a loading significantly less than what could be achieved by the fiber compressive strength.

Similarly, there will always be some level of imperfect fiber alignment associated with any fabrication method. A ply intended to be oriented exactly at 0° in perfect alignment with loading will always have some nonzero level of misorientation. Although slight misalignments of a few degrees usually do not appreciably affect strength, ply-level failure theories show that an approximately 25% decrease in strength corresponds to a 5° ply misalignment [2].

5. THEORETICAL PLY STRENGTH PREDICTION

Accurate prediction of composite strength based solely on constituent properties remains elusive, despite decades of research and a range of specialized tests devoted to achieving a quantifiable understanding of composite strengths and microstructure-to-property relationships. Theoretical predictions generally assume strong interface bonding, exact fiber alignment, a lack of voids in the matrix, and straight fibers with no kinking. The best use of theory in strength prediction is to set an upper bound that is approachable in an optimized fabrication process and to provide a sanity check on experimental results.

5.1 THEORETICAL AXIAL TENSILE STRENGTH

The axial strength of a carbon fiber polymer matrix composite is dominated by the strength of the fiber. The fiber is the preferred load path in such composite materials, thus a laminate should be designed to align strong fibers in the directions of highest loadings. When a ply is loaded axially, the fiber and matrix phases share a common strain or elongation. Although they are stronger and stiffer than the matrix, the fibers are less ductile. At ply failure (σ_C^{max}), as described in Eq. (1), the stress in the matrix (σ_{matrix}) is less than its failure stress.

$$\sigma_c^{max} = V_f\left(\sigma_{fiber}^{max}\right) + V_m(\sigma_{matrix}) \#(1)$$

At the point of fiber failure, the stress in the matrix is the same as that in the fibers only if they share the same stiffness. In general, this will not be the case, and the level of stress in the matrix will be less, proportional to its lower stiffness [3].

$$\sigma_{matrix} = \left(\sigma_{fiber}^{max}\right)\left({}^{E_{matrix}}\!/\!{}_{E_{fiber}}\right) \#(2)$$

Substituting Eq. (2) for σ_{matrix} in Eq. (1) yields Eq. (3). For carbon fiber polymer matrix composites, the fiber is usually around 50 times stiffer (and stronger) than the matrix material, and fiber volume fractions should be 50% to 75%. Thus, the load taken by the matrix in this case is minimal, and the theoretical longitudinal strength is the product of fiber strength and fiber volume fraction:

$$\sigma_c^{max} = V_f\left(\sigma_{fiber}^{max}\right) + V_m\left(\sigma_{fiber}^{max}\right)\left(\frac{E_{matrix}}{E_{fiber}}\right) \approx V_f\left(\sigma_{fiber}^{max}\right)\#(3)$$

This approximation has two notable exceptions: (1) if the fiber volume fraction is low, then the matrix will fail before any fiber breakage (this design would not be appropriate), and (2) if the fiber–matrix interface is very weak, then the fibers will pull out of the matrix before fracture (and once a fiber is detached from the surrounding matrix, it is largely removed from the load path)

Figure 1 shows a graphical representation of the effect of fiber volume fraction on composite axial strength (σ_C^{max}). If there are no fibers ($V_f = 0$), then the "composite" strength is simply the matrix strength. As fiber volume fraction increases, the fiber content is initially low and only acts to dilute the matrix without adding sufficient strength ("matrix controlled"). At higher fiber volume fractions, strength increases ("fiber controlled") and eventually reaches the fiber strength at $V_f = 100\%$ (although perfect fiber packing is physically impossible).

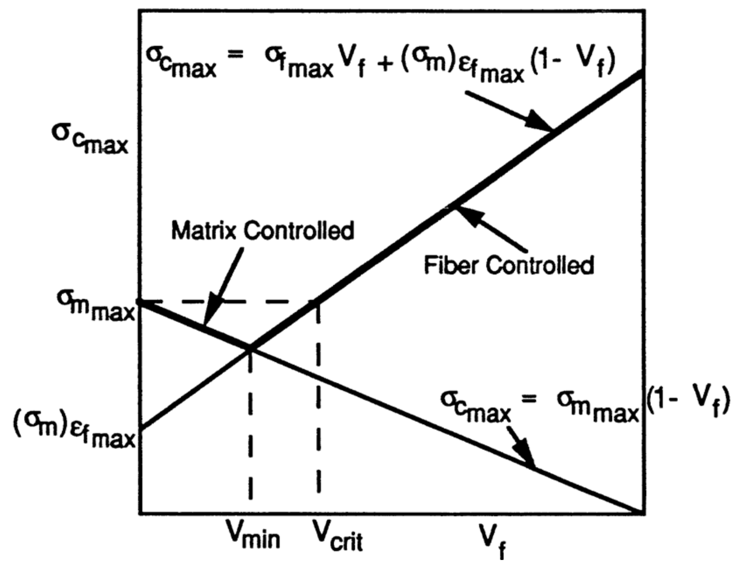

Figure 1. The effect of fiber volume fraction on composite axial strength.

5.2 THEORETICAL AXIAL COMPRESSIVE STRENGTH

Although Eq. (3) can predict idealized fiber crushing in compression by substituting constituent compressive strength into the equation, fiber buckling or microbuckling (local to a kink in the fiber) is much more likely to occur and will do so at a lower stress level. This phenomenon is difficult to consistently predict because it depends on the geometry of the fiber architecture (i.e., level of fiber kinking or misalignment) and on the fiber and matrix stiffness. However, because of this fiber microbuckling failure mode, longitudinal ply compressive strength is generally lower than ply longitudinal tensile strength. Even without fiber buckling, many fibers tend to be weaker when loaded in compression owing to features of their own microstructure (Kevlar fibers are a good example). Some approaches have modeled the fiber as a Timoshenko beam in buckling; however, this model does not incorporate the controlling factor of misalignment or waviness (kinking). Accounting for fiber misalignment has been approached [4] as in Eq. (4):

$$\sigma_{c,comp}^{max} = \frac{G_{12}}{1 - \left(\frac{\Phi}{\gamma}\right)}\#(4)$$

where Φ characterizes the fiber initial misalignment, and γ is the strain at matrix yielding. This fiber misalignment parameter is not necessarily practical, nor is it consistent in magnitude. The authors do set practical observed limits on values $0 < \Phi/\gamma < 8$, giving the expression some relevance in understanding likely limits on strength.

5.3 THEORETICAL TRANSVERSE STRENGTH

The transverse strength of a composite ply is dominated by the matrix strength. In this mode, the fibers are essentially floating freely within the matrix, which can be pulled apart to fracture well before any fiber fracture strength. Under transverse loading, the fiber–matrix interface is sometimes the first mode of failure (i.e., preceding even the matrix microcracking failure). Furthermore, many fibers are themselves orthotropic, such as carbon and Kevlar fibers, and the transverse fiber strength is less than the axial strength. However, transverse fiber failure is unlikely to precede matrix failure or interface failure.

Theoretical predictions of transverse strength have been formulated by assuming that fibers act as a loss of cross-sectional area in the loading plane. Figure 2 shows a square fiber-packing array, consisting of fibers of radius r that are spaced at a center-to-center distance of $2R$.

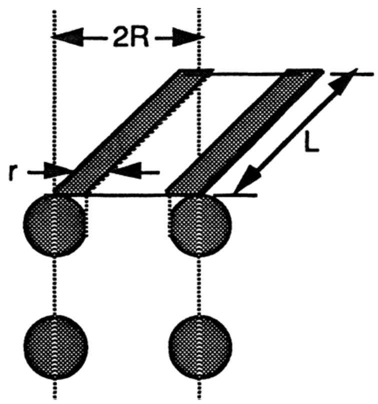

Figure 2. Schematic rectangular fiber packing for micromechanical analyses.

If the fibers are modeled as cylindrical holes, then the repeat unit cell with original surface area $2RL$ loses cross-sectional loading surface area of $2rL$.

$$A_{reduced} = 2RL - 2rL = 2RL\left(1 - \frac{r}{R}\right) \#(5)$$

Fiber volume fraction can be expressed by Eq. (6):

$$V_f = \frac{\pi r^2 L}{2RL(2R)} = \frac{\pi r^2}{4R^2} \#(6)$$

and then rearranged to show the relations given by Eqs. (7a) and (7b):

$$\sqrt{\frac{4V_f}{\pi}} = \sqrt{\frac{r^2}{R^2}} \#(7a)$$

$$2\sqrt{\frac{V_f}{\pi}} = \frac{r}{R} \#(7b)$$

Substituting Eq. (7b) into Eq. (5) shows that the cross-sectional area (*A*) of a fiber-embedded matrix is reduced by a factor determined by Eq. (8):

$$A_{reduced} = A\left(1 - \frac{r}{R}\right) = A\left(1 - 2\sqrt{\frac{V_f}{\pi}}\right) \#(8)$$

Applying this area knockdown factor to the transverse strength yields Eq. (9):

$$\sigma_c^{max} = \sigma_{matrix}^{max}\left(1 - 2\sqrt{\frac{V_f}{\pi}}\right) \#(9)$$

Equation (9) should be applicable to both tensile and compressive loading. Neat matrix compressive strength, although not often reported in supplier or experimental data, is always significantly higher than tensile strength because internal cracks do not propagate in compression, and crack propagation leading to crack coalescence is the failure mechanism of this material (cracks only propagate in tension or shear).

These approaches assume completely regular fiber-packing arrays. They neglect the stress concentration factor of the fiber discontinuity as well as the complex nature of a continuous curved load path throughout the matrix phase. This simplification does not yield accurate predictions, but it does reflect the fact that transverse strength is consistently lower than the neat matrix strength.

5.4 THEORETICAL SHEAR STRENGTH

Ply shear strength is also a heavily matrix dominated failure mode. Although the fibers may offer some level of reinforcement, the matrix load path provides a weakest-link mode of failure. If the ply material directions are labeled as longitudinal (*L*), transverse (*T*), and through-thickness (*Z*), then the *LT* and *LZ* plane shear loadings are largely interchangeable (at the ply level). The *LT* plane represents in-plane shear strength. The *LZ* plane in a multi-ply laminate is often affected by delamination failure modes, but this is not a ply property. The *TZ* plane represents transverse shear. The transverse shear strength is typically observed to be higher than that of the in-plane or delamination shear strength, but appreciable transverse shear loading is uncommon (cases might include thickness direction loading on short beams, short plates, or on tubes with low radius-to-thickness ratios).

Often, the lamina shear strength is estimated as the matrix shear strength. This estimate is approximate but conservative (as shown in Table 2). Some approaches calculate a correction factor based on volume fraction, matrix shear stiffness, and single-fiber shear stiffness [5] as shown in Eq. (10). Fiber shear stiffness is rarely, if ever, reported, and it is difficult to obtain because single-fiber tests are problematic in setup and actuation.

$$\tau_c^{max} = \frac{\tau_{matrix}^{max}}{K_T} \#(10a)$$

$$K_T = \frac{1 - V_f\left[1 - \left(\frac{G_m}{G_f}\right)\right]}{1 - \left(\frac{4V_f}{\pi}\right)^{0.5}\left[1 - \left(\frac{G_m}{G_f}\right)\right]} \#(10b)$$

6. EMPIRICAL COMPOSITE PLY STRENGTH

The most reliable way to characterize composite material strength is via direct testing, maintaining specimen morphology as close as possible to the in situ morphology. However, some approximations and rules of thumb can consistently guide the design process and consideration of alternatives.

To this end, material properties from nine carbon/epoxy material systems were collected and reviewed [2,5,6] to assess differences between theoretical and experimental axial ply strength along with trends in values that can be consistently applied to predict orthotropic strength. Reported values are presented in Table 1, showing experimental axial ply strength (S_L^+) in tension, axial ply strength (S_L^-) in compression, transverse ply strength (S_T^+) in tension, transverse ply strength (S_T^-) in compression, and ply in-plane shear strength (S_{LT}). The fiber volume fraction (V_f) is also shown to indicate fiber content as a fraction of total volume.

Table 2 presents several comparisons. Column 1 shows the ratio of experimental axial layer strength vs. the theoretical maximum strength (from Eq. [3] and Table 1), which is less than one since measured properties are lower than assumption-based theory, as previously detailed. Column 2 shows the ratio of experimental axial strength in compression vs. tensile loading, which is less than one since fiber buckling leads to lower compression strength. Column 3 shows the ratio of experimental transverse ply tensile strength vs. experimental neat matrix tensile strength. Column 4 shows the ratio of experimental transverse strength in compressive loading vs. tensile loading, which is usually larger than one since these materials are stronger in compression as such loads do not lead to microcrack coalescence. Finally, Column 5 shows the ratio of experimental ply shear strength vs. experimental neat matrix strength.

Table 1. Experimental orthotropic prepreg ply strength values [2,5,6]

MATERIAL	S_L^+	S_L^-	S_T^+	S_T^-	S_{LT}
	GPa	GPa	MPa	MPa	MPa
AS4 3501-6 $V_f = 0.63$	2.28	1.44	57.0	228	71.0
AS4 APC2 $V_f = 0.58$	2.06	1.08	78.0	196	157
IM6 SC1081 $V_f = 0.65$	2.86	1.88	49.0	246	83.0
GY70 934 $V_f = 0.57$	0.59	0.49	29.4	98.1	49.0
IM7 8551 $V_f = 0.60$	2.58	1.62	75.8	-	-
T300 5208 $V_f = 0.60$	1.45	1.45	44.8	248	62.1
AS4 3501 $V_f = 0.55$	1.45	1.17	48.0	248	62.0
HM63 8552 $V_f = 0.55$	2.49	1.35	45.0	-	10.0
IM7 8552 $V_f = 0.60$	2.72	1.69	111	-	128

Table 2. Overview of calculated trends in orthotropic strength values

MATERIAL	S_L^+ vs. theory	Axial + vs. -	S_T^+ vs. matrix	Transverse + vs. -	S_{LT} vs. matrix
AS4 3501-6 $V_f = 0.63$	0.90	0.63	1.17	4.00	1.45
AS4 APC2 $V_f = 0.58$	0.89	0.52	0.78	2.51	1.57
IM6 SC1081 $V_f = 0.65$	0.84	0.66	0.98	5.02	1.66
GY70 934 $V_f = 0.57$	0.52	0.83	0.36	3.34	0.59
IM7 8551 $V_f = 0.60$	0.78	0.63	0.76	-	-
T300 5208 $V_f = 0.60$	0.68	1.00	0.92	5.54	1.27
AS4 3501 $V_f = 0.55$	0.60	0.81	0.98	5.17	1.27
HM63 8552 $V_f = 0.55$	0.94	0.54	0.62	-	0.83
IM7 8552 $V_f = 0.60$	0.82	0.62	0.92	-	1.06
Average	0.78	0.69	0.83	4.26	1.21
Std. Dev.	0.14	0.15	0.22	1.08	0.34

6.1 DESIGN GUIDELINE: PLY STRENGTH IN AXIAL TENSION

Based on the results shown in Table 2, some reasonable design guidelines for strength values can be inferred. Ply axial tensile strength (S_L^+) in a carbon–epoxy composite can be predicted from fiber strength and fiber volume fraction as per Eq. (1), which must be decreased by a factor of 0.78 to include the effects of voids, misalignments, and other features that are natural outcomes of the fabrication and cure process. For a fiber volume fraction of 65%, Eq. (11) amounts to about half of the constituent fiber strength.

$$S_L^+ = 0.78\, V_f \left(\sigma_{fiber}^{max}\right) \#(11)$$

6.2 DESIGN GUIDELINE: PLY STRENGTH IN AXIAL COMPRESSION

Based on the results shown in the second column of Table 2, some reasonable design guidelines for strength values can be inferred. Ply axial compressive strength (S_L^-) in a carbon–epoxy composite can be referenced from ply tensile strength (Eq. [11]), which must be decreased by a factor of 0.69 to include the effects of fiber anisotropy, kinking, misalignment, and resultant buckling modes.

$$S_L^- = 0.69\ S_L^+ \#(12)$$

6.3 DESIGN GUIDELINE: PLY STRENGTH IN TRANSVERSE TENSION

Based on the results shown in the third column of Table 2, some reasonable design guidelines for strength values can be inferred. Ply transverse tensile strength (S_T^+) in a carbon–epoxy composite can be referenced from matrix strength, with a reduction factor to account for the presence of fibers aligned against the loading direction as previously discussed in section 5.3.

$$S_T^+ = 0.83\ \sigma_{matrix}^{max} \#(13)$$

In the case of a particularly weak interface strength, such a failure mechanism might be the limiting case. However, interface strength is generally of a similar magnitude to the matrix strength; the distinction can be negligeable.

6.4 DESIGN GUIDELINE: PLY STRENGTH IN TRANSVERSE COMPRESSION

Based on the results shown in the fourth column of Table 2, some reasonable design guidelines for strength values can be inferred. Ply transverse strength (S_T^-) in compression is also a heavily matrix-driven property and is significantly stronger in compression.

$$S_T^- = 4.26\ S_T^+ \#(14)$$

Equation (14) is referenced from the transverse tensile strength (S_T^+) rather than from a constituent matrix compression strength property, largely because of the pervasive lack of availability of such constituent compression test data. A weak interface strength would not be likely to significantly affect the accuracy of Eq. (14) because load is still transferred to the fibers in transverse compression after the fiber–matrix interface has failed.

6.5 DESIGN GUIDELINE: PLY STRENGTH FOR IN-PLANE SHEAR

Based on the results shown in the fifth column of Table 2, some reasonable design guidelines for strength values can be inferred. Ply in-plane shear strength (S_{LT}) in a carbon–epoxy composite can be determined from the neat matrix strength (i.e., from the constituent matrix material property). In general, the fiber-reinforced ply will be about 20% stronger than the pure matrix strength property.

$$S_{LT} = 1.21\ \sigma_{matrix}^{max} \#(15)$$

This strength value should apply to both in-plane (*LT*) and out-of-plane (*LZ*) shear strengths. The same value can be conservatively used for transverse shear strength (*TZ*), although it is observed to generally be a higher strength. For cases in which transverse shear strength is the limiting structural design load, some level of through-thickness reinforcement would be a worthwhile consideration if possible (e.g., 2D weaves, 3D orthogonal or angle-interlock weaves, stitched).

6.6 LIMITATIONS OF EMPIRICAL APPROACHES

The above guidelines were prepared based on a survey of experimental data for carbon fiber epoxy matrix composite laminae. Equation (11) is based on the fiber breakage failure mode. For fiber volume fractions around 40% or less, matrix failure may well precede any fiber failure. In general, deviations from these guidelines could indicate a poor fabrication process. The guidelines should not be used for other material systems. For example, glass fibers are isotropic, and Eq. (13) would likely be somewhat conservative in this case. The guidelines should not be used for ceramic matrix or metal matrix composites, for which the matrix properties may match or exceed those of the fibers. For different fiber or matrix material systems, Tables 1 and 2 should be repopulated to determine appropriate correlations. The guidelines are intended as an estimate of the effect of constituent properties on ply strength properties. They may be used to effectively down-select a variety of material system (fiber and matrix) choices. With the inherent variability of composite material design and fabrication, physical testing is ultimately the best way to find ply properties for component and system designs.

7. USING PLY STRENGTH IN LAMINATE FAILURE ANALYSIS

Failure analysis of a laminate is accomplished with a layer-by-layer analysis, as informed by the ply strength of each layer. Once stresses in each ply have been determined via analytical or finite element approaches, an appropriate failure theory must be applied to determine whether layer stresses exceed an allowable limit. Commonly employed failure theories include the Tsai–Wu or Tsai–Hill orthotropic failure theories. These failure theories quantify the effects of concurrent failure modes and multiaxial loadings to determine a safety factor for each layer. Once any individual ply has failed ("first-ply failure"), this is generally considered to represent structural failure.

8. ACKNOWLEDGEMENTS

Special thanks to Dr. Nancy Ma of ORNL for many detailed discussions that were essential to the foundation of this paper.

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