AIAA JOURNAL
Vol. 30, No. 4, April 1992

Strength Scaling in Fiber Composites

Sortiris Kellas* and John Morton†
Virginia Polytechnic Institute and State University, Blacksburg, Virginia 24061

Factors responsible for scale effects in the tensile strength of graphite/epoxy composite laminates have been studied. Four layups, $(+30_n/-30_n/90_{2n})_s$, $(+45_n/-45_n/0_n/90_n)_s$, $(90_n/0_n/90_n/0_n)_s$, and $(+45_n/-45_n/+45_n/-45_n)_s$, were chosen with appropriate stacking sequences so as to highlight individual and interacting failure modes. Four scale sizes were selected for investigation, including full-scale size, 3/4, 2/4, and 1/4, with n equal to 4, 3, 2, and 1, respectively. The full-scale specimen size was 32 plies thick as compared to 24, 16, and 8 plies for the 3/4, 2/4, and 1/4 specimen sizes, respectively. In general, large-size specimens exhibited a lower tensile strength. Fiber dominated layups were found to be less sensitive to scale effects compared to the matrix dominated layups. It has been observed that the presence of transverse ply cracks, which were associated with the ply thickness, were largely responsible for both strength and failure mode variations in different size specimens. In some cases, such cracks were found in virgin laminates, which meant that subscale models and full-scale laminates of the same layup were effectively dissimilar.

Nomenclature

b	= thickness of $\pm\theta$-deg laminas
C	= constant
d	= semithickness of 90-deg laminas
E_x	= extensional modulus of the laminate
E_θ	= extensional modulus of $\pm\theta$-deg laminas
G_{Ic}	= critical strain energy release rate (mode I)
G_{23}	= transverse shear modulus of 90-deg laminas
m	= constant (shape parameter)
V	= volume
V_i	= volume of the ith scaled specimen (component)
ϵ_2^{TC}	= transverse failure strain of 90-deg laminas
ϵ_{ult}	= strain at failure
λ	= scaling factor (ratio of full scale to model dimension)
σ_f	= critical stress in full-scale structure
σ_i^{ult}	= ultimate tensile strength of ith scaled specimen (component)
σ_m	= critical stress in model
σ_{ult}	= ultimate tensile strength

Introduction

MANY engineering structures evolve from small-scale models, which can be manufactured and tested under controlled laboratory conditions. Consequently, it is important that any effects associated with scaling be identified, well understood, and correlated to model size. Also, in the case of advanced composites, material properties such as strength and stiffness are obtained from small coupons tested under laboratory conditions. It is important to determine whether such measurements are representative of the behavior of large-scale components.

Because of the intricacy of their internal microstructure, fiber reinforced composite materials belong to a special category of materials presenting some complex and, hence, challenging scaling problems. Complications may arise from factors upon which standard similitude laws cannot be satisfied. Such factors include fabrication, fiber diameter, fiber/matrix interface, ply interface, and test method. If these limitations are disregarded, one is then left with two obvious scaling options for laminated composite materials: 1) ply-level scaling, which, superficially, appears to be the better of the two options; and 2) sublaminate-level scaling. Ply-level scaling is achieved when a large-scale laminate, of a given stacking sequence, is constructed from thick layers of the same fiber orientation, each built from a number of standard thickness plies. On the other hand, sublaminate-level scaling is achieved by the introduction of basic sublaminates that are stacked together to form thicker laminates.

Previous research[1-7] has indicated that the strength and stiffness of fiber reinforced composites depend on the size of the composite laminates. It has been demonstrated that the degree of influence depends on the type of scaling level used, the stacking sequence, and the mode of loading. For example, Lagace et al.[3] presented results that showed that each of the three ply-level scaled laminates $(+15_n/-15_n)_s$, $(+15_n/-15_n/0_n)_s$, and $(0_n/+15_n/-15_n)_s$ exhibited a different tensile strength degradation when n was increased from 1 to 5 (in this study, the in-plane dimensions of the scaled specimens were kept constant). The same authors have shown that the tensile strength of otherwise similar, sublaminate-level scaled laminates remained unchanged when n was varied from 1 to 3. In conclusion, these authors have attributed the problem of strength degradation to interlaminar stress effects. Camponeschi,[4] on the other hand, has presented compression data that indicate strength degradation in sublaminate-level scaled laminates. Furthermore, he showed that the degree of compressive strength degradation depends on the material system as well as the laminate thickness.

As a result of the complexity of the problem, due to the large number of variables involved (geometry, material system, layup, stacking sequence, environment, and loading mode), research studies, to date, have failed to establish the exact causes of strength degradation in scaled composite laminates. Consequently, various researchers have different views on what is causing the scale effect. Some have associated the problem with edge effects,[3] whereas others have attacked the problem purely from a statistical point of view,[1,5] or a combination of the two,[6] and some have considered the fracture mechanics approach.[8] Batdorf,[5] for example, has proposed a size relationship for un-notched unidirectional composites that is a slightly modified version of the Weibull theory, which states that the size-strength relationship of brittle materials failing in tension is given by

$$\ln\sigma_{ult} = C-(1/m)\ln V \tag{1}$$

Received Jan. 28, 1991; presented as Paper 91-1144 at the AIAA/ASME/ASCE/AHS Structures, Structural Dynamics, and Materials Conference, Baltimore, MD, April 8-10, 1991; revision received July 31, 1991; accepted for publication Aug. 1, 1991.

*Assistant Professor, Engineering Science and Mechanics Department. Member AIAA.

†Professor, Engineering Science and Mechanics Department.

For geometrically similar brittle materials, Eq. (1) may be written as

$$\sigma_1^{\mathrm{ult}}/\sigma_2^{\mathrm{ult}} = (V_2/V_1)^{1/m} \tag{2}$$

where the shape parameter m is thought to be constant for a given material. Thus, if m can be evaluated from two model size specimens, the strength of geometrically similar components can be predicted.

Atkins and Caddell[8] used a fracture mechanics approach to derive a simple size-strength relationship for notched brittle and isotropic materials. Using elementary similitude laws, they have shown that the stress required to propagate a crack in a full-scale structure σ_f and the equivalent stress in a model structure σ_m is given by

$$\sigma_f = \sigma_m/\sqrt{\lambda} \tag{3}$$

The main objective of the present work was to study and isolate the factors responsible for scale effects on the tensile strength of graphite reinforced epoxy laminates, geometrically scaled at the ply level. A single graphite reinforced epoxy system has been studied, Magnamite AS4/3502. Four layups, $(+30_n/-30_n/90_{2n})_s$, $(+45_n/-45_n/0_n/90_n)_s$, $(90_n/0_n/90_n/0_n)_s$, and $+45_n/-45_n/+45_n-45_n)_s$, denoted A, B, C, and D, respectively, have been chosen with appropriate stacking sequences so as to highlight individual and interacting failure modes. Note that, for convenience, the stacking sequences A, B, and C may be written as $(\pm 30_n/90_{2n})_s$, $(\pm 45_n/0_n/90_n)_s$, and $(\pm 45_n/\pm 45_n)_s$, respectively. Four scale sizes have been selected for investigation, including full-scale size, 3/4, 2/4, and 1/4, with n equal to 4, 3, 2, and 1, respectively. The full-scale specimen size was 32 plies thick as compared to 24, 16, and 8 plies for the 3/4, 2/4, and 1/4 specimen sizes, respectively. All three dimensions—length, width, and thickness—were proportionally scaled. The nominal dimensions of the full-scale specimens were 508 mm long, 50.8 mm wide, and 4 mm thick. The nominal gauge length for the full-scale size was 356 mm.

Results are presented in the form of tensile strength, stress/strain curves, and damage development. Factors responsible for strength degradation with increasing specimen size are isolated and discussed. Inconsistencies associated with strain measurements are also identified. Enhanced X-ray radiography is employed for damage evaluation, following incremental loading.

It is shown that, in general, the tensile strength deteriorates with increasing specimen size. The fiber dominated layups B and C are less sensitive to scale effects compared to the matrix dominated layups A and D. It is observed that transverse ply cracks, which were associated with ply thickness increases, are largely responsible for the observed strength and failure mode variations in different size specimens. In some cases, such cracks are found in virgin laminates, which meant that model and full-scale laminates of the same layup are effectively dissimilar.

Extrapolation to the full-scale strength is attempted by means of two basic methods: a Weibull statistics based model, Eq. (2), and a fracture mechanics based model, Eq. (3). The predictive performance of each one of these models is assessed and their applicability to the scaling problem of laminated composites is discussed.

Experimental Program

Mechanical Testing

Prior to mechanical testing, all specimens were stored in a dry environment at room temperature. All tests were carried out on a universal screw driven testing machine at a constant strain rate equal to 0.028/min. The stress/strain behavior of several specimens, per size and layup, was carefully monitored using custom built extensometers. Four such extensometers were designed and built to accommodate the four scaled specimen sizes. The observed stress/strain behavior was then verified against data from specimens instrumented with both strain gauges and one of the extensometers. Apart from layup C, specimens from the other layups were instrumented with both central and edge gauges. Only one specimen per layup and size was tested in this way.

Damage Evaluation

Penetrant enhanced X-ray radiography was employed as a nondestructive damage evaluation technique. Specimens were soaked in zinc-iodide solution prior to being X-rayed. Since the same specimen was used in a load increment/damage evaluation procedure, it was assumed that the penetrant solu-

Table 1 Strength and strain at failure (average values from six or more valid tests)

Size	σ_{ult}, MPa	ϵ_{ult}, %	$\sigma_{\mathrm{ult}}/(\sigma_{\mathrm{ult}})_{\mathrm{f.s}}$	$\epsilon_{\mathrm{ult}}/(\epsilon_{\mathrm{ult}})_{\mathrm{f.s}}$
	Layup A: $(+30_n/-30_n/90_{2n})_s$			
1/4	208.8	0.60	1.83	1.88
2/4	156.5	0.55	1.37	1.74
3/4	131.1	0.33	1.15	1.04
Full scale	114.3	0.32	1.00	1.00
	Layup B: $(+45_n/-45_n/0_n/90_n)_s$			
1/4	557.0	1.20	1.39	0.82
2/4	498.9	1.18	1.24	0.81
3/4	427.3	1.47	1.06	0.97
Full scale	402.3	1.47	1.00	1.00
	Layup C: $(90_n/0_n/90_n/0_n)_s$			
1/4	884.4	1.38	1.07	1.48
2/4	872.6	1.17	1.05	1.26
3/4	865.9	1.25	1.04	1.34
Full scale	830.3	0.93	1.00	1.00
	Layup D: $(+45_n/-45_n/+45_n/-45_n)_s$			
1/4	135.3	1.05	1.56	2.49
2/4	117.8	0.96	1.36	2.29
3/4	103.1	0.74	1.19	1.77
Full scale	86.6	0.42	1.00	1.00

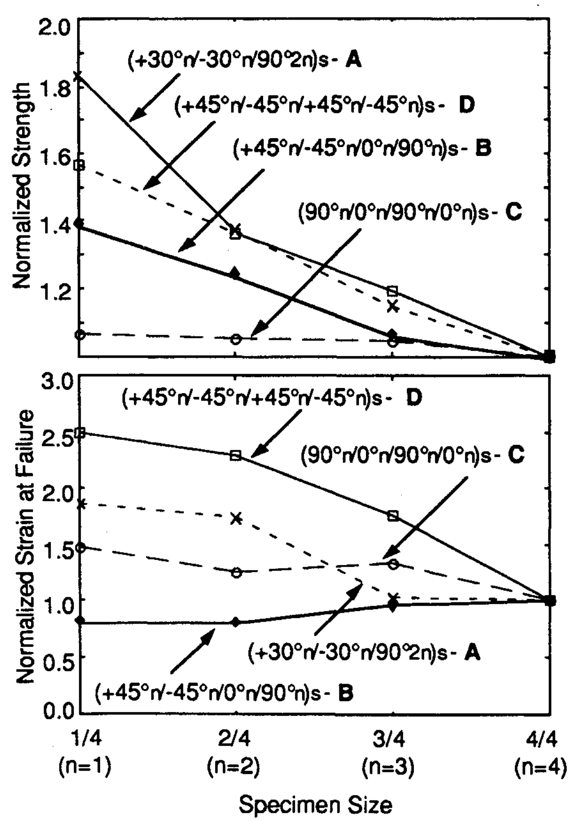

Fig. 1 Normalized strength and strain at failure vs specimen size.

Table 2 Initial stiffness values (0.2, 0.5, and 0.35% strain for A, B, C, and D, respectively)

	Initial stiffness, GPa			
Size	Extensometer	Central gauge	Edge gauge	Lamination theory[a]
	Layup A: $(+30_n/-30_n/90_{2n})_s$			
1/4	35.2	37.2	37.9	46.2/44.1
2/4	35.9	36.5	38.6	46.2/44.1
3/4	35.9	35.2	33.1	46.2/44.1
Full scale	42.1	34.5	33.8	46.2/44.1
	Layup B: $(+45_n/-45_n/0_n/90_n)_s$			
1/4	46.9	48.3	49.0	55.8/53.8
2/4	46.9	46.9	49.6	55.8/53.8
3/4	—	—	—	55.8/53.8
Full scale	44.8	44.1	48.3	55.8/53.8
	Layup C: $(90_n/0_n/90_n/0_n)_s$			
1/4	64.8	67.6	—	77.2/73.8
2/4	70.3	69.0	—	77.2/73.8
3/4	—	—	—	77.2/73.8
Full scale	—	—	—	77.2/73.8
	Layup D: $(+45_n/-45_n/+45_n/-45_n)_s$			
1/4	15.2	16.5	16.5	18.6/20.0
2/4	16.5	17.2	17.9	18.6/20.0
3/4	16.5	16.5	15.2	18.6/20.0
Full scale	19.3	16.5	15.2	18.6/20.0

[a]Two sets of material properties were used. The first set of data was supplied by the material manufacturer and the second by Jackson.[7]

tion had no effect on the fracture characteristics of the epoxy. To validate this assumption, as far as possible, tests for a given specimen were carried out in as short a period as possible to reduce the amount of penetrant absorbed by the matrix. Several sequential loading increments per specimen were used, each one followed by damage evaluation using X-ray radiography. The load increments were determined from predetermined stress/strain plots.

Results

Strength

The tensile strengths $\sigma_{\rm ult}$ and strains at failure $\epsilon_{\rm ult}$ can be found in Table 1, together with normalized values for strength and strain at failure. Normalizations have been carried out with respect to the full-scale values. The normalized results are also plotted in Fig. 1. Strength is defined as the maximum attained load divided by the cross-sectional area of each specimen. Likewise, strain at failure is defined as the maximum recorded displacement divided by the extensometer gauge length. Three points should be noted in Table 1 and Fig. 1:

1) The tensile strength depends on the specimen size: the greater the size, the smaller the strength. This is true for all four layups. However, the degree of influence depends on the percentage of 0-deg plies in a given layup: the more 0-deg plies, the lower the strength related scale effect.

2) So far as the strength is concerned, the scale effect appears to be diminished with increasing specimen size; that is, it would appear that when a certain specimen size is reached (not necessarily the full-scale size used here) scaling effects tend to a limiting value. For example, see layup A, Fig. 1.

3) The strain at failure is also affected by the specimen size. However, it appears that the strain at failure depends on the stacking sequence rather than just the number of 0-deg plies in a given laminate.

Stiffness

The stiffness for each specimen type was determined from the stress/strain curves, obtained from both the extensometer and the strain gauge readout. Apart from layup C, specimens from all other layups exhibited a nonlinear stress/strain response. Therefore, the reported values for stiffness, shown in Table 2, represent the initial slope and are valid for small strains only.

Even though the stiffness value derived from strain-gauge data is not strictly statistically meaningful, the results listed in Table 2 suggest that, for small strains at least, the value of the measured stiffness is independent of the location of measurement. Furthermore, it would appear from the gauge results that all specimen sizes of a particular layup share approximately the same initial stiffness value. However, the entire stress/strain response depended on the layup, the specimen size, and the method of measurement.[9] For example, specimens of layup C displayed approximately the same stress/strain response throughout the loading range. On the other hand, specimens of layup B exhibited a stress/strain response that was more sensitive to specimen size and the method of strain measurement. Typical examples of stress/strain responses for 1/4- and full-scale size specimens of layup B are shown in Fig. 2. It is shown that a sudden drop in the stiffness of the full-scale specimen that was detected by the extensometer, Fig. 2b, was not registered by the strain gauge. This sudden drop in stiffness was later found to be associated with the formation of delamination. In general, strain gauges were inadequate for providing ultimate strain values since the gauges were usually damaged (detached) prior to specimen final failure.

An inconsistency between the extensometer and the strain-gauge reading was recorded for layups A and D, where the extensometers indicated a slight increase in specimen stiffness with increasing specimen size and applied load. A typical example of such an inconsistency is depicted in Figs. 3 for a full-scale size specimen of layup A. In the case of the 1/4-scale size specimen, both strain indicators recorded a similar material response, as shown in Fig. 3a. However, in the case of the full-scale specimen, Fig. 3b, a lower strain was detected by the

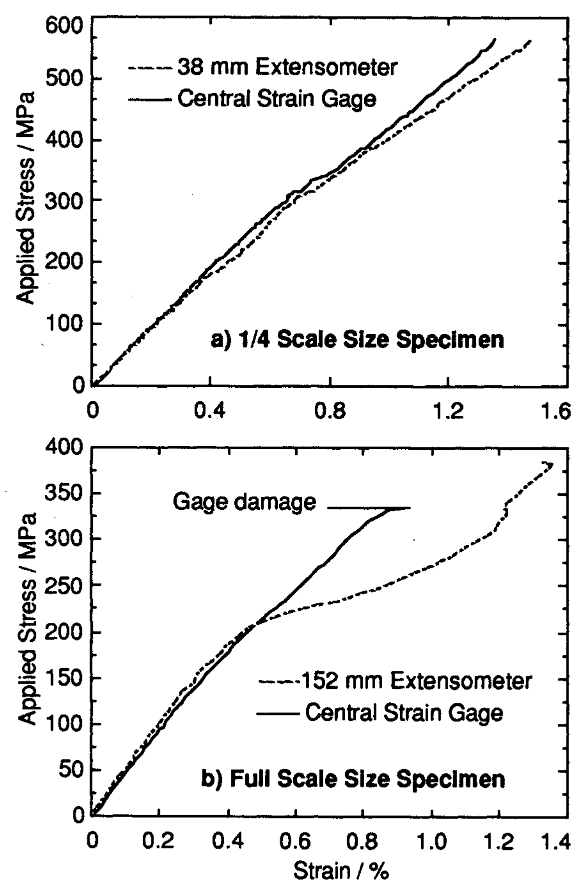

Fig. 2 Typical stress/strain responses of 1/4- and full-scale specimens of layup B $(+45_n/-45_n/0_n/90_n)_s$: a) 1/4-scale size specimen; b) full-scale size specimen.

extensometer. This behavior was thought to be a result of fiber scissoring in the cracked surface plies, which leads to extensometer knife-edge slipping.

Failure Modes

For the fiber dominated layups (B and C), the modes of final failure depended on the specimen size. In fact, contrary to the strength behavior, the layup containing the largest amounts of 0-deg plies was much more sensitive to failure-mode related scale effects than laminates with less or no 0-deg plies. For example, so far as the tensile strength is concerned, specimens of layup C (50% 0 plies) showed very little dependence on size. However, even though the strength in all four sizes was comparable, the mode of final failure was completely different. The mode of failure changed from a clean fracture in the 1/4-size specimens to a brush-like fracture in the full-scale specimens. On the other hand, specimens of layups A and D (no 0-deg plies) that showed large strength related size dependency exhibited no apparent failure related size effects.

Nondestructive Examination

Following curing and postcuring, all panels were C-scanned for quality assessment. Results indicated a slight but consistent deterioration in panel quality with panel thickness. This was particularly true in an area close to the panel edges.

X-ray radiography (on sliced specimens) indicated that even before the load was applied, large-scale specimens contained substantial interply matrix cracks. This type of damage was related to layup configurations but also the number of plies grouped together in a given laminate.

In most cases, delamination appears to have evolved as a result of extensive matrix damage at the free edges. Such delamination was more pronounced and, in general, appeared at a lower percentage of strength in the large-size specimens. For example, for layup A, substantial delamination has occurred in the full-scale size specimen at 103 MPa (approximately 90% of the average value of strength). On the other hand, equivalent delamination in the 2/4-scale specimens has occurred at 152 MPa (approximately 97% of the average value of strength). Likewise, delamination along the entire gauge length was evident in the full- and 1/4-scale size quasi-isotropic specimens when the applied stress was 207 and 414 MPa, respectively. These stresses corresponded to 51% of the average failure stress for the full-scale size specimen as compared to 74% of the failure stress for the 1/4-scale size specimen.

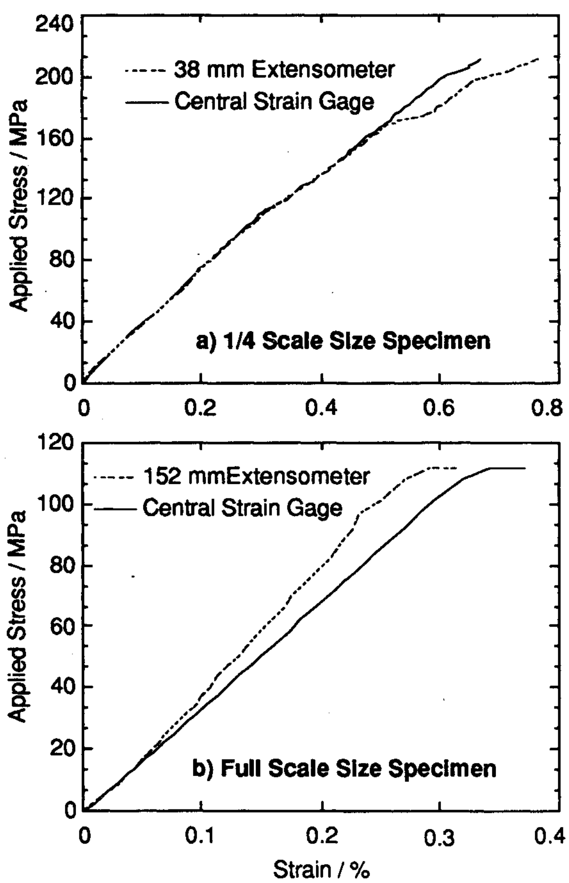

Fig. 3 Typical stress/strain responses of 1/4- and full-scale specimens of layup A $(+30_n/-30_n/90_{2n})_s$: a) 1/4-scale size specimen; b) full-scale size specimen.

Discussion

Experimental Program

Laminate stacking sequences were chosen so as to promote a variety of failure mechanisms including fiber fracture, delamination, and matrix transverse cracking. By doing so, at least one structurally inadequate layup had to be considered, layup A.

The specimen minimum and maximum sizes were selected to satisfy certain constraints. The first constraint was set by test standards. For example, the 2/4-scale specimen size was chosen to comply with existing ASTM D 3039 standards for specimen geometry. The second constraint was set by the capacity of the loading frame. This determined the limit for the full-scale specimen size.

Damage

Transverse Cracks

Interply cracks, which are often referred to as transverse matrix cracks, were present in all thick virgin specimens. Some of these cracks pre-existed in the cured panels and some formed during cutting and propagated transversely, along the fiber directions, from one free edge to the other. In general, the thicker the plies, for a given layup, the larger the density of the observed cracks. It is believed that some cracks pre-existed in the uncut panels since noise was emitted from thick panels during cool down, following the manufacturer's recommended curing cycle. However, the exact density and location of such cracks could not be verified by subsequent ultrasonic C-scanning. The technique is not sensitive enough to distinguish between a collection of microvoids and a collection of matrix microcracks. Further cracking was noted during specimen slicing since more noise was emitted from the thick panels. Cutting induced vibration, or large free-edge stresses, or a combination of the two, are thought to have been responsible for triggering these extra cracks.

In addition to the triggering mechanism, there must have existed a driving mechanism that would cause the cracks to propagate. It is generally accepted that thick laminates, scaled on a ply level, may suffer more from free-edge interlaminar stresses compared to corresponding thin laminates.[3,6] Such stresses, however, cannot be solely responsible for the observed cracks since delamination rather than transverse cracking would be a more appropriate resulting damage mode. Residual curing stresses could also be responsible for driving the cracks since these stresses possess all of the right attributes (sign and direction) to give rise to the observed damage. However, lamination theory would suggest that residual stresses should be the same in all scaled sizes for a given layup. Thus, if matrix cracks should develop due to curing stresses, these should be observed in all four sizes, and not only in the laminates with the thickest plies. If one accepts this argument and the validity of the lamination theory, there remain at least two possible explanations for the premature development of transverse cracks in the thick laminates: a fracture mechanics or a statistical distribution based approach such as those described by Eqs. (2) and (3). Both of these equations suggest that thicker plies, in otherwise identical layups, will crack at a lower applied, or residual, stress.

From an entirely different point of view, it is possible to relate the development of transverse cracks to the in situ transverse strength of off-axis plies, which is thought to approach a minimum value (the unconstrained strength) as the ply thickness increases.[10] The subject of ply constraint has received a lot of attention; however, most studies have dealt with laminates in which only the transverse ply thickness was increased.[10-14] In these cases, the ply thickness ratio b/d was a variable, which meant that the influence of the residual stresses on the strength of the transverse plies was also a variable. It can be shown that, for geometrically scaled specimens where the ratio b/d is a constant, typical expressions for the transverse failure strain of the 90-deg plies, such as Eq. (4), which has been derived by Flaggs and Kural[10] from the model of Parvizi et al.,[11]

$$\epsilon_2^{TC} = \left[\frac{bE_\theta G_{23} G_{1c}^2}{d^2 E_2^3 (bE_\theta + dE_2)}\right]^{1/4} \tag{4}$$

can be reduced to Eq. (3). Thus, no additional information is gained into the driving mechanism of the transverse cracks. It would appear that the subject of constrained cracking in geometrically scaled specimens deserves more attention.

Modes of Failure

The modes of final failure in the fiber dominated layups (B and C) depended on the specimen size, whereas the contrary was true for the two matrix dominated layups (A and D). These transitions in the modes of final failure can be explained through the effect of transverse cracking on the load bearing plies. For example, in the matrix dominated layup A, final failure is largely controlled by the load bearing ± 30-deg plies. In this case, the mode of final failure was more or less uniform in all four sizes because failure along the ± 30-deg directions was the only alternative: no fibers could be broken. The effect of transverse cracking was merely reflected in the tensile strength. The large-size specimens, in addition to the 90-deg ply cracks, suffered early cracking in the ± 30-deg directions, which effectively reduced the tensile strength.

For the fiber dominated layups (B and C), the effect of matrix cracks was largely reflected in the mode of final failure, rather than in the tensile strength. For example, matrix cracks in the 90-deg plies of layup B, of the small-size specimens, appeared to be responsible for promoting fracture in the load bearing 0-deg plies. In other words, transverse cracks in the 90-deg plies imposed a stress concentration on the neighboring 0-deg plies. As a result, a clean 0-deg ply fracture occurred. It is believed that the difference in the mode of final failure, between the small- and the large-size specimens, lies in the decoupling rate between the 90-deg plies (source of stress concentrations) and the 0-deg plies (load bearing plies). Delamination in the large-size specimens occurred at a much lower percentage of strength, as compared to the smaller size specimens. Hence, the 0-deg plies in the large-size specimens can survive the local stress concentrations imposed by the transverse cracks in the 90-deg plies. Furthermore, the -45-deg plies that were already cracked delaminated from the 0-deg plies at a faster rate. Thus, as the applied load increased, the chance of localized fracture in the load bearing 0-deg plies was reduced further.

In the case of the cross-laminates (layup C), the presence of extensive matrix cracks in the load bearing plies of the large-size specimens effectively served as the decoupling mechanism, as delamination did in the quasi-isotropic laminates. Since the 0-deg plies of the large-size specimens were severely split, a local fiber fracture could not have propagated transversely through the whole width of the 0-deg plies as it did in the case of the small-size specimens.

Delamination

Tensile strength reductions with increased ply thicknesses have been attributed to edge stress effects that were thought to be responsible for causing delamination at lower applied loads.[3,6] Although this may be true for certain specific layups and sizes, the generality of such an approach is questionable. In the present research work, it has been shown that ply decoupling influences both the final failure mode as well as the strength of scaled laminated composites. However, such ply decoupling is not always associated with pure delamination, as described by Lagace et al.[3] and Rodini and Eisenmann.[6] As the ply thickness increases, interply transverse cracking becomes the primary strength controlling mechanism and delamination is simply a secondary damage mode. Since ply decoupling depends on the specimen size, as demonstrated by the modes of final failure, this will always be a strength controlling factor in the scaling of laminated composites and has to be taken into account in any mechanistic approach for strength prediction. In fact, in a follow-up study,[15] angle-ply laminates were examined from a sublaminate scaling point of view. It was found that both the strength and the strain at failure were increased with increasing specimen size. In this case, first ply failure occurred in the form of interply cracks that developed in the surface plies. The effect of these interply cracks had a completely opposite effect, having an adverse effect in the thinner laminates. Therefore, simple stand-alone strength prediction models based on volume of the material, free-edge stress distribution, or the void size are not expected to be sufficient for general applications. The effect of damage, such as transverse cracking and delamination, cannot be disregarded.

Strength

Failure criteria, which are presently used to determine the strength of composite materials, are empirical in nature and phenomenological; that is, the mode of failure cannot be predicted. Moreover, their success in predicting the strength of a composite depends on the accuracy of empirical parameters such as the strength degradation factor,[16] for example. Mechanistic approaches, although more effective, require more input data and, often, require a powerful computer. Bearing in mind that none of the existing theories or failure criteria, phenomenological or otherwise, can account for size effects in composite laminates, it is expedient that the simplest available techniques be considered first. In this case, a reasonable first attempt for predicting the full-scale strength of a composite laminate from the strength of modest sizes could involve techniques that are often used in the prediction of the strength (or a critical stress) in homogeneous, brittle, and isotropic materials, such as those described by Eqs. (2) and (3).

Nevertheless, laminated composites are neither homogeneous nor isotropic, and moreover, ultimate failure does not occur as a result of a self-similar crack propagation; hence, the application of either Eq. (2) or (3) in the prediction of the laminate strength should be viewed with caution. A reasonable approach would be to apply these equations to determine the onset of self-similar growing cracks such as transverse cracks, where Eq. (4) could be replaced by Eq. (3). However, the successful application of Eq. (2) or (3) to predict ultimate failure and, hence, strength depends on the layup, as shown in Fig. 4, which contains a comparison of the Wiebull model, Eq. (2), and the fracture mechanics based model, Eq. (3). For the purpose of direct comparison, the ratio of the predicted to the measured strength was plotted (for each layup) vs the specimen size. Note that the closer this ratio is to 1, the better the agreement between theory and experiment. It appears that, in general, the tensile strength of full-scale size specimens, as predicted by Eq. (3), is underestimated for all four layups: from 3%, in the case of layup A, to as much as 26%, in the case of cross-ply layup C. Clearly, in the case of the matrix dominated layups, this simple relationship between size and strength [as described by Eq. (3)] appears to be valid to a reasonable degree. The large error, for the fiber dominated layups, is not unexpected since the self-similar crack propagation concept is completely violated. Note that, according to the damage propagation observations, the full-scale specimens with fiber dominated layups have retained their strength by

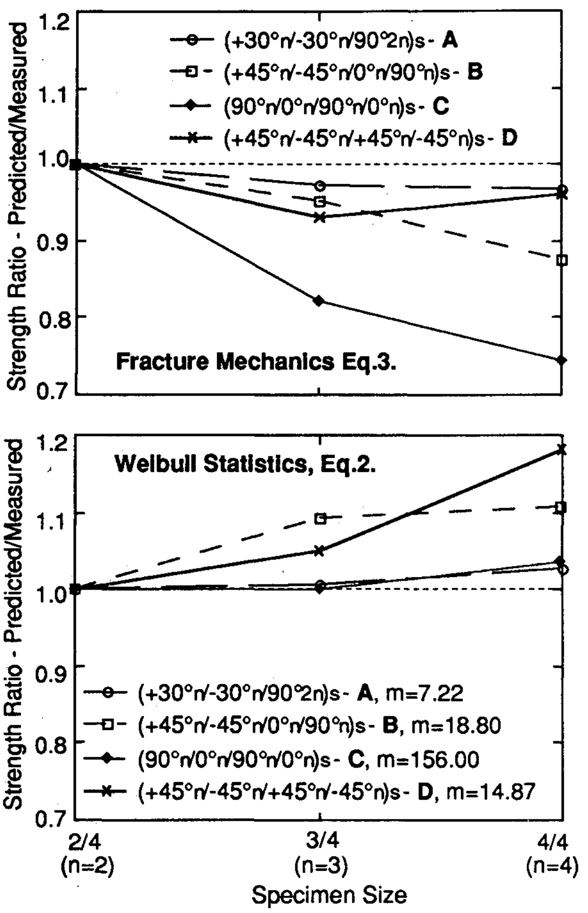

Fig. 4 Strength predictions of scaled specimens using Eqs. (2) and (3).

means of ply decoupling (which is effectively a crack arrestment process).

In contrast to Eq. (3), Eq. (2) requires data from two model sizes for the calculation of the empirical parameter m. The shape parameter m was evaluated for each layup from the two smallest sizes. Note that, for both the 1/4- and 2/4-scale sizes, the ratio of the predicted to the measured strength is equal to 1, hence, the 1/4-scale size was omitted from Fig. 4. Since the model of Eq. (2) involves an empirically obtained parameter, its predictive capability appears to be much better than that of Eq. (3).

Unlike the strength predictions of Eq. (3), the strength predicted by Eq. (2) has been overestimated in all four layups. In particular, the strength of the full-scale specimens of layup D has been overestimated by as much as 18%. This overestimate can be attributed to the shape parameter m, which was evaluated from the 1/4- and 2/4-scale size specimens. For layup D, these two sizes, unlike the 3/4- and full-scale size specimens, were uncracked prior to testing. On the contrary, in the case of layup A, only one of the two specimens used to calculate the shape parameter m was cracked. Thus, the strength of the full-scale size specimens of layup A was predicted within 2.5%. Clearly, the parameter m is a function of the initial state of the specimens used in its determination and a function of how the initial state of such specimens affects the ultimate mode of failure.

Obviously, the predictive performance of Eq. (2) will depend on a correct estimate for the shape parameter m, which depends on the uniformity of the material's internal micro- or macrostructure as the specimen size increases. Note that the word uniformity refers to any material changes that may have a positive or negative effect on strength. For example, transverse cracks were present in approximately equal amounts in both layups C and D. The ±45-deg cracks present in layup D had a negative effect on the strength of that laminate. For layup C, on the other hand, the negative effect due to the 90-deg ply cracks was offset by the simultaneous development of 0-deg transverse cracks (which effectively led to ply decoupling, a positive contribution to tensile strength). Thus, as shown in Fig. 4, the full-scale strength of layup C was predicted correctly within 3.7% as opposed to 18% difference for layup D. The mode of final failure in the quasi-isotropic specimens suggested that some degree of ply decoupling did take place. However, unlike the cross-ply layup, the strength of the quasi-isotropic laminates depended on the integrity of the 0-deg plies as well as the integrity of the ±45-deg plies (which make up 50% of the laminate). Thus, as one would expect, the predictive accuracy of the full-scale strength of the quasi-isotropic layup lies halfway between that of the cross-ply and the angle-ply layup.

Strain at Failure

The strain at failure was found to be much more sensitive to the method of measurement as compared to strength. Furthermore, results showed that there was no simple correlation between the strain at failure, the type of layup, and the specimen size. For layups A, C, and D, the failure strains tend to increase with decreasing specimen size; however, an opposite effect was observed in the case of layup B. It would seem appropriate, therefore, to conclude that the sensitivity of the strain at failure to the specimen size depends on the stacking sequence as well as the layup and the method of measurement.

Strain-based failure criteria should have an advantage over stress-based criteria since strains in the longitudinal and transverse directions are coupled directly. Conversely, the stresses in most failure theories are considered to be independent.[17] However, the correct application of strain-based criteria requires ply strains at failure to be measured experimentally. Results from the present study showed that such measurements are sensitive to the type of method used. Therefore, additional and sometimes substantial errors can be introduced into such criteria.

In this study, all four chosen layups had off-axis plies on the outer surface, which meant that strain gauges attached to those surfaces would be incapable of measuring the maximum strain at failure due to damage. In contrast, the extensometers exhibited a relatively more consistent behavior, although some slipping did occur, especially for specimens with severely cracked surface angle plies.

Stiffness

In addition to design considerations, understanding of scale effects becomes important when standard methods have to be specified. The measured strains at failure and the theoretical predictions by lamination theory are presented in Table 2. These results suggest that for small strains the stiffness is approximately independent of the specimen size, but depends somewhat on the method of strain measurement.

Despite material property conflicts, lamination theory provided an acceptable agreement with experiment, with the best match occurring in layup D and the worst in layup A. The fact that experiment and lamination theory are approximately 25% apart, in the case of layup A, may be attributed to the extensive matrix damage associated with this layup due to ply grouping. This particular layup had the largest number of grouped plies compared to the other three layups.

It has been shown that, as the strain increased, the stress/strain response became more sensitive to both the method of measurement as well as the specimen size. For example, a significant loss in stiffness took place in specimens of layup A at a stress of just over 138 MPa in the 2/4-size specimens. X-ray radiography showed that this applied stress level was associated with the initiation of delamination. An interesting deviation of the stress/strain curve has occurred also in the large-size specimens of layup B. This deviation was registered only by the extensometers and was later associated with the initiation of edge delamination. This is an important observation since it demonstrates the limitation of small gauge length strain measurement sensors. In general, global effects that may significantly alter the specimen's elastic behavior cannot be registered by local strain measuring devices, unless of course the sensor (strain gauge in this case) happens to be located in the vicinity of the damage.

Opposite to the strain gauge reading, the extensometers registered a marginal increase in stiffness in the largest size specimens of layups A and D, see Table 2. These two layups contained angle plies on the outside and no 0-deg plies. The reason for this behavior is knife-edge slipping due to fiber scissoring. The heavily cracked outer plies, in the full-scale specimens, have a tendency to align themselves with the loading direction as the applied load increased, hence causing extensometer slipping. Such extensometer slipping was successfully eliminated when the knife edges were located in V notches cut in a layer of soft resin brushed across the specimen width.[18]

Conclusions

Damage

Damage development and the final mode of failure were found to be size sensitive. It was observed that the degree of size sensitivity depended on the layup. Matrix dominated layups showed the least dependence on scale size. In contrast, modes of final failure in the fiber dominated layups were more sensitive to size effects. It has been shown that the premature development of interply transverse cracks has contributed to this observed behavior. Such damage was a function of the ply thickness and often led to ply decoupling, which is not necessarily associated with the word delamination. The rate of ply decoupling was the most important factor in controlling the evolution of damage and the mode of final failure and, consequently, the ultimate strength of the laminate.

Strength

So far as strength is concerned, all four layups were found to be scale-size sensitive. The degree of sensitivity was very much dependent on the given layup. An 82.7% increase in strength was observed in 1/4-size specimens, of the matrix dominated layup A, as compared to the full-scale size specimens. In contrast, the strength of the 1/4-size specimens of the fiber dominated layup C was only 6.6% higher than the strength of full-scale specimens.

Prediction of the full-scale strength has been attempted, with limited but reasonable success, by the use of two basic methods that are usually applicable to homogeneous, brittle, and isotropic materials: 1) a Weibull statistics based model, Eq. (2), and 2) a fracture mechanics based model, Eq. (3). The full-scale strength was overestimated by Eq. (2) and underestimated by Eq. (3) for all layups. The best full-scale strength predictions were obtained from the first method, which is thought to be the most appropriate to the strength scaling problem.

It has been shown that the predictive performance of the Weibull statistics based model depended on the material uniformity with increasing size. The more uniform a composite material was, the better the extrapolated full-scale strength.

Strain at Failure

The strain at failure was found to be sensitive to the method of strain measurement. Furthermore, results showed that there was no simple correlation between the strain at failure, the type of layup, and the specimen size. Although in most cases (layups A, C, and D) the failure strains tend to increase with decreasing specimen size, an opposite effect was observed for layup B, which may have been associated with ultimate strain measurement errors rather than material response.

Stiffness

For small strains, the stiffness values appeared to be independent of specimen size as lamination theory would predict. However, at large-strain values the stiffness depended on both the specimen size as well as the method of measurement. It has been shown that the stress/strain behavior can be correlated to observed damage mechanisms, provided that an appropriate method of strain measurement is used.

An apparent strain hardening that was recorded by the extensometers was attributed to a scissoring effect of the surface angle plies in the matrix dominated layups A and D.

Acknowledgments

This study was supported by NASA Langley Research Center under NASA Grant NAS1-18471. Thanks are due to the contract monitors Huey D. Carden and Karen E. Jackson.

References

[1]Zweben, C., "The Effect of Stress Nonuniformity and Size on the Strength of Composite Materials," *Composites Technology Review,* Vol. 3, No. 1, 1981, pp. 23–26.

[2]Morton, J., "Scaling of Impact-Loaded Carbon-Fiber Composites," *AIAA Journal,* Vol. 26, No. 8, 1988, pp. 989–994.

[3]Lagace, P., Brewer, J., and Kassapoglou, C., "The Effect of Thickness on Interlaminar Stresses and Delamination in Straight-Edged Laminates," *Journal of Composites Technology & Research,* Vol. 9, No. 3, 1987, pp. 81–87.

[4]Camponeschi, E. T., Jr., "Compression Testing of Thick-Section Composite Materials," David Taylor Research Center, Ship Materials Engineering Dept., Research and Development Rept. DTRC-SME-89/73, Oct. 1989.

[5]Batdorf, S. B., "Note on Composite Size Effect," *Journal of Composites Technology & Research,* Vol. 11, No. 1, 1989, pp. 35–37.

[6]Rodini, B. T., and Eisenmann, J. R., "An Analytical and Experimental Investigation of the Edge Delamination in Composite Laminates," *Fibrous Composites in Structural Engineering,* edited by E. M. Lenoe, Pleum, New York, 1978, pp. 441–457.

[7]Jackson, K. E., "Scaling Effects in the Static and Dynamic Response of Graphite-Epoxy Beam-Columns," NASA TM 102697, July 1990.

[8]Atkins, A. G., and Caddell, R. M., "The Laws of Similitude and Crack Propagation," *International Journal of Mechanics Science,* Vol. 16, 1974, pp. 541–548.

[9]Kellas, S., and Morton, J., "Strength Scaling in Laminated Composites," NASA CR 4335, Nov. 1990.

[10]Flaggs, D. L., and Kural, M. H., "Experimental Determination of the In Situ Transverse Lamina Strength in Graphite/Epoxy Laminates," *Journal of Composite Materials,* Vol. 16, 1982, pp. 103–116.

[11]Parvizi, A., Garrett, K. W., and Bailey, J. E., "Constrained Cracking in Glass Fibre-Reinforced Epoxy Cross-Ply Laminates," *Journal of Materials Science,* Vol. 13, 1978, pp. 195–201.

[12]Bader, M. G., Bailey, J. E., Curtis, P. T., and Parvizi, A., "The Mechanisms of Initiation and Development of Damage in Multi-Axial Fibre-Reinforced Plastic Laminates," *Proceedings of the ICM 3, Mechanical Behaviour of Materials,* Vol. 3, Cambridge, England, UK, Aug. 1979, pp. 227–239.

[13]Crossman, F. W., Warren, W. J., and Wang, A. S. D., "Influence of Ply Thickness on Damage Accumulation and Final Fracture," *Proceedings of the Symposium on Mechanics of Composite Materials and Structures,* American Society of Mechanical Engineers, Boston, MA, 1983, pp. 215–226.

[14]Wang, A. S. D., "Fracture Mechanics of Sublaminate Cracks in Composite Materials," *Composites Technology Review,* Vol. 6, No. 2, 1984, pp. 45–62.

[15]Kellas, S., and Morton, J., "Scaling Effects in Angle Ply Laminates," NASA CR 4423, Feb. 1992.

[16]Tsai, S. W., *Composites Design,* 4th ed., Think Composites, Dayton, OH, 1988.

[17]Hart-Smith, L. J., "A New Approach to Composite Laminate Strength Prediction," Douglas Paper 8366, MIL-HDBK-17 Committee Meeting, Riviera Beach, FL, Oct. 1989.

[18]Kellas, S., Morton, J., and Jackson, K. E., "An Evaluation of the ±45° Tensile Test for the Determination of the In-Plane Shear Strength of Composite Materials," *Proceedings of the Eighth International Conference on Composite Materials (ICCM/VII),* Paper 36-M-3, Honolulu, HI, July 1991.