Polyester Composites Reinforced with Noncrimp Stitched Carbon Fabrics: Mechanical Characterization of Composites and Investigation on the Interaction between Polyester and Carbon Fiber

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ABSTRACT: The primary purpose of the study is to investigate the anisotropic behavior of different noncrimp stitched fabric (NCF) reinforced polyester composites. Carbon fiber composite laminates were manufactured by vacuum infusion of polyester resin into two commonly used advanced noncrimp stitched carbon fabric types, unidirectional and biaxial carbon fabric. The effects of geometric variables on composite structural integrity and strength were illustrated. Hence, tensile and three-point bending flexural tests were conducted up to failure on specimens strengthened with different layouts of fibrous plies in NCF. In this article an important practical problem in fibrous composites, interlaminar shear strength as measured in short beam shear tests, is discussed. The fabric composites were tested in three directions: at 0°, 45°, and 90°. Extensive photomicrographs of multilayered composites resulting from a variety of uniaxial loading conditions were presented. It was observed that broken fibers recede within the matrix in composites with weak interfacial bond. Another aim of the present work was to investigate the interaction between carbon fiber and polyester matrix. The experiments, in conjunction with scanning electron photomicrographs of fractured surfaces of composites, were interpreted in an attempt to explain the instability of polyester-resin–carbon-fiber interfaces. J Appl Polym Sci 102: 4554–4564, 2006

Key words: carbon fiber; anisotropy; scanning electron microscopy; vacuum-assisted resin transfer molding

INTRODUCTION

Carbon-fiber-reinforced epoxy resins are widely used in the aerospace applications and defense industries because of the superior combination of stiffness, strength, and fatigue resistance that these materials offer, whereas polyester resins that command the greatest attention in the field of glass reinforcement comprise a major part of the reinforced plastics market today. The combination of carbon fibers and polyester matrix is becoming more important as the cost of carbon fibers is decreasing, and because of the development of new composites manufacturing technologies.[1] The vacuum-assisted resin transfer molding (VARTM) process allows a greater degree of automation than does hand lay-up, provides an improved work environment (fewer volatile emissions), improves product quality, and is more cost-effective. The VARTM or the patented SCRIMP (Seemann composite resin infusion molding process)[2,3] processes have been developed as alternative low cost methods for the manufacture of composite structures. The resin infusion processes lend themselves to the use of near net shape textile preforms manufactured through a variety of automated textile processes such as knitting and braiding.[4] The challenge facing the resin infusion techniques is to design a robust process that will consistently ensure complete infiltration and cure of a geometrically complex shape preform with the high fiber volume fraction needed for structural applications. One major disadvantage of the resin infusion processes is that they require long duration and high temperature cure cycles to cure the resin-saturated performs fully.

Relatively little research has been reported on the mechanical performance of carbon fiber/unsaturated polyester composites despite its importance in many dynamically loaded structures.[5,6] Although the number of combinations of carbon fabrics and unsaturated polyester resin are popular among designers, relatively small amount of data is available.

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TABLE I
Specifications of Multiaxial Multiply Carbon Fabrics

		Fibrous layers				Stitching			
Preform ID	Description	Areal density (g/m^2)	Fiber	Fiber count in tow	Orientation (°)	Stitch	Linear density (tex)	Knit pattern	Gauge (needles/in.)
UDCF	Unidirectional carbon fabric	200	Toray T700 50E	12K	0	PES	5	Tricot	5
BACF	Biaxial carbon fabric	400	Toray T700 50E	12K	+45; −45	PES	5	Tricot/chain	5

The textile industry has developed the ability to produce net-shape/near-net-shape fabrics using highly automated techniques such as stitching, weaving, braiding, and knitting. Multiaxial multiply fabrics (MMF), also called ''noncrimp fabrics'' (NCF) are a promising class of composite preforms that consists of unidirectional (UD) plies arranged in a number of possible orientations relative to the fabric warp direction; the individual plies are stitched together by warp knitting process by stitching yarns piercing through the fibrous plies.

During NCF-based composites manufacturing, preforms are laid up with desired stacking sequence on the mold tool and infiltrated by a thermoset resin to form the composite. Compared with the time-consuming and expensive UD tape layout, the MMFs are produced in one step, and so the lay-up time is drastically reduced. The use of this preforms overcomes the disadvantages of the wrinkling that is normally experienced with standard woven fabric and prepreg tape. For this reason, the use of composites reinforced with NCF is growing rapidly in complex structural components.

All of the variables that affect the performance of composite structures have been the subject of numerous investigations.[7–14] The success of these studies is indicated by the constantly increasing strength levels obtained in reinforced plastic systems.

The relationship between the mechanical properties and the process in which the MMF and composite are manufactured has received considerable attention in recent years. Many of these studies, both theoretical and experimental, the individual properties of multiaxial multilayer warp knit (noncrimp) fabric reinforced polymer composites, and some of the predictive models available for determining them have been extensively reviewed elsewhere[15] and will be discussed further only where directly applicable to this study.

The effect of through-the-thickness stitching on the in-plane mechanical properties of fiber-reinforced polymer composites has been studied from many viewpoints using both empirical and theoretical methods. In spite of the difficulties of correlation and the differences of test conditions between the classical models and reinforced composites, many of the controlling parameters should be common to both systems. Mouritz et al.[16] gave an excellent review of the stitching problems encountered in the fiber-reinforced polymer (FRP) composites.

Recent articles by Drapier and Wisnom[8,9] attempted to investigate and model the behavior of MMF composites subjected to compression and shear loading.

It is unfortunate that very few significant work efforts are based upon experiments carried out with damage development in MMF composites.[12,17] Bibo et al.[18] have shown how crimp in the tows has a pronounced effect on the mechanisms of failure in the NCFs, but with subtle differences driving failure in tension and compression.

The work of Truong et al.,[19] where the mechanical properties of the composites were measured in a number of orientations relative to the stitching direction for different NCF-based laminates, had shown that the stitching has limitations on stiffness of MMCF. On the other hand, when the damage development is investigated by C-scan and X-ray imaging, the relation between stitching and damage (as a result of crack initiating resin rich pockets created by stitching) patterns is observed.

The major goal of the research described in this article has been to study the anisotropy of the carbon noncrimp fabric reinforced polyester composites. The study of anisotropy was carried out to investigate the effects of geometric variables, such as directions of fiber orientation, on structural integrity and strength of the quasi-isotropic (QI) laminate built from biaxial carbon NCF blankets and UD carbon fabric laminates. Hence, tensile, three-point bending and short beam shear tests were conducted at different off-axial angles (0°, 45°, and 90°) with respect to the longitudinal direction. Another aim of this study was to identify

TABLE II
Compositions of Noncrimp Stitched Fabric Composites

Composite	Layup	Number of fabric layers	Laminate thickness (mm)
Unidirectional carbon fabric	$[0]_{12}$	12	5.78 ± 0.02
Quasi-isotropic carbon fabric	$[45/-45/90/0]_6$	12	6.10 ± 0.03

TABLE III
Mechanical Properties of Carbon Fiber/Polyester Composites in Different Directions[a]

Test direction	Fiber volume fraction V_f (%)	Tensile modulus (GPa)	Tensile strength (MPa)	Elongation at break (%)	Flexure modulus (GPa)	Flexure strength (MPa)	ILSS (MPa)
Unidirectional carbon fabric							
Machine		71.345 ± 2.6	641.07 ± 14.1	0.866 ± 0.04	50.445 ± 1.3	542.27 ± 12.8	21.545 ± 0.73
Bias	23.1 ± 1.1	6.7401 ± 0.4	23.480 ± 0.8	0.387 ± 0.03	6.6410 ± 0.4	57.530 ± 1.5	4.3530 ± 0.24
Cross		4.5440 ± 0.3	9.8980 ± 0.5	0.203 ± 0.03	4.3770 ± 0.3	22.068 ± 0.76	2.1940 ± 0.1
Quasi-isotropic carbon fabric							
Machine		35.248 ± 1.1	353.17 ± 11.5	1.293 ± 0.06	21.674 ± 0.7	200.68 ± 10.5	5.2290 ± 0.34
Bias	44.5 ± 1.8	38.324 ± 1.2	386.63 ± 13.1	1.039 ± 0.04	21.682 ± 0.7	172.85 ± 8.76	8.4630 ± 0.52
Cross		39.615 ± 1.1	391.12 ± 13.6	1.032 ± 0.03	22.306 ± 0.7	211.76 ± 10.8	7.4230 ± 0.45

[a] The data quoted are all average results taken from a minimum of six tests.

the dependence of fracture surface on this off-axial variation. In addition to the extensive efforts in elucidating the variation in the mechanical properties of UD and quasi-isotropic laminate, the work presented here focuses, also, on the type of interactions that are established between polyester and carbon fiber. Scanning electron microscope (SEM) was also helpful and additionally used to describe the morphological features of fractured surfaces of carbon–polyester composites.

EXPERIMENTAL

Materials and sample preparation

Unidirectional (UD) and biaxial (BA) carbon fabric provided by Metyx Telateks Tekstil Urunleri San. ve Tic. A.S. (Turkey) using a Liba multiaxial warp-knitting machine was used for experimental characterization. The carbon fabric parameters, as specified by the manufacturer, are shown in Table I. In the quasi-isotropic carbon fabric laminates, the fabric layers were separated, giving twelve layers, the first layer being (0/90) orientated (BA fabric turned by 45°), the second (+45/−45), the third (0/90), and the fourth layer being (+45/−45) orientated. This sequence is repeated three more times to make up the twelve layers. The compositions of these materials, which in this article will be referred to as the noncrimp stitched fabric composites, are given in Table II.

Polipol polyester 383-T resin system was used as resin in the composite. The resin (specific gravity, 1.11; viscosity brookfield, 950), which is isophthalic acid type resin, was mixed before VARTM with the catalyst cobalt octoate (0.35 pph, of a 41% solution in white spirit), the retarder 2.4-pentanedione (0.10 pph), and methylethylketone peroxide (2.2 pph, of a 40% dimethyl phthalate solution).

The VARTM equipment and a hot press at the GOVSA Composites Ltd. were used to manufacture composite plates. Two types of composite plates having different fiber volume fractions, shown in Table III

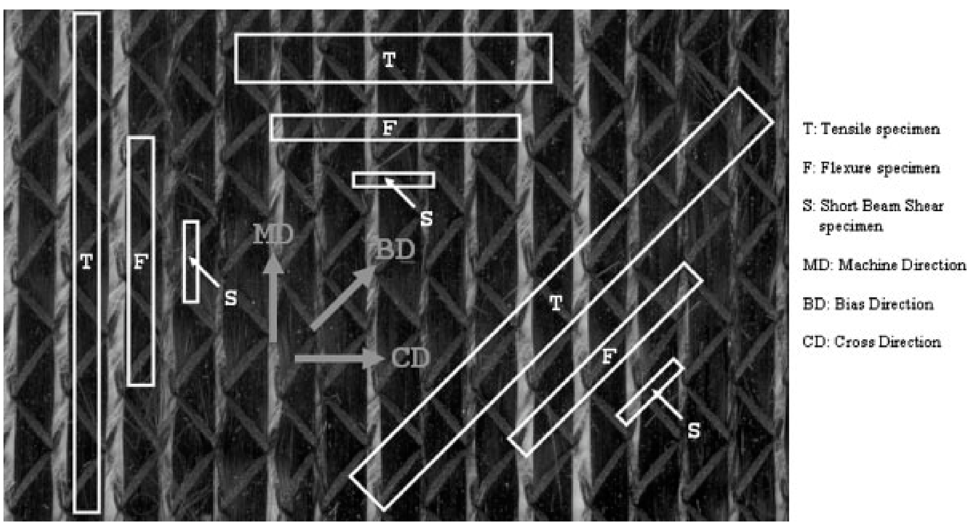

Figure 1 Location of test pieces cut from the laminates used for mechanical testing and directions of measurements.

(the data obtained by a resin burn-off method are the reported mean values from the manufacturer), were used for the test specimens. The testing specimens were cut as illustrated in Figure 1. The final dimensions of each test specimen are identified as in Figure 2.

Figure 1 illustrates the test directions performed on the specimens where machine direction (MD) is the flow direction of fabric in the machine during production, while bias (BD) and cross direction (CD) are 45° and 90° relative to MD, respectively.

Tensile strength testing

The tensile experiments were conducted by longitudinal (0°), transverse (90°), and balanced crossply (45°) tension (ASTM D 3039) on a Shimadzu AUTOGRAPH AG-G Series universal testing machine with a video extensometer system (SHIMADZU noncontact video

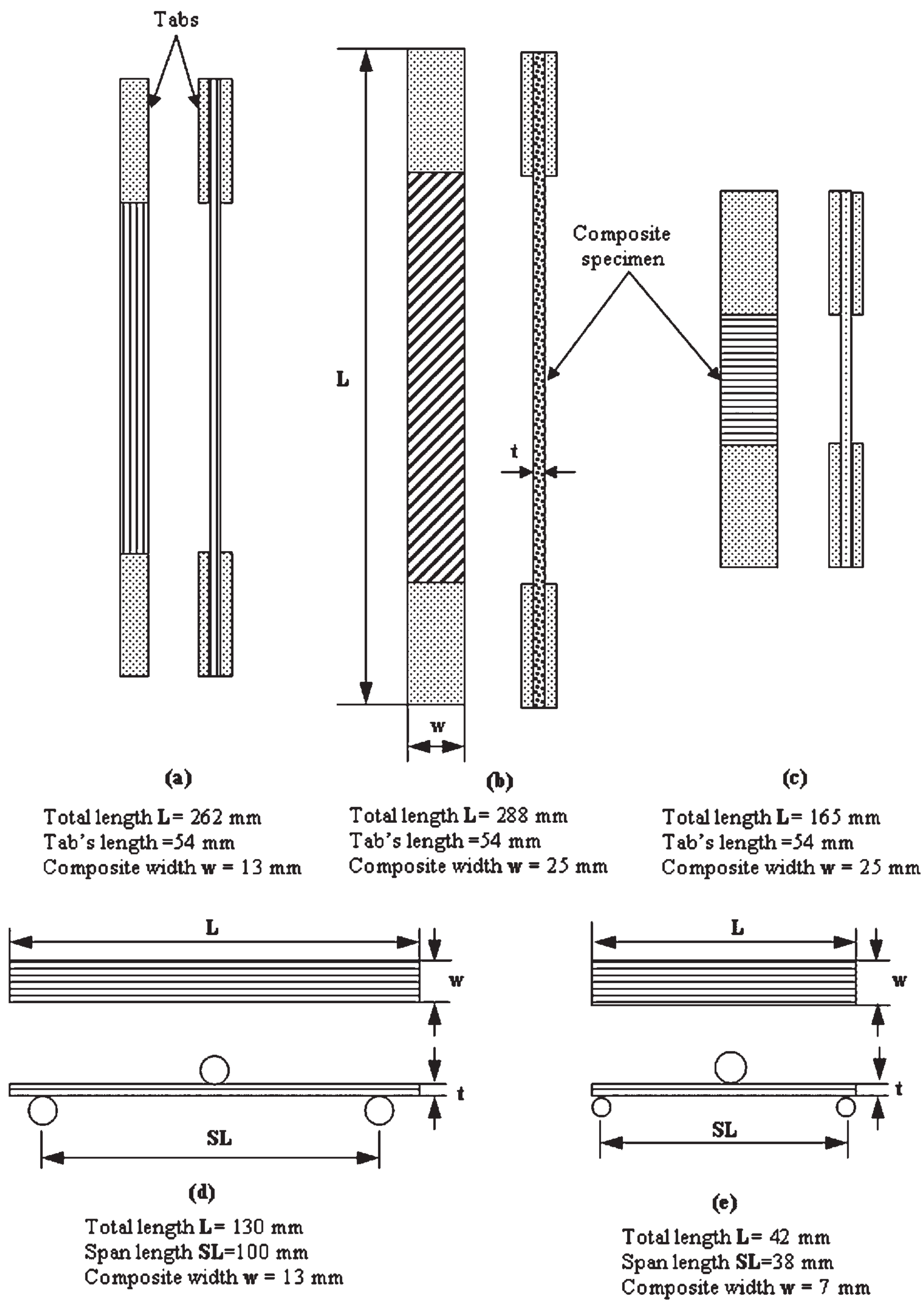

Figure 2 Schematic diagrams of the testing geometries: (a) longitudinal, (b) off-axis, and (c) transverse tension; (d) flexure; (e) short beam shear.

extensometer DVE-101/201), with Trapezium (advanced software for materials testing) for machine control and data acquisition. Four rectangular tabs to be used with the tensile specimens were produced from carbon fiber reinforced epoxy with the lay-up $[0/90]_s$ and were bonded to the griping length of each test specimen using a cold hardening epoxy resin. Tensile tests were performed at a constant cross-head speed of 2 mm/min at room temperature under air. Six tests were made for each material per orientation.

Flexure test

One of the most common experiments to characterize materials in flexural conditions is the three-point bending test. Specimens were machined from flat panels with a high-speed diamond saw with a 50/50 mix of water and ethylene glycol coolant. This machining operation resulted in constant width specimens having very smooth cuts. The gauge lengths for the three-point bend test were determined according to ASTM D 790 standard, and were set to 100 mm for all of the composite specimens. Flexural strength is derived from the simple beam theory as follows:

$$\sigma_f = \frac{3PL}{2wt^2} \tag{1}$$

where σ_f denotes the flexural strength, P the applied load that leads the specimen to rupture, L the support span, w the specimen width, and t specimen thickness.

For the case where the three-point flexure specimen is not strain-gauged, flexural modulus may be determined from the slope of the initial straight-line part of the load–deflection curve by means of this equation:

$$E_f = \frac{L^3}{4wt^3}\frac{\Delta P}{\Delta \delta} \tag{2}$$

where $\Delta P/\Delta\delta$ represents the slope of the force–displacement curve, and E_f the flexural modulus.

Short beam shear test

To determine the interlaminar shear strength of the composites, short beam shear tests were performed following ASTM 2344 ''Apparent interlaminar shear strength of parallel fiber composites by short-beam method.'' A sliding roller three-point bending fixture, which included a loading pin (diameter, 6.4 mm) and two support pins (diameter, 3.2 mm), was used for the room temperature short beam shear tests. The test fixture was mounted in a 5-kN capacity, screw-driven load frame. The apparent interlaminar shear strength of composites was determined from samples that were tested with a support span/sample thickness ratio of 5 : 1. The simply supported specimens allow lateral motion and a line load is applied at the mid span of the specimens. The apparent shear strength was then calculated as follows:

$$V = 0.75\left(\frac{P_{\max}}{wt}\right) \tag{3}$$

where V is the apparent shear strength, $P_{\max}$ is the failure load, and w and t are the width and thickness of the specimen respectively. Further details of the test procedures are given in the ASTM 2344.

Scanning electron microscopy observation

The fracture surfaces of tensile-tested specimens were examined using the SEM (JEOL JSM 6060) at excitation voltage equal to 20 keV in the secondary electron mode. To reduce the extent of sample arcing, the samples were coated with a thin layer of metallic gold in

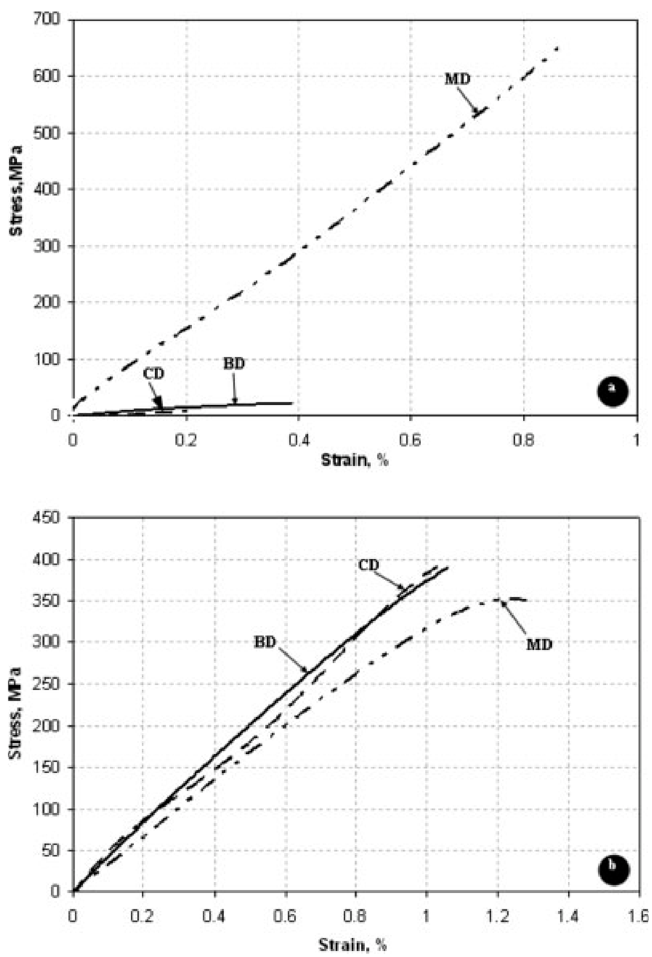

Figure 3 Tensile stress–strain curves of laminates in different directions: (a) unidirectional carbon fabric and (b) quasi-isotropic +45/−45/90/0° carbon fabric laminate.

an automatic sputter coater (Polaran SC7620) prior to examination by SEM.

RESULTS AND DISCUSSION

Experimental results and discussions on the mechanical properties

Tensile properties

Comparisons of the values of tensile strength determined experimentally with the tensile specimen cut at different angles from the same UD glass/polyester laminate are shown in Table III. These results for a $[0]_{12}$ UD carbon/polyester composite, given in Figure 3(a), show that the strength falls rapidly when the fiber orientation is not parallel to tensile loading direction. The variation in stiffness in a highly anisotropic UD carbon fabric/polyester and the decrease in stiffness over that of the laminate, even at right angles to the lay of the fibers, are shown in Table III.

Laboratory tests have long established that the laminate ply stacking sequence and the ply fiber orientations greatly influence the onset and growth of free edge delamination.[20,21] Under tensile loading, the delamination in most laminates, especially those containing 90° plies, is preceded by a number of transverse cracks.[22] The stacking sequence of $[0°]_{12}$, where there is no an abrupt change in the various laminating plies, produces an interlaminar normal stress at the free edge under applied uniaxial tension. Because of the absence of transverse cracks in the $[0°]_{12}$ specimen of carbon/polyester under tension in MD, a delamina-

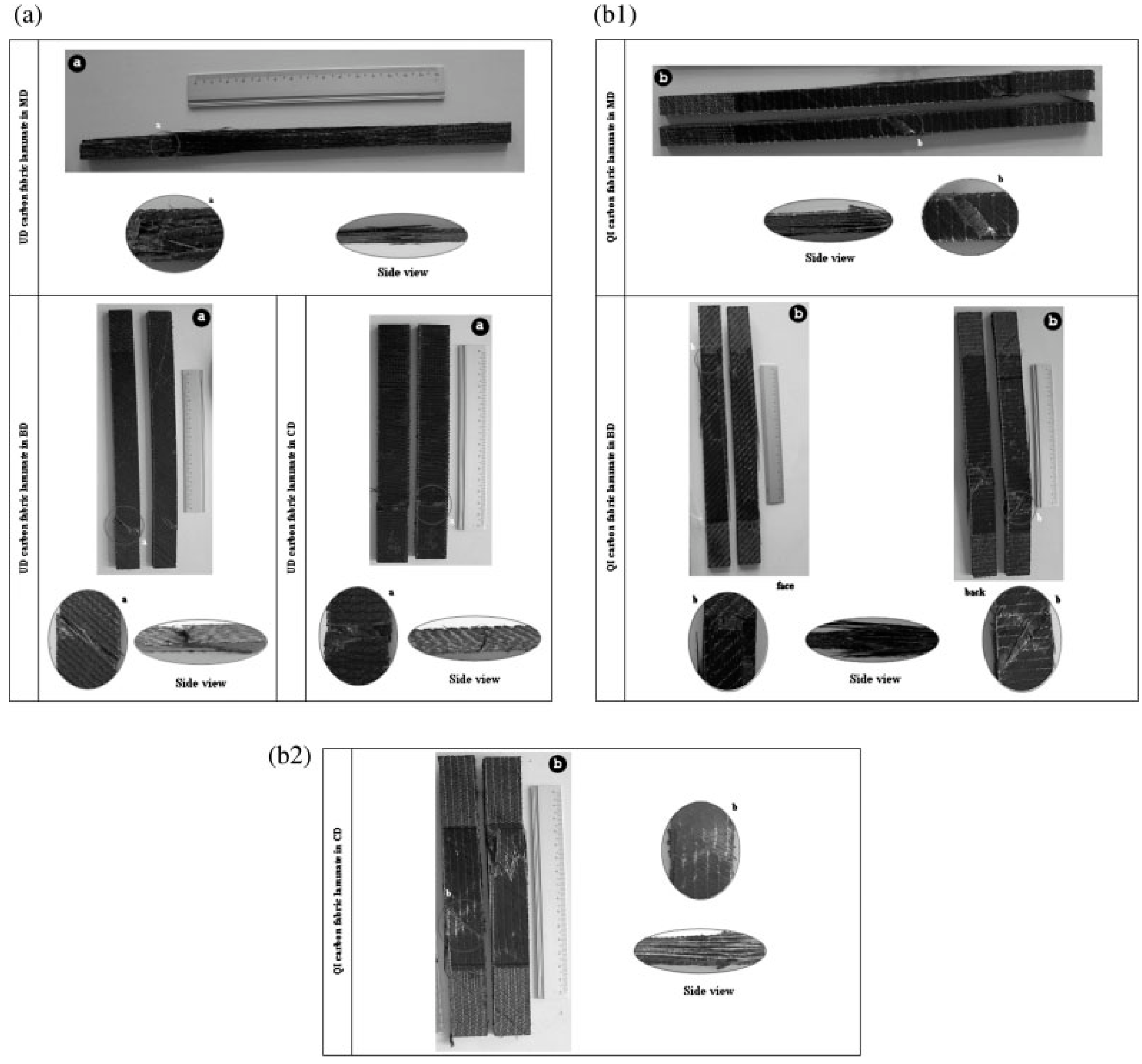

Figure 4 Photomicrographs showing side and front views of representative failures loaded in tension along the MD, BD, and CD, for laminates produced from (a) unidirectional carbon fabric and (b) quasi-isotropic carbon fabric laminate.

tion crack cannot propagate in the interface of the laminating plies. Consequently, this specimen did not show any evidence of delamination at an applied axial tension of 641.1 MPa. Figure 4(a) is the photomicrograph of fractured tensile specimen with intermediate bond strength. This fracture surface of UD composite is irregular and has some fiber pullout.

In the case of UD laminate of carbon/polyester subjected to transverse tensile loading, a number of longitudinal fractures of matrix and debonding of fiber–matrix interface were formed in the laminating plies [Fig. 4(a)]. Note that each of the fracture planes was, macroscopically speaking, parallel to the fibers and normal to the applied tension. Each longitudinal fracture is formed and arrested at the 0°/0° interfaces wherein a series of localized longitudinal cracks can be developed as the applied load advances to higher levels; and consequently, carbon–polyester laminate subjected to transverse tension failed without exhibiting delamination.

It is interesting to note that the strength of quasi-isotropic laminate in test directions performed on the composites are not equal [Fig. 3(b)]. A somewhat surprising feature is that the strength in BD and CD is about 10% higher than that of MD, whereas the stiffness differs merely by a few percent (Table III). Improvements in the tensile strength in BD and CD cases have been attributed to the combination of fiber orientation and the orientation of the stitch rows, since the tricot knitting pattern runs nearly parallel to these directions as a consequence of the lay-up of 12 BA carbon fabric with [+45/−45/90/0] orientations (Fig. 5). It is worth noting that for MD samples there is not stitching in either + 45/−45 or 0/90 layers which goes along the loading direction (MD). Experimental data taken from the literature imply that the orientation of the stitch rows with respect to the loading axis greatly influences the tensile properties of noncrimp stitched carbon fabric composites.[23,24] Both studies report that when the stitch rows were aligned parallel to the direction of tension loading, the strength was improved. Also, the anisotropy of strength for quasi-isotropic laminates is believed to be because the nominal 45° direction of the fibers was not controlled during the laminate fabrication by VARTM process.

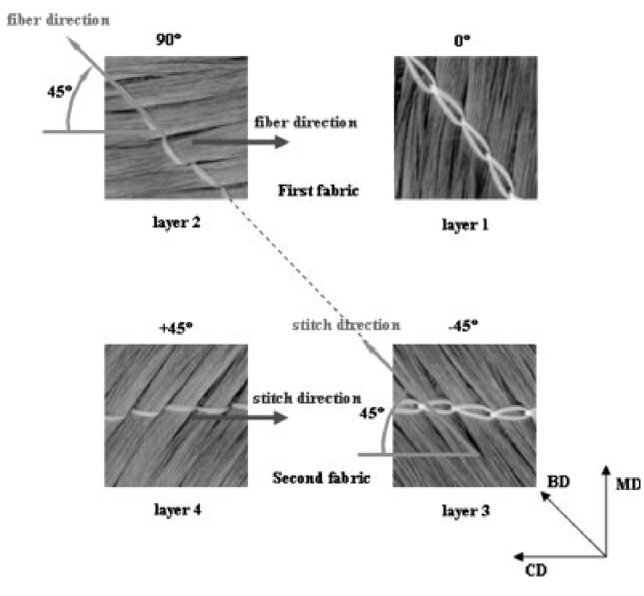

Figure 5 The effect of the orientation of stitch rows and fiber orientation on tensile strength behavior with respect to bias/cross directions.

Another interesting point is demonstrated by the photomicrographs showing side and face views of quasi-isotropic carbon fabric laminates after failure in the test along MD, BD, and CD [Fig. 4(b)]. The implication is that matrix cracks form longer cracks oriented both in + 45/−45° relative to loading direction. Most of them run from one edge of the specimen to the other. Despite the fact that delaminations were relatively localized at each side of the fracture site in all test directions, the appearance of failed specimens in the side plane tested in the BD is similar to that of the MD and CD tested coupons but with more debris. Comparisons between the actual photomicrograph of fractured tensile specimens show that damage at fracture, in the case of UD fabric laminate, was relatively localized when compared with the quasi-isotropic laminate and that the UD fabric laminate retained its through-thickness integrity.

Flexural properties

It is of interest to note here that a catastrophic failure was observed neither in the UD nor in the quasi-isotropic carbon fabric laminates. From a closer observation on the tested specimens, it can be revealed that there was some visible interplane slippage and crack initiation in the midplane because the through-the-thickness stitching increased delamination resistance of laminates. Hence, delaminations did not grow as readily in the stitched laminate.

It is of interest to note here that the recorded values of the flexural strength of both laminates, except for the UD laminates in the MD, were significantly lower than those obtained from tensile tests at all the considered test directions (Table III). When the fractured bent specimens were investigated, it was observed that the upper layers of UD carbon fabric specimens fractured in fiber dominated MD direction as a result of compressive stresses. On the other hand, matrix dominated BD and CD directions had a fracture at the lower layers having tensile stresses. For quasi-isotropic laminate, the layers at the upper side had damaged where the loading in MD, BD, and CD directions resulted in compressive stresses. It is well known that carbon fibers have higher strength for tensile loading than compressive while polyester shown better strength properties in compressive directions than tensile. The specimens under pure tensile load had higher strength than did the specimens subjected to bending which showed compressive failure. The specimens tested for bending and failed in tensile side had higher strength than did the specimens tested in pure tension.

In the case of UD carbon fabric laminates, the excitation aspects of the strength–deflection response were confirmed to agree well with the shape of the stress–strain curves in tensile tests [Fig. 6(a)]. Table III demonstrates that the composite strength falls rapidly with increasing fiber orientation angle (θ). It has already been mentioned in literature that when the angle θ is high, the strength of the composite tends to be matrix- or interface-dominated.[25] This statement is true not only of UD composites but also of any lamina in a multidirectional laminate. It can be seen from Figure 6(a) that the magnitude of the deflection is quite small for the CD case. This could be because when the load is directed toward the laminate the matrix-dominated specimen is a very feeble structure and only local inward deformation causes the specimen to bulge out at the bottom surface layer.

By examining the strength–displacement curves of the quasi-isotropic laminates [Fig. 6(b)] it is clear that these distributions are quite different. The great improvement in middle-span deflection for the BD case when compared with that of the MD and CD is seen to occur at a load of about 150 N. From an overall point of view, however, we feel that increasing the displacement in BD is not perhaps as important as reducing the strength intensities in BD. The effect of stitching on the flexural properties has been studied by numerous investigators and discussed critically by Mouritz et al.[16] In a study of the stress concentration effects of modified lock Kevlar® stitches in GFRP composites under four-point bending, Mouritz[26] observed that damage was localized around the stitch loops, which clearly revealed that the bending stresses were highest around the thread. By the same reason the flexural strength was reduced about 15%. It was concluded that not only was the stitch density controlled by selecting the stitch spacing, but the orientation of the stitch rows affected the flexural properties of stitched noncrimp fabric composites.

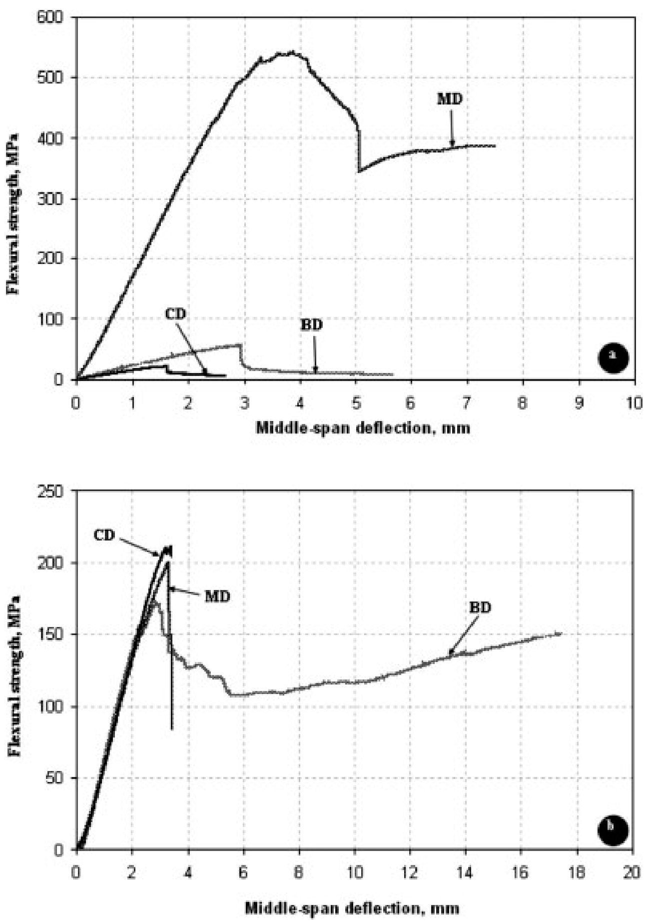

Figure 6 Flexural strength–deflection curves of laminates in different directions: (a) unidirectional carbon fabric and (b) quasi-isotropic carbon fabric laminate.

Short beam shear test results

It is not surprising to find that the interlaminar shear strengths at the time of failure recorded are at least an order smaller than the flexural strengths of both laminates (Table III).

Comparison of the results with data in Table III suggests that carbon composites fabricated from UD fabric are more susceptible to interlaminar shear strength degradation as a function of test directions than composites produced with the fibers in a quasi-isotropic lay-up by using suitably arranged BA carbon fabrics. It has been established, both in theory and experiment, that the lamina stacking sequence influences interlaminar stresses, and consequently, delamination in laminates. An approximate elasticity solution has been used by Pipes and Pagano[27] to study the effect of the stacking sequence on the interlaminar shear stress in $+45/-45$ laminates. The results of their study indicated that stacking sequence had a significant effect on the interlaminar shear stress and that these stress concentrations (when layers having the $[45/45/-45/-45]_s$ orientation) could be reduced by use of the layers having the opposite $[45/-45/45/-45]_s$ orientation. Experimental results, in conjunction with analytical models, showed that the distribution and magnitude of interlaminar normal stress and interlaminar shear stress are widely varying because of the stacking sequences and laminate type.[28,29] Multiply composites containing cross-ply layers of parallel filaments also experience complex forms of shear failure, including delamination between layers, cracking parallel to the fibers of one layer which eventually causes fracture of neighboring crossed layers, and combinations of these two events. Observation of these modes of failure suggests that often crack extension parallel to the fibers in one layer precedes and causes the other modes of failure.[30]

It can be observed, from Table III, that specimens made from quasi-isotropic $[+45/-45/90/0]_3$ lay-ups of carbon fiber reinforced polyester, in MD case, exhibited the lowest loading capacity with an ultimate strength of 5.2 MPa, which is 61.8% and 42% lower when compared with the 8.5-MPa and 7.4-MPa loading capacity of the specimens in BD and CD cases, respectively. A lay-up of 12 BA carbon fabrics with $[+45/-45/90/0]_3$ orientation, combined with the orientation of the stitch rows and fiber orientation, is held responsible for the different behavior with

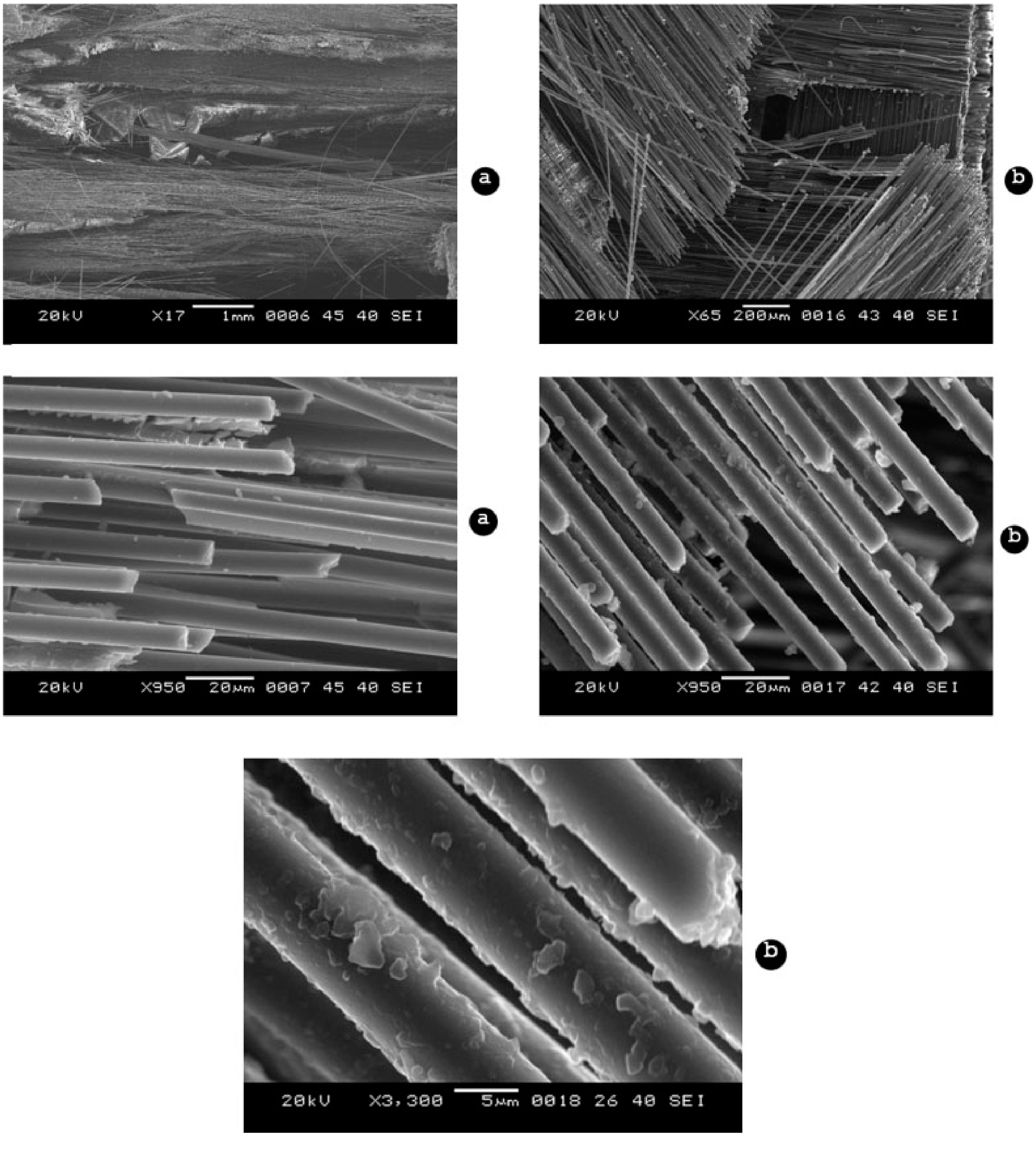

Figure 7 Representative SEM images of the breakage region for (a) unidirectional carbon fabric and (b) quasi-isotropic carbon fabric laminate that were tensile tested in machine direction (at different magnification levels).

respect to the BD/CDs (Fig. 5). The effect of stitching on interlaminar shear strength has been studied by Du et al.[31] for a prepreg CFRP tape laminate, and improvements in the shear strength due to stitching were recorded to be 14.4 or 25.6% depending on the orientation of the stitch rows.

Fractographic analysis

It is possible to observe clear local differences in the representative images of the breakage region for UD carbon fabric specimens [Fig. 7(a)]. In the upper inset a general view is presented showing ductile deformation of the matrix but a poor matrix–carbon fiber adhesion. In the lower photomicrograph, a higher magnification image clearly shows a clean fiber surface, although a small amount of polymer can be seen between fibers. These observations suggest an adhesive failure in the interface after fiber fracture. Although the UD carbon sample shows adhesive failure [Fig. 7(a)], its relatively high tensile strength can be explained taking into consideration a highly ordered carbon structure along the fiber axis, while chemical breakage takes place. Pullout effect of fibers was observed from these images.

In Figure 7(b) the 11th and 12th reinforcement layers are shown, which are (+45/−45) and (0/90) orientated, respectively. Although none of the 45° tows are fully fractured, there is evidence of considerable damage to one tow on the right hand side micrographs, corresponding to the side of the coupon that had the visible surface damage. None of the 45°and 90° tows are fractured and there is little or no evidence of individual fiber fracture. It should be noted that on the micrographs the exposed reinforcements are not ''clean''; there are flaky deposits on the reinforcement surface, which are remnants of the resin that was not fully decomposed.

CONCLUSIONS

We would like to draw attention to the fact that some conclusions of this study are of general applicability, while some others refer to the special problem discussed. So it has been pointed out that when the fiber orientation angle in UD composites is high, the strength of the composite tends to be matrix- or interface-dominated. Having recognized the nature of this problem, it was noted that a continuous decrease in strength of multidirectional composites can be achieved by deviating the test angles θ from 0°. This behavior is quite general. On the other hand, it has been pointed out that for quasi-isotropic laminate the experimental data do not allow for an adequate description of the quasi-isotropic model. Improvements in tensile strength in BD and CD cases are believed to be due to the fact that the combination of fiber orientation and the stitch rows run nearly parallel to these directions as a consequence of the lay-up of BA carbon NCF blankets with [+45/−45/90/0] orientations.

Perhaps the most significant result of this investigation is the strong correlation between the changes in interlaminar shear strength values and fiber orientation angle, in the case of UD carbon fabric laminates. In addition, the lamina stacking sequence and laminate type have a significant effect on the recorded values of the interlaminar shear strengths.

The interlaminar shear strengths at the time of failure recorded were significantly lower than those obtained from three-point bending tests at all the considered test direction.

Photomicrographs of fracture surfaces of multiaxial composites loaded in tension along the fiber direction exhibit pronounced irregularity and fiber pullout.

Composites broken in tension and observed under different moderate magnification levels will show obvious differences between good bonding rarely and poor bonding commonly at the interface. A chemically noncoupled composite showed uncoated clean fiber ends pulled from the polymer. Even the highest magnification could not show the presence of a monolayer of polyester resin on fiber.

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