Title: Hierarchical Carbon Nanotube Carbon Fiber Unidirectional Composites with Preserved Tensile and Interfacial Properties

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Abstract: Hybrid hierarchical carbon-nanotube (CNT)-based composites, such as radially-aligned CNT arrays grown onto microfiber filaments, have significant potential to expand the performance and functionality of fiber reinforced composites. Here, a novel method for high-yield growth of aligned CNTs on aerospace-grade carbon fibers (CFs) is demonstrated at the composite level for the first time. Fuzzy carbon fiber reinforced plastics (fuzzy CFRP) unidirectional composites with >60% microfiber volume fraction are fabricated via vacuum-assisted resin infusion of CNT-grafted tows using an unmodified aerospace-grade epoxy. Preservation of microfiber tensile modulus and strength are demonstrated by longitudinal composite tensile testing, consistent with single-fiber tensile tests. Fiber-matrix interface strength is also unchanged by the CNT growth as revealed through continuously-monitored fiber fragmentation tests. Taken together, the results provide needed new composite-level understanding of hierarchical structural composite laminates and motivate future work on structural CF composite laminates with integrated multifunctionality and improved interlaminar and intralaminar performance.

Keywords: Carbon Fibres (A.), Carbon nanotubes (A.), Nano composites (A.), Mechanical Properties (B.) Hierarchical Composites

1. Introduction:

Hierarchical nano-engineered composites have become of great interest in order to leverage superior properties of constituents such as carbon nanotubes, and thus integrate mechanical improvements into existing composite structural materials, e.g., in aerospace structural applications of advanced composites. Considerable work with various nano-architectures have been explored [1, 2], however, one that is particularly advantageous is the “fuzzy” fiber reinforced plastic architecture (FFRP) in which all micro-fibers are coated with radially aligned carbon nanotubes (CNTs) prior to matrix impregnation [3, 4]. This hierarchical fiber architecture (see Fig. 1a) is attractive for its ability to tailor CNT distribution within the matrix without running into dispersion issues commonly encountered through shear-mixing of CNT within resin systems prior to infiltration [5]. More importantly, prior work with FFRP implemented on alumina/epoxy composites have demonstrated a wide array of composite property improvements including increases in mode I fracture toughness, interlaminar shear strength, bolt-bearing strength, thermal conductivity, and electrical conductivity among others [3, 4, 6].

Despite large interest of applying this architecture onto CF composites particularly for aerospace structures, many barriers have compromised attainable improvements. Due to the poor wettability of fibers within catalyst solutions (Ni, Co, Iron) typically used for CNT growth via chemical vapor deposition (CVD), harmful fiber surface modifications techniques were employed to promote catalyst adhesion [7, 8]. Common metal catalysts also dissolve fiber surfaces at high temperatures (~ 700 °C) typically used in chemical vapor

deposition for CNT Growth [7, 8]. Consequently, although prior work has attained circumferential CNT growth on CF, single fiber tensile strength reductions as high as 55% were observed. Alternative strategies employing sol-gel “barrier coatings” that shield catalyst from the fiber surface have also demonstrated single fiber strength losses of 53%, thereby suggesting CVD processing also contributes to CF damage. Even heat treatment of CFs in inert atmospheres has shown tensile strength loss that is time and temperature dependent above a critical strength loss temperature [9]. Thus a large tradeoff was observed between high-yield CNT growth and CF tensile properties, which is critical to in-plane properties of the composite.

In order to obtain high-yield circumferential CNT growth without degrading fiber properties, it is crucial to enable catalyst adhesion without damaging fiber exteriors since strength is highly susceptible to surface defects [10, 11]. As a result, we utilize our recently reported polymeric functional coating Poly(styrene-alt-[dipotassium maleate]) (K-PSMA) that adsorbs onto the CF surface and leaves surface potassium ions available for ion exchange with iron upon immersion in a catalyst solution. Subsequently, catalysts adhere to the functional coating without being in direct contact with the CF, such that surface dissolution is prevented. Further, a low temperature (below a previously characterized threshold strength loss temperature) thermal CVD process for CNT growth was implemented as outlined by Steiner III et al. based on the process introduced by Magrez et al. [12]. Such processing has been shown to retain CF tensile properties at the single fiber level. Leveraging this strategy for non-covalent functionalization and low temperature CVD [9], this work brings to fruition, for the first time, a fuzzy CFRP composite without degraded in-plane properties. Unidirectional fuzzy CFRP composites were fabricated and

tested for longitudinal tensile properties at high fiber volume fractions (>60%) as is utilized in high-performance aerospace structural applications. Further, to better assess the effects of CNTs on load transfer between the fiber and matrix, continuously monitored single fiber composite articles were fabricated and fragmentation responses were used to characterize interfacial properties within the fuzzy CFRP.

2. Method:

2.1. Non-covalent Functionalization and Low Temperature CNT Growth

An aqueous solution of 0.25 wt % K-PSMA was prepared as follows: 1.4 grams of Polystyrene-alt-maleic anhydride (Sigma-Aldrich, 99%, Molecular weight of 350,000) was dissolved in 25 mL of acetone and slowly mixed with 300 mL of 0.3 M aqueous sodium hydroxide before stirring for 3 hours. 0.1 M nitric acid was then added to bring the pH to 8 and acetone was removed using a rotary evaporator to produce a 0.5 wt % h-PSMA solution similar to the method used by Stroock et al. [13, 14]. Potassium carbonate (K_2CO_3 Sigma-Aldrich part number 209619, $\geq 99.0\%$) was added to the h-PSMA until the pH was 11, and an equivalent mass of water was added to the solution to yield 0.25 wt % K-PSMA.

Unsize CFs, TohoTenax HTR40 24K (research variant of HTA40 F22 24K with 3.5 GPa strength and 204 GPa modulus [9]), were selected for this study since these fibers have previously been evaluated for tensile property loss on a single fiber level. These CFs do not require any desizing procedure that may introduce damage to the fibers. 30 mm-long tow segments were extracted from the spool using light tension and adhesive tapes on either end to maintain fiber collimation before being rinsed in deionized water and dried in a vacuum heated chamber at $\sim 80^\circ\text{C}$ for 5 minutes. Tow segments were then dip-coated in 0.25 wt % K-PSMA solution for 5 minutes, spread over smooth rollers to remove excess

liquid, and dried in a vacuum heating chamber at $\sim 80^\circ\text{C}$ and pressure of -91 kPa for 5 minutes under light tension. A catalyst solution of $0.05\text{ M Fe(NO}_3)_3 \cdot 9\text{H}_2\text{O}$ in isopropanol was mixed and stirred for 1 hour to develop iron oxide particles in solution. The K-PSMA functionalized tow was then dip-coated in the catalyst for 5 minutes, before undergoing controlled drying in a vacuum heating chamber at $\sim 80^\circ\text{C}$ and -91 kPa for 5 minutes.

Fe^{3+} /K-PSMA coated tows were inserted into a fused quartz tube (25 mm outer diameter, 22 mm inner diameter, 30 cm length) enclosed within a Lindberg/Blue M Minimate furnace. 750 sccm of Ar was first flushed into the tube for 2 min before flowing H_2 and Ar at 400 and 200 sccm, respectively. Iron catalysts were then reduced during temperature ramp (50°C/min) to 480°C before growth was started by switching off H_2/Ar , and flowing $\text{CO}_2/10\%\text{ C}_2\text{H}_2$ in Ar at 17 sccm / 167 sccm. Lastly, the gasses and furnace were turned off, and the tube was flushed with 750 sccm of Ar during cooling.

2.2. Unidirectional Fuzzy CFRP Fabrication

To directly assess in-plane properties of hierarchical composites composed of CNT coated CFs, a fabrication method for unidirectional fuzzy CFRP specimens and baseline CFRP (without nanotubes) was necessary. Utilizing the same unsized CFs that had been characterized in-depth from prior work, we created resin-impregnated tow specimens in accordance with the widely reported ASTM D4018 standard [15]. A resin impregnation mold and vacuum assisted resin infusion (VARI) system [16] utilizing low viscosity aerospace grade epoxy resin RTM6 (2.89 GPa tensile modulus, 75 MPa tensile strength, 33 mPa.s initial viscosity at 120°C) was chosen [17]. Dry tow segments (both baseline and fuzzy) were first lightly tensioned and inserted into the groove half of a set of Teflon molds. Tension was still applied to the tow segment to preserve fiber collimation while the

Teflon plug was controllably lowered into the groove. Next, the mold and tow assembly was wrapped inside vacuum bagging and connected to a resin source and vacuum on opposite ends before being heated to 90 °C to lower resin viscosity. Once the target temperature was reached, valves were opened to allow resin infusion through the full length of the mold channel. Valves were then closed, and temperature was set to 160 °C for 75 min and 180 °C for 120 min to cure the resin following the manufacturer's recommendation. After cooling, mold halves were separated and the resin-impregnated tows were gently pulled out of the mold. Since microfiber (CF in this study) volume fractions (V_f) for aerospace applications range from 40-50% in woven materials to ~60% for unidirectional materials, specimen V_f was targeted to be at least 60% to be directly relevant to even demanding aerospace structural applications. Both baseline and fuzzy CFRP specimens were manufactured and tested at CF $V_f = 67-68\%$. To prevent tow crushing and failure at the mechanical testing grips, specimen ends were embedded in a uniform layer of thermoset epoxy that is sandwiched between laser-cut sandpaper and cured on molds to ensure consistent 150 mm sample gauge lengths specified by ASTM D 4018.

2.3. Longitudinal Tensile and Interface Shear Strength Testing

Tensile testing was conducted on the Instron 1332 load frame with load measured through a 220 kN capacity load cell. Gripping pressure was set to 2.1 MPa onto each specimen tab to prevent sample pull-out as was determined from preliminary testing. Samples were aligned with the loading axes to ensure a symmetric and uniaxial stress state in the center of the specimen. In accordance with the ASTM D 4018, the displacement-controlled test was conducted at a quasi-static displacement rate of 1 mm/min until specimen fracture. Strain measurements were obtained through optical strain mapping as a

non-contact method for thin (1.15 mm-wide) specimens. Prior to mounting, specimens were sprayed with a white base paint and speckled with black ink. During testing, a high-resolution camera (Point Grey Grasshopper 2), placed 1 m away from the testing plane, captured optical images of the speckled sample in 250 ms intervals. Images were synced with load data and imported into Correlated Solutions Vic2D software, which calculated strain from speckle displacements within a selected area of interest (0.5 mm X 12 mm).

Synthesis of fuzzy CFs involves non-covalent functionalization of filament surface as part of the approach [9] to retaining fiber strength and stiffness. Microfiber-matrix adhesion needs to be established before such fuzzy CFRP properties can be predicted and understood. Fuzzy and as-received fibers were used to make single fiber composites for continuously-monitored single-filament composite (CM-SFC) fragmentation tests as discussed in previous work [18]. To ensure representative results, individual fibers were extracted from fuzzy CF tows (and baseline tows) using tweezers and adhesive tape. The individual fibers were then embedded in EP-502 from Polymer Gvulot, a bisphenol-A (DGEBA) based liquid epoxy resin, and mixed with the curing agent EPC-9 triethylenetetramine (TETA). A single fiber composite specimen, with typical cross-sectional dimensions of 3.1 X 0.15 mm² and neat matrix Young's modulus of 1.34 GPa, was prepared and tested as in previous work [18]: CM-SFC tests, at least four of each specimen type, were performed in a computer-controlled MiniMat2000 tensile tester while monitored under a microscope with a video CCD camera. Samples were held at a gauge length of 12 mm and deformed at 50 µm/min, while digital stress and strain data were recorded with real-time color video monitoring, where polarized light is used to identify fiber break sites by the birefringent gaps created under load. Fiber-breaking is then

correlated with the fiber stress, calculated from the measured stress and CF/matrix Young's moduli ratio. The CF modulus is known from the single fiber tensile testing in air.

3. Results and discussion

3.1. Processing and Manufacturing Fuzzy CFRP

Scanning Electron Microscopy images (JEOL 6010 LA) of unsized CFs subjected to 0.25 wt % K-PSMA / iron nitrate dip-coating and CVD processing (see Fig. 1b) indicate high-yield radial CNT growth around CFs. Growth in between fibers (see Fig. 2a and 2b) and within large area tows (see Fig. 2c) demonstrate an improvement in conformal CNT coating around fibers due to a decrease in K-PSMA concentration and resultant coating thickness as compared with prior work. Moreover, this facile incipient wetness method is advantageous in process-scaling. Transmission electron microscopy reveals that the nanotubes are multi-walled and have diameters of ~25 nm [9]. CNT lengths range from 1 to 2 microns, and growth coverage is consistent across different samples which we attribute primarily to the pressure and temperature control during the drying processes. As interfiber distances are less than 0.5 microns for a 67% V_f unidirectional composite (hexagonal close packing serves as an upper bound), CNT lengths are sufficiently long to span across the polymer matrix in between fibers. Cured composite specimens were cross-sectioned and examined via optical microscopy and microCT, revealing that void-free specimens were achieved by utilizing resin infusion through the close-mold manufacturing (see Fig. 2d) that is typical of aerospace composites. This processing is consistent with past alumina FFRP results that show the ability of CNTs to facilitate capillary-driven wetting of fibers with resin.

3.2. Unidirectional Composite Testing

A typical stress strain plot (see Fig. 3 left) of a unidirectional composite tensile test specimen (see photograph in Fig. 3 right) displays consistency with previously acquired single fiber testing data for HTR40 CFs [9]. Given that longitudinal composite properties are fiber dominated, strain to failure at $\sim 1.7\%$ is within the range of that acquired from single fiber data. Tangent modulus of the composite specimen also increases with strain, which is expected from the straightening of ribbon-like graphitic crystallites for PAN-based CFs under tension [10]. Composite chord modulus values were calculated between the strain limits of 1000-6000 microstrain as specified in the ASTM standard.

Moreover, composite chord modulus values (see Fig. 4 left) are in close agreement with micromechanical predictions using previously acquired single fiber testing results. From the isostrain conditions of slender fibers and matrices, a rule of mixtures (ROM) prediction can be utilized to estimate a resultant baseline composite chord modulus [19]. Single fiber tensile testing data of fuzzy CFs (20 fibers tested in accordance with D3379 in the Nano UTM) was revisited to calculate fiber chord modulus values [20] (between 1000 and 6000 microstrains), which were used with RTM6 literature values for predictions [17].

No significant change in chord composite modulus was observed for fuzzy CFRP compared to baseline. This result is also expected from multi-scale ROM analysis [21]: matrix regions surrounding each CF have radially aligned nanotubes and can locally be modeled as aligned polymer nanocomposites (A-PNCs). Since nanotubes are oriented perpendicular to fiber axis, the longitudinal modulus of the matrix in the fuzzy CFRP case can be estimated by the transverse modulus of A-PNCs, which has been well characterized utilizing different techniques [22]. Even for high CNT volume fractions, the transverse A-

PNC modulus is roughly the same as neat epoxy values. Application of ROM yields a 137 GPa chord modulus prediction that is nearly centered within the 138 ± 6.5 GPa range from fuzzy CFRP testing results. Thus longitudinal elastic properties of the fuzzy CFRP are preserved.

To directly compare fiber performance within the context of a composite, implied fiber chord moduli were calculated per the ASTM Standard, which does not stipulate a specific fiber volume fraction of specimens. Since the polymer matrix has a stiffness two orders of magnitude lower than that of carbon filaments, it can be assumed to have a much smaller axial load-bearing contribution. As a result, implied fiber stresses can be calculated by dividing loads by the cross-sectional area of the fibers via $(E_{c_implied} = (P_{upper} - P_{lower}) / (n\pi r_f^2 (\epsilon_{upper} - \epsilon_{lower})))$ for implied fiber chord modulus. Table 2 demonstrates no significant difference in implied fiber chord modulus between baseline and fuzzy CFs. Additionally, implied fiber chord moduli are in close agreement with single fiber testing results, thus corroborating the consistency of our testing methodology across different scales.

More importantly, the aforementioned agreements are also seen for both composite strengths and implied fiber strengths as plotted in Figure 4 (right) and shown in Table 2. Implied fiber strength means are close to the average strengths of single carbon filaments, and well within the larger scatter that is typical for single filament testing. The centers of the implied strength ranges for both baseline and fuzzy fibers are in line with those of the corresponding single filament strength ranges. The lower variance of impregnated tow testing is expected due to effective matrix load transfer around fiber ruptures to adjacent filaments [19], a fact that renders the testing methodology suitable for fiber production

quality control assessments [23]. A critical result is observed here: fuzzy CFRP retain the same strength as baseline CFRP composites at the tested fiber volume fraction. Further, the lack of degradation in implied fibers strengths is consistent with the preserved strengths previously observed of single fuzzy carbon filaments [9].

3.3. Fuzzy CFRP Fiber-Matrix Interfacial Strength

The interface of advanced fiber composites is highly engineered, typically via application of a ‘sizing’ polymer to passivate the CF after production, protect it from additional processing, and allow a carefully controlled interface to be formed with the polymer matrix when the CF is finally processed into a composite. As discussed in [24], the interface is not engineered to be as strong as possible for such composites, primarily for damage tolerance reasons. It is therefore of interest to investigate whether the fiber-matrix interface is altered significantly by the CNTs grown on the surface [25]. The CM-SFC is a favorable testing method for this purpose, since a comparably small amount of test specimens enables the measurements of both Weibull shape and scale parameters of a fiber embedded in matrix. A typical progression of a CM-SFC test is shown in Fig. 5a-c. After the initial unloaded state (Fig. 5a), increased loading causes sparse independent fiber breaks (Fig. 5b), and finally individual fragments interact after reaching saturation. (Fig. 5c). The results, namely the embedded fiber critical length, and its strength at that critical length, can then be used to calculate the fiber-matrix interfacial shear strength (IFSS) using the Cottrell-Kelly-Tyson theory [18, 26, 27]. It should be noted that Weibull values from the CM-SFC tests are usually different than Weibull parameters from single-fiber tests, due to the well-known increase in strength and reduced variability that occur for fibers embedded in compliant matrices [18], and are presented in Table 1 for completeness.

The continuous single fiber fragmentation results reveal that there is no change between the IFSS of the baseline and CNT-grown fuzzy CFs. Fiber fracture in the baseline and fuzzy CF system is shown to be similar (see Fig. 5d). The ln-ln plots of the CM-SFC tests results are presented in Figure 5e, and the data summarized in Table 1. The plots of baseline and fuzzy CF overlap each other, indicating that both the Weibull scale (indicating average fiber tensile strength, normalized by length) and shape parameters (indicating the variability around the scale parameter) of the fuzzy CF are within the standard deviation of the baseline CF characteristics. Note that although the shape parameter of the fuzzy CF appears to be smaller, suggesting higher variability, the difference is not statistically significant. The fragment saturation length, expressed as the lowest data point in each line on the ln-ln plots is also similar between baseline CF and fuzzy CF, showing that the fiber critical length – the smallest length enabling effective stress-transfer to the fiber (below which the stress transferred to the fiber falls below the fiber strength) – is also similar. This similarity, when combined with similarity in their tensile strength [18], shows that the IFSS is preserved after CNT-growth.

The unchanged IFSS and mode of failure (fuzzy CF pullout diameter is located at the fiber-matrix interface) is expected since (1) aligned CNTs are bound to the surface through weak non-covalent interactions so as not to damage/etch the CF, and (2) ~99% of the microfiber surface area is not covered by CNTs and is available for direct wetting and bonding with resin during infusion. Consequently, the interfacial properties of the fuzzy CF system are dominated by the carbon microfiber-epoxy bonding. Further, an increase in IFSS would typically indicate etching of the carbon microfiber surface, which is avoided to prevent microfiber strength loss. Note that the IFSS value of the fuzzy CF/epoxy is within

the range of commercially available intermediate modulus sized CFs in similar epoxy matrices (25-70MPa) [28-31]. Thus, as shown from Figure 5 and Table 1, longitudinal properties of the fibers are preserved with circumferentially grown CNTs, consistent with impregnated tow testing results. As the longitudinal and interface properties of the resultant hierarchical CF composite can now be considered equivalent to those of the baseline CF composite, it is expected that the fuzzy CF composites will have similar elastic properties. Fuzzy CF composites thus have the potential to enhance CFRP interlaminar and intralaminar strength and fracture properties (subject for future research), as demonstrated previously for alumina-fiber composites [4, 6]. Future work may also entail in-plane shear strength and transverse tensile strength testing as these property improvements are also expected based on previously observed CNT-reinforced matrix toughening and bolt-bearing strength and stiffness increases for alumina-fiber FRP composites [4].

4. Conclusions

Unidirectional fuzzy CFRP with high-yield radial CNT growth was developed and shown to have preserved tensile and polymer-CF interface strengths, two key attributes missing to date in fuzzy, sometimes called grafted, CFs. This development circumvents the drastic in-plane strength reductions, which have limited the full potential for hierarchical composite integration into structural applications. Interfacial properties of CFs and polymer matrix was observed to remain unchanged with the addition of radial CNT growth, which is advantageous for preserving the fiber load sharing characteristics. Since the IFSS is a key parameter that will govern composite failure, the similarity of interfacial characteristics between fuzzy and unsized CFs in epoxy serves as an advantageous base for further tailoring of the interface, such as the addition of traditional sizing onto fuzzy CF. Future

work will quantitatively assess transverse stresses as a function of CF volume fraction and CNT loading to assess the limitations on microfiber packing and CNT lengths before inducing fiber damage or exceeding composite tooling pressures similar to prior work conducted on alumina/epoxy FFRP [32]. Additionally, scale-up to laminate-level fuzzy-CFRP specimens will be performed to allow characterizations of interlaminar mechanical properties, electrical/thermal conductivity, and CNT-based multifunctionality.

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6. References

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7. Figure Captions

Figure 1 Fuzzy fiber reinforced plastic (FFRP) architecture concept consists of radially aligned growth of CNT on microfibers (a). Conformal and high-yield radial growth of CNTs is observed on CFs using 0.25 wt % K-PSMA/Fe³⁺ coatings and low temperature CVD (b).

Figure 2 CNT growths are observed across the tow (c), with sections exhibiting radial alignment (a), and conformal coverage around each fiber (b). Cross section of fuzzy CFRP shows resin fully infused between fibers (d).

Figure 3 Typical stress-strain curves for ~67% Vf baseline CFRP and fuzzy CFRP under tensile testing until fracture (left). Fuzzy CFRP curve is offset 2000 μ strain for visualization. Optical image of an impregnated tow/unidirectional composite specimen that was tabbed and tensile tested at a gauge length of 150 mm (right).

Figure 4 Both ~67% Vf fuzzy CFRP chord modulus (left) and strength values (right) are preserved and very close to ROM micromechanical predictions based on single-fiber tensile data. The vertical whiskers represent the minimum and maximum data points, whereas the shaded rectangular box ends represent the 1st and 3rd quartiles. The band inside the box displays the median value, and the diamond marker represents the mean.

Figure 5 CM-SFC birefringent stress patterns during initial state showing no breaks (a), at intermediate levels of stress showing sparse independent fiber breaks (b), and after saturation in which individual fragments interact (c). Note that the fuzzy CF pull-out diameter remains unchanged from the as-received CF baseline case (d). Continuously

monitored single fiber fragmentation test data of baseline (as received) and fuzzy CFs showing decrease in fragment lengths with increasing applied load (e).

7. Tables

Table 1 Parameter results from CM-SFC of as received and fuzzy carbon fibers

Fiber type	Shape parameter β	Scale parameter α (GPa)	Fiber strength at ℓ_c $\sigma_f(\ell_c)$ (GPa)	Critical length ℓ_c (mm)	Interface shear strength τ_i (MPa)
Baseline CFs	26±20	8.24±0.39	8.21±0.68	0.81±0.1	35.4±3.91
Fuzzy CFs	15±7.9	8.68±0.60	8.48±0.48	0.8±0.13	37.4±4.00

Note: Plus-minus values represent one standard deviation. Shape parameters are estimated from the linear fit slopes in the ln-ln plots of Figure 5e, and exhibit larger standard deviations due to slope-estimation of 5 samples. Additionally, scale parameters of both baseline and CNT-grown fibers are twice the tensile strength measured in single-fiber tensile testing. As explained, this result is expected, and attributed to the surrounding matrix minimizing surface defect effects [18].

Table 2 Summary of composite and implied fiber tensile properties from unidirectional baseline and fuzzy CFRP fabricated at ~67% V_f compared with single fiber testing data.

Specimen	Implied Fiber Modulus (GPa)	Implied Fiber Strength (GPa)	Composite Chord Modulus (GPa)	Composite Strength (GPa)
Baseline CFRP	198.3 ± 5.0	3.81 ± 0.15	134.7 ± 5.0	2.59 ± 0.12
Fuzzy CFRP	203.9 ± 7.3	3.77 ± 0.10	138.6 ± 5.0	2.56 ± 0.07
Single CF	203.6 ± 13.4	3.46 ± 0.79	139.2 ± 10.0*	2.36 ± 0.55*
Single Fuzzy CF	201.7 ± 11.9	3.81 ± 0.85	138.0 ± 9.0*	2.61 ± 0.59*

Note: Plus-minus values represent one standard deviation

*Denotes ROM predictions from single fiber testing data.

Figure1

Figure2

Figure3

Figure4

Figure5