

## **Flexural behaviour of CNT-filled glass/epoxy composites in an in-situ environment emphasizing temperature variation**

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### **Abstract**

The paucity of structural defects in carbon nanotube (CNT) with unrivalled mechanical properties has always posed an interest to material scientists for its potential incorporation in soft polymer resins to achieve superior mechanical stability. Present investigation focuses on the assessment of flexural behaviour of glass/epoxy (GE) and multiwalled carbon nanotubes (MWCNT) embedded glass/epoxy (0.3 wt. % of epoxy) (CNT-GE) composites at different in-service environmental temperatures. In-situ 3-point bend tests were performed on GE and CNT-GE composites at -80 °C, -40 °C, room temperature (20 °C), 70 °C and 110 °C temperatures at 1 mm/min crosshead speed. The results revealed that at 110 °C temperature, the flexural strength of GE and CNT-GE composites was significantly decreased by 67% and 81% respectively in comparison to their strength at -80 °C temperature. Similarly, 38% and 77% decrement in modulus was noted for GE and CNT-GE composites respectively. Dynamic mechanical thermal analysis (DMTA) was carried out in the temperature range of -100 °C to 200 °C to correlate the mechanical and thermo-mechanical response of both the material systems. Addition of 0.3 wt. % MWCNT in GE composite resulted in lowering of glass transition temperature ( $T_g$ ) by 12°C. Furthermore, to understand various possible deformation and failure mechanisms, the post failure analysis of the fractured specimens, tested at different temperatures, was carried out using scanning electron microscope (SEM). The critical parameters needed during designing composite structures were calculated and modelled using Weibull constitutive model.

Keywords: Polymer-matrix composites (PMCs); Mechanical properties; Environmental degradation; Fractography; Carbon nanotube

### **1 Introduction**

In contrast to the constituent materials, FRP composites do exhibit quite reliable material properties. The tailorability of the interface/interphase in composite structures has successfully generated a new prospective era for material scientists to design materials with well desired properties for various high performance engineering applications. Restricting the premature failure of FRP composites in various isolated cases has been a major challenge for material engineers from the last decade. The durability and structural integrity of these materials are strongly affected by the in-service environments like high and low temperature [1], loading rate, humidity [2] and other harsh environmental conditions [3]. For laminated composites, the in-plane mechanical properties are governed by the fiber phase whereas the z-direction (perpendicular to the plane of the laminate) properties are mostly limited by the matrix phase [4,5]. Scientifically engineered nano-fillers have been reported to be more

reliable choice to improve these properties, which have been acknowledged round the globe [6]. Along with this improvement, it is of great interest to investigate the effect of nanofiller incorporation on the environmental degradation of these potential materials. Undoubtedly, carbon nanotube (CNT) is one of the most promising nanofillers for effective enhancement of the mechanical, electrical and thermal properties of polymer matrix composites (PMCs) [7–12]. The low and cryogenic temperature strength of epoxy has been reported to be significantly enhanced due to addition of CNTs [13]. As temperature induced damages further may degrade the properties of the composite material, its structural integrity must be ensured at the service temperature [14]. Several literatures are available explaining the pros and cons of CNT modified polymer, but current time demands fabrication and effective evaluation of mechanical performance and integrity of CNT modified hierarchical composite materials at various temperature environments [15,16].

Present study aims to investigate the mechanical and thermo-mechanical performance of MWCNT modified glass fiber/epoxy (CNT-GE) composites at elevated and low temperature environments. This, in turn, proposes the applicability, reliability and predictability of the CNT-GE composite.

## 2 Experimental Section

### 2.1 Materials and processing of composite laminates

The epoxy resin used in the present study is diglycidyl ether of Bisphenol A (DGEBA) and the hardener is Triethylene tetra amine (TETA), supplied by Atul Industries Ltd, Gujarat, India under the trade name of Lapox L-12 and K-6 respectively. The glass fiber used is a 3K plain weave type with 15  $\mu\text{m}$  filament diameter supplied by Saint-Gobain, India. The MWCNTs having an outer diameter of 6-9 nm with 5  $\mu\text{m}$  length were purchased from Sigma-Aldrich. Some salient properties of the constituents of laminated composite are provided in Table-1.

Table-1

Salient properties of epoxy and glass fiber

Property	Epoxy	Glass fibres
Density ( $\text{g}/\text{cm}^3$ )	1.162	2.58
Tensile modulus (GPa)	4.1	72.3
Tensile strength (GPa)	0.11	3.4
Strain at failure (%)	4.6	4.8
Poisson's ratio	0.3	0.2

The volume fraction of fibres is approximately 50%. For fabricating CNT-GE composite, the epoxy resin was modified with MWCNT prior to hand lay-up method. The MWCNT content in the epoxy resin was kept at 0.3 wt. % of the epoxy taken. Required amount of CNTs was taken and dispersed in 250 mL acetone. The suspension was then stirred at 1000 rpm for 30 minutes followed by sonication for 30 minutes at room temperature. At this stage, stirring and sonication were done to facilitate deagglomeration of the as supplied CNTs in acetone

medium. The CNT/acetone mixture was then transferred to a flask containing the pre-weighed epoxy, which was already heated at 70 °C for 30 minutes to impart sufficient fluidity. Then the epoxy/CNT/acetone mixture was stirred at 1000 rpm for 2 hrs at 70 °C. Sonication was again carried out at 70 °C for 1 hr. At this stage, removal of all acetone was ensured. Then the epoxy/CNT suspension was kept in vacuum for 12 hrs. This was done for the removal of air bubbles entrapped in the suspension during stirring and sonication. Then to the suspension, required amount of hardener (10 wt. % of epoxy as specified by the supplier) was added and mixed well. The laminated composites were prepared by hand lay-up method with 14 layers of woven fabric cloth of reinforcement and then placed in a hot press. Then the curing of the laminate was carried out at 60 °C temperature and 10 kg/cm<sup>2</sup> pressure for 20 minutes. Similarly, the controlled glass fiber/epoxy (GE) composite laminate was fabricated using 14 layers of glass fiber and neat epoxy by hand lay-up method followed by hot compression using the same parameters as were in CNT-GE composite. The laminates were then kept at room temperature for 24 hrs. The flexural (as per ASTM D7264) and DMTA (as per ASTM D7028) samples were cut from the laminates using a diamond tipped cutter. After cutting, all the samples were post-cured at 140 °C for 6 hrs.

## **2.2 Thermo-mechanical and mechanical characterization**

The thermo-mechanical analysis was carried out as per ASTM D7028 in the dynamic mechanical thermal analyser (DMTA) (Netzsch DMA 242E) from -100 °C to 200 °C temperature range at a heating rate of 10 °C/min. In DMTA, the specimens were loaded in a 3-point bending mode using a frequency of 10 Hz.

In-situ flexural testing of the samples was carried out at -80 °C, -40 °C, room temperature (20 °C), 70 °C and 110 °C (with a holding time of 10 minutes) with a loading rate of 1 mm/min using the 3-point fixture in the environmental chamber of universal testing machine (Instron 5967) following ASTM D7264 standard. The instrumental set-up with the loading fixture for DMTA and in-situ flexural testing is shown in figure 1.

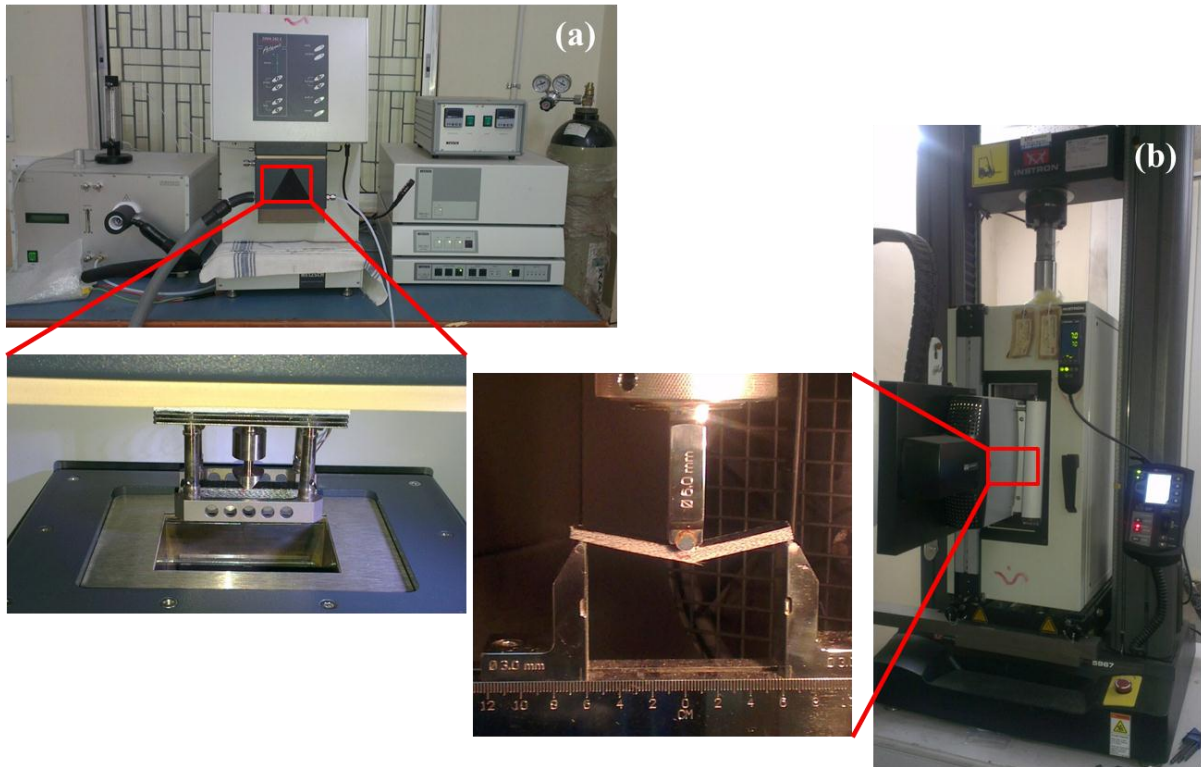


Figure 1: Experimental set-up used for (a) DMTA using Netzsch DMA 242E and (b) in-situ flexural testing using Instron 5967 with environmental chamber.

### 3 Results and Discussion

#### 3.1 Dynamic Mechanical Thermal Analysis (DMTA)

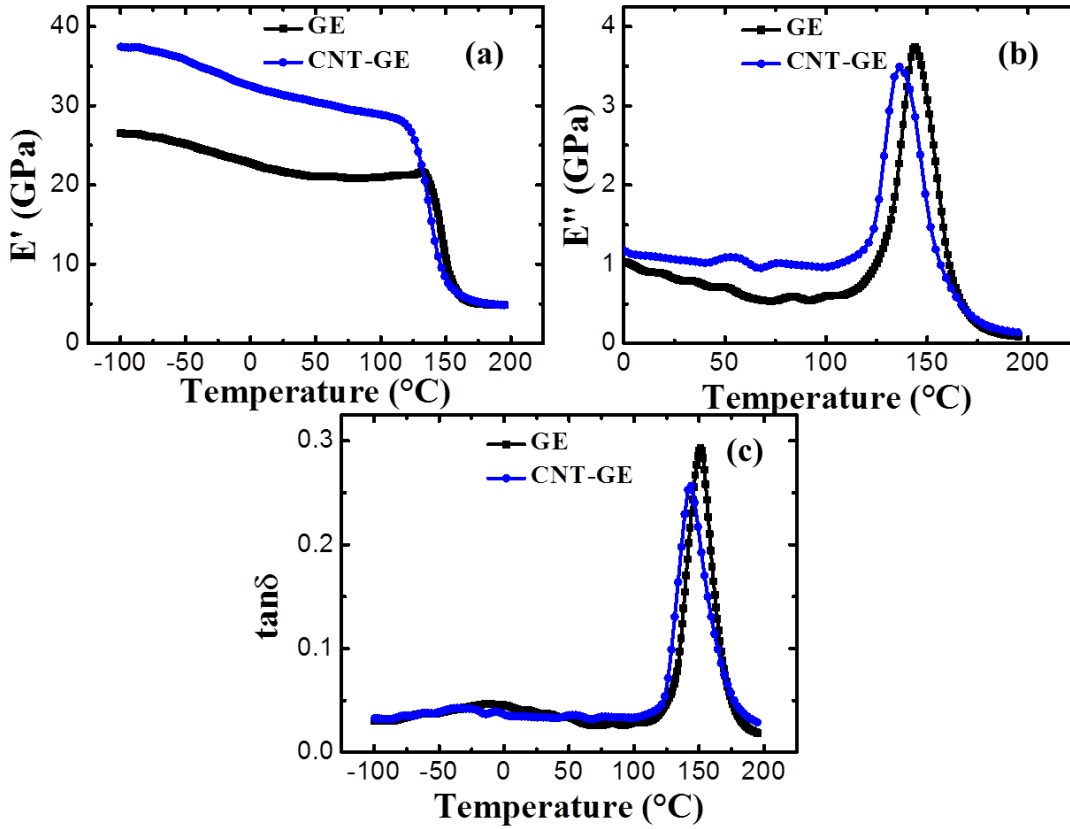


Figure 2: Variation in (a)  $E'$ , (b)  $E''$  and (c)  $\tan\delta$  with temperature for GE and CNT-GE composites.

DMTA is a tool to evaluate the viscoelastic response of the material for a wide range of temperature. The instrument applies a dynamic load to the sample and the response of the material is recorded in the form of dynamic displacement. For a perfectly elastic solid, the applied stress and resulted strain remain in phase, whereas there is a phase difference in case of polymeric (visco-elastic) material. The storage modulus ( $E'$ ) obtained from DMTA is a representation of the elastic modulus of the material, whereas the loss modulus ( $E''$ ) reflects the viscous modulus. The damping tendency of the material is determined from the parameter  $\tan\delta$  (ratio of  $E''$  to  $E'$ ). The  $E'$ ,  $E''$  and  $\tan\delta$  are determined using the following equations.

$$E' = \frac{\sigma_o}{\varepsilon_o} \cos \delta \quad (1)$$

$$E'' = \frac{\sigma_o}{\varepsilon_o} \sin \delta \quad (2)$$

$$\tan \delta = \frac{E''}{E'} \quad (3)$$

where,  $\sigma_o$  and  $\varepsilon_o$  represent the peak stress and peak strain respectively and  $\delta$  is the phase difference between the stress and strain.

Figure 2 refers the variation in  $E'$ ,  $E''$  and  $\tan\delta$  with temperature for GE and CNT-GE composites. It is evident from figure 2(a) that at temperature below glass transition temperature ( $T_g$ ), the  $E'$  for CNT-GE composite is higher than that of GE composite. The rate of decrement of  $E'$  with temperature (upto  $T_g$ ) is higher in CNT-GE composite than that of GE composite. The onset of sharp change in slope of the  $E'$  verses temperature curve is taken as the  $T_g$  of the material. From figure 2(a) it can be observed that addition of 0.3 wt. % CNT results in lowering of the  $T_g$  from 139 °C to 127 °C. This decrement in  $T_g$  may be attributed to the hindrance of crosslink formations due to the entrapment of CNT between the polymeric chains. Figure 2(b) shows the change in  $E''$  due to CNT incorporation in the composite. The  $\tan\delta$  value reduces due to CNT addition showing more induced brittleness in the composite as observed from figure 2(c).

### 3.2 Flexural behaviour at various testing temperatures

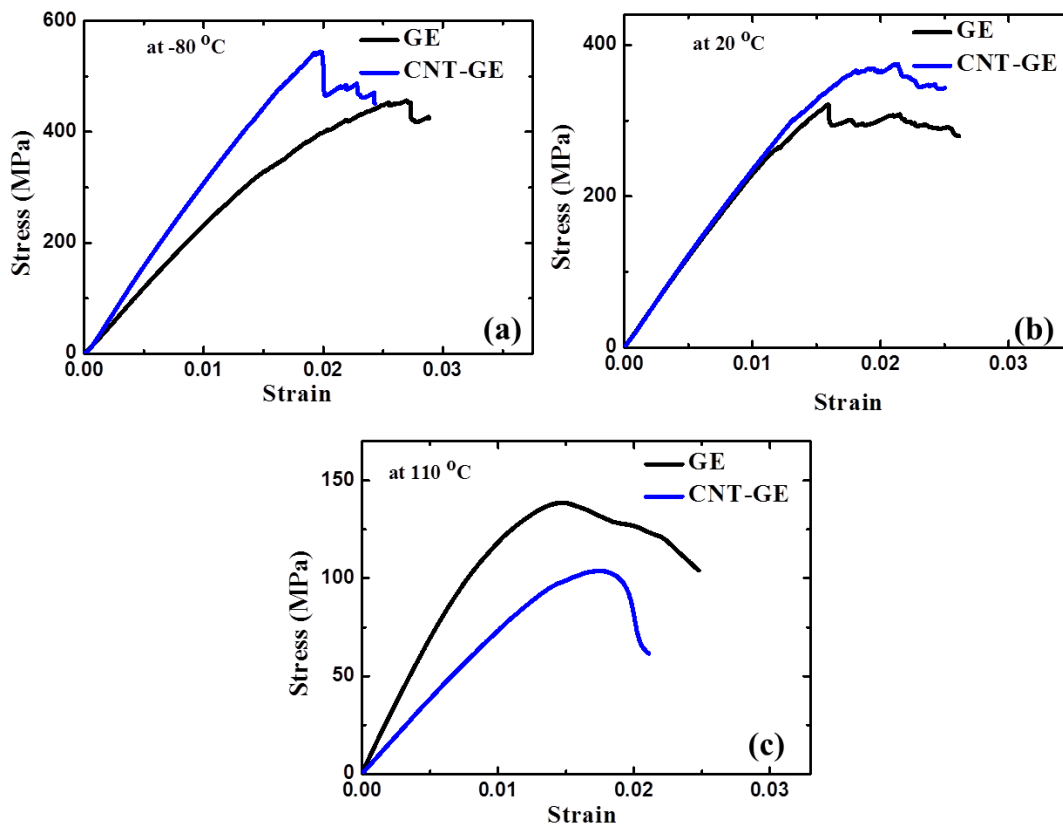


Figure 3: Flexural stress-strain curves for GE and CNT-GE composites at (a) -80 °C, (b) room temperature ( 20 °C) and (c) 110 °C.

The stress-strain curves of the GE and CNT-GE composites at low, room and elevated temperature environments are shown in figure 1. As the  $T_g$  of both the composites are below 140 °C, in-situ flexural testing was carried out up to the maximum testing temperature of 110 °C. The flexural properties i.e. flexural modulus and flexural strength were plotted against testing temperature for both the composites as shown in figure 4.

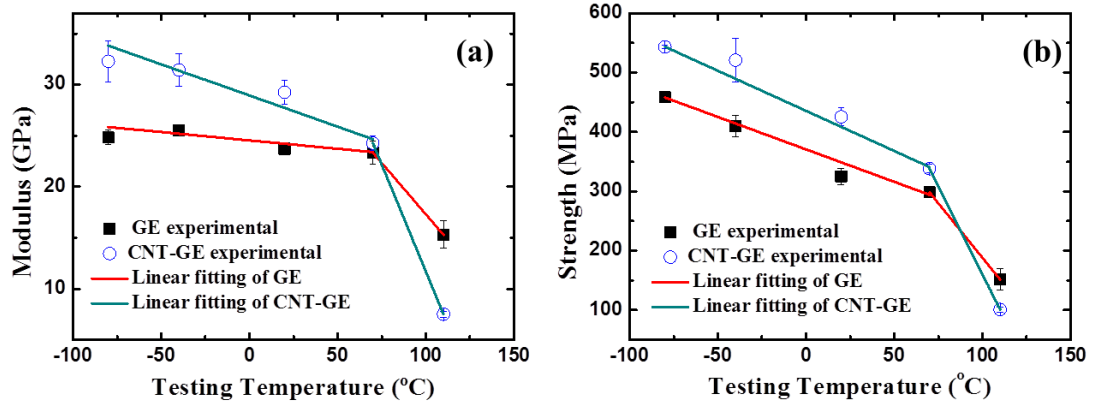


Figure 4: Variation in (a) Flexural Modulus and (b) Flexural Strength with testing temperatures for GE and CNT-GE composites.

It is clearly evident from figure 4 that the flexural modulus and strength are strongly dependent on the testing temperature for both the composite systems. The modulus and strength dependency of GE and CNT-GE composites on testing temperature (within -80 °C to 70 °C) is expressed below.

For GE composites:

$$E \text{ (GPa)} = 24.54 - 0.02 T \text{ (}^\circ\text{C)} \quad (4)$$

$$\sigma \text{ (MPa)} = 370.66 - 1.09 T \text{ (}^\circ\text{C)} \quad (5)$$

For CNT-GE composites:

$$E \text{ (GPa)} = 28.94 - 0.06 T \text{ (}^\circ\text{C)} \quad (6)$$

$$\sigma \text{ (MPa)} = 435.55 - 1.35 T \text{ (}^\circ\text{C)} \quad (7)$$

From equation (4) and (6), the rate at which the modulus decreases with increase in temperature for GE and CNT-GE composites is 0.02 GPa/°C and 0.06 GPa/°C respectively. Similarly the rate of reduction in strength with high testing temperature for GE and CNT-GE composites is 1.09 MPa/°C and 1.35 MPa/°C respectively as seen from equation (5) and (7).

As CNTs possess very high surface area, which in turn exhibits very high interfacial area in CNT/epoxy matrix, the stress transfer ability across the interface is easily facilitated. Hence a higher strength in case of CNT-GE composite is observed than GE composite at room temperature environment.

The modulus and strength enhancement in GE composite with decreasing temperature may be attributed to two factors: (i) Matrix hardening - with decrease in temperature, due to decrease in polymer chain flexibility and free volume and increase in rigidity of the macromolecules, the bond strength in epoxy increases and (ii) interfacial interlocking – as temperature is lowered, due to higher radial contraction of epoxy than glass fiber, a

mechanical gripping is established at the glass fiber/epoxy interface. Hence a high interfacial strength due to mechanical interlocking/bonding is achieved. These two factors are combinedly responsible for high flexural strength and modulus of the GE composite with decreasing temperature. These two factors also remain valid in case of CNT-GE composite. In addition, the extent of matrix hardening is further high due to CNT incorporation. The reason for this higher hardening behaviour may be again attributed to the interfacial interlocking at the CNT/epoxy interface as epoxy shrinks at a higher rate than CNT [13,14]. This immense matrix hardening behaviour is reflected in the higher flexural strength and modulus of CNT-GE composite than GE composite at lower temperatures.

As temperature is increased from room temperature to 70 °C, the radial expansion of epoxy being higher than CNT and glass fiber, some microcracks nucleation is expected at the glass fiber/epoxy and CNT/epoxy interfacial zone. This in addition to polymer softening reduces the strength of both the composites with respect to their properties at room temperature. As temperature is further increased to 110 °C, the density of the microcracks becomes very high. Hence coalescence of these microcracks takes place, converting them into a potential crack. This crack under the action of applied stress propagates and hence reduces the strength of the composite drastically, which can be seen from figure 4(a) and 4(b) as a sharp fall in both strength and modulus between 70 °C and 110 °C. Further, the rate at which strength/modulus decreases with increase in temperature is higher in case of CNT-GE composite than GE composite in the temperature range of 70 °C to 110 °C. This high degree of reduction is ascribed towards the formation of additional microcracks/ microvoids due to high interfacial debonding at CNT/epoxy interface in case of CNT-GE composite. The sharp decrease in the modulus can be correlated to the reduction of  $E'$  obtained in DMTA close to the  $T_g$  temperature (figure 2(a)).

### 3.2.1 Constitutive flexural deformation model

In general, the deformation/failure of laminated composite is the net result of a number of deformation/failure micro mechanisms such as formation of shear cusps, kink bands, local flexure, micro buckling, etc. which ultimately results in failure modes like fiber fragmentation, matrix cracking, fiber/matrix interfacial debonding. The stress( $\sigma$ )-strain( $\varepsilon$ ) relationship for a fiber reinforced composite can be modelled with Weibull distribution function [17,18].

$$\sigma = E\varepsilon \exp \left[ - \left( \frac{E\varepsilon}{\sigma_o} \right)^\beta \right] \quad (8)$$

where  $E$  is the flexural modulus of the composite.  $\sigma_o$  and  $\beta$  are the Weibull scale parameter and shape parameter respectively.  $\sigma_o$  is a measure of nominal strength of the composite, whereas  $\beta$  is a measure of randomness in strength.  $\sigma_o$  and  $\beta$  can be evaluated by taking double logarithm on both sides of equation 8.

$$\ln \left[ \ln \left( \frac{E\varepsilon}{\sigma} \right) \right] = \beta \ln(E\varepsilon) - \beta \ln(\sigma_o) \quad (9)$$



Equation 9 represents a straight line between  $\ln(E\varepsilon)$  and  $\ln\left[\ln\left(\frac{E\varepsilon}{\sigma}\right)\right]$  which can be seen from figure 5. The slope of the straight line gives the value of  $\beta$  and from the value of intercept and  $\beta$  the value of  $\sigma_0$  can be determined.

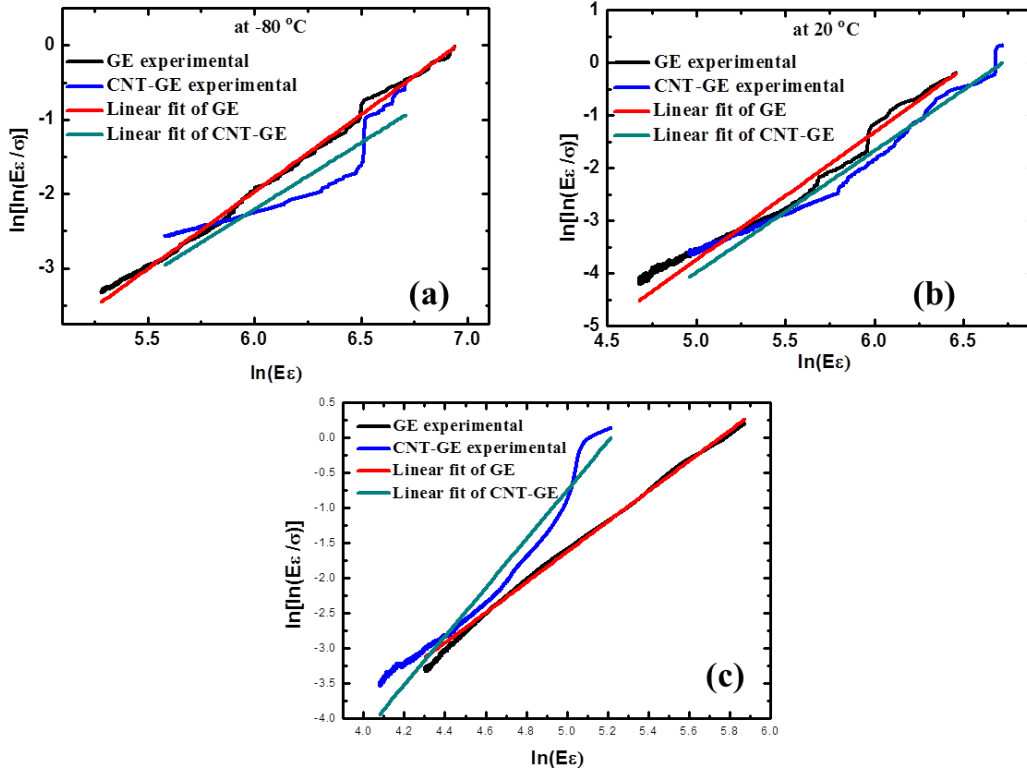


Figure 5: Weibull fitting of experimental data for GE and CNT-GE composites at (a) -80 °C, (b) room temperature (20 °C) and (c) 110 °C.

The parameters of the Weibull function for GE and CNT-GE composites obtained from figure 5 at various testing temperatures are reported in table 2.

Table-2

Weibull scale ( $\sigma_0$ ) and shape ( $\beta$ ) parameters for GE and CNT-GE composites at various temperatures

Temperature (°C)	$\sigma_0$ (MPa)		$\beta$	
	GE	CNT-GE	GE	CNT-GE
-80	1052.6±5.3	1571.5±293.6	2.02±0.13	1.72±0.23
-40	853.4±45.3	1169.3±228.7	2.19±0.08	2.42±0.18
20	758.7±51.7	818.2±33.8	2.39±0.06	2.40±0.12
70	612.5±11	734.1±22.4	2.61±0.06	2.22±0.10
110	338.4±18.4	181.2±5.4	2.16±0.04	3.40±0.09

For both composites,  $\sigma_0$  follows a trend similar to that of flexural strength as observed in figure 4(b). For GE composites, the value of  $\beta$  increases steadily with increase in testing temperature till 70 °C, showing better consistency in strength at high temperature. Addition of CNT to GE composite unevenly alters the value of  $\beta$ , thus providing no such trend in case of CNT-GE composite at various testing temperature.

Using the  $\sigma_0$  and  $\beta$  values listed in table-2, the stress-strain curve was drawn for GE and CNT-GE composites at -80 °C, 20 °C and 110 °C and superimposed with the experimental data as shown in figure 6. It is quite evident from the figure that the experimental results are in close agreement with the simulated data.

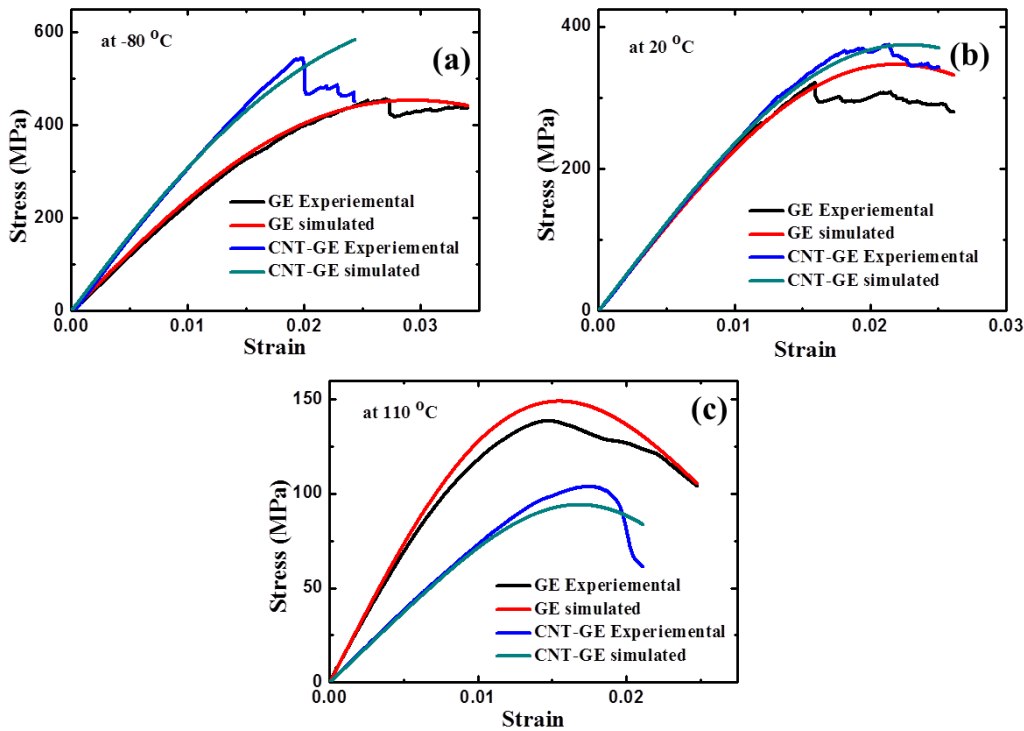


Figure 6: Comparison between experimental and simulated stress-strain curves for GE and CNT-GE composites at (a) -80 °C, (b) room temperature ( 20 °C) and (c) 110 °C.

### 3.3 Fractography

The post failure analysis of the GE and CNT-GE composites was carried out in order to identify various damage and degradation micro-mechanisms responsible for the bulk failure of the material using scanning electron microscope (SEM). Figure 7 shows the SEM images of the fractured surfaces of GE composites at various testing temperatures. Exposure to low temperature (-80 °C), promotes the residual tensile stresses in the matrix causing matrix embrittlement and ultimately reduces the composite toughness. It is clearly evident from figure 7(a) that brittle rupture of the matrix is the dominating failure mode for GE composites at such low temperature environment. At this temperature, the microstructure (figure 7(b)) indicates delaminated surface in conjunction with debris and loose materials. The formation of debris and loose materials is the net result of the abrasion between the fractured surfaces

due to induced shear loading. The delaminated surfaces (figure 7(b)) indicate fiber imprints and river line markings on the polymer in the inter-filamentary region.

At room temperature, fiber/matrix debonding (figure 7(c)) constitutes the major failure micro mechanism and further leading to delamination (figure 7(d)). But neither fiber fragments nor river line markings was noticed on the polymer as shown in figure 7(d). The smooth fiber imprints were observed due to interfacial debonding between the fiber and matrix. Furthermore, the appearance of the fractured glass fiber surface resembles with the three distinct zones, namely mirror, mist and hackle [19].

At elevated temperature (70 °C), softening of polymer matrix may release the residual stresses which can accelerate the fiber/matrix debonding (figure 7(e)) and matrix ductility (figure 7(f)) may become the primary factor for governing the failure in composite.

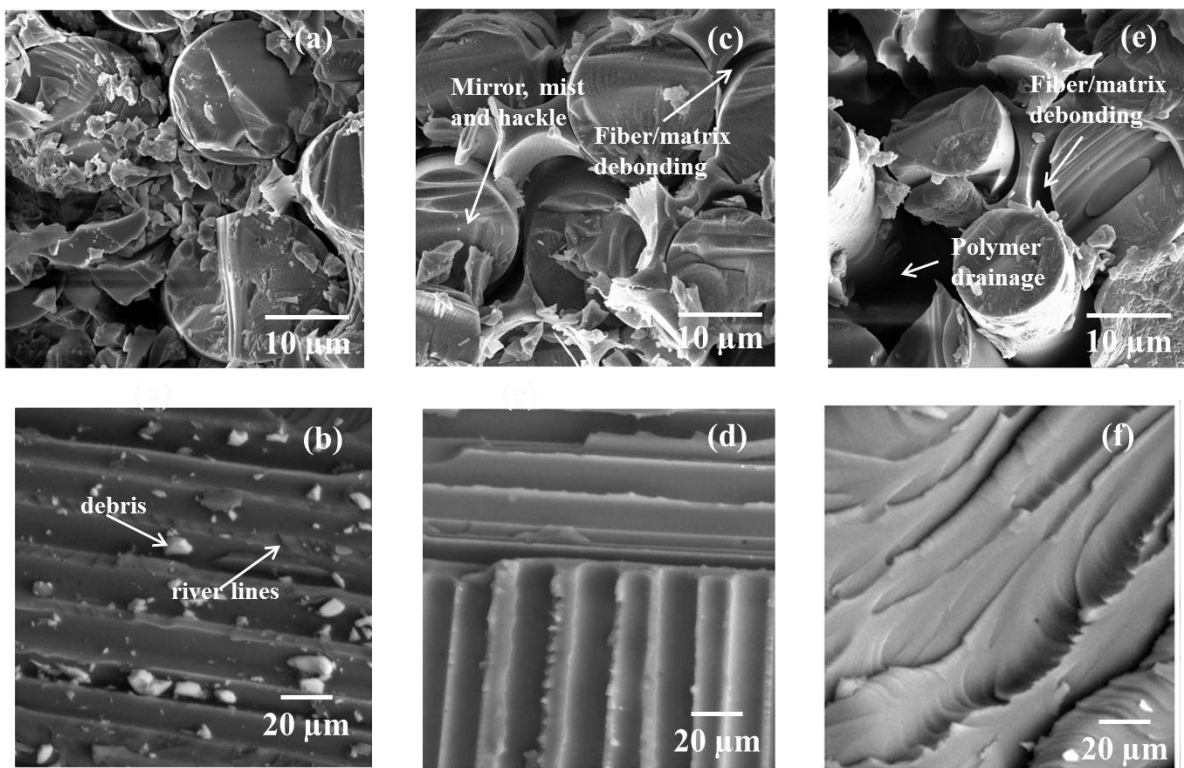


Figure 7: SEM images of the fractured surfaces of GE composites after flexural testing at (a, b) -80 °C, (c, d) room temperature ( 20 °C) and (e, f) 70 °C.

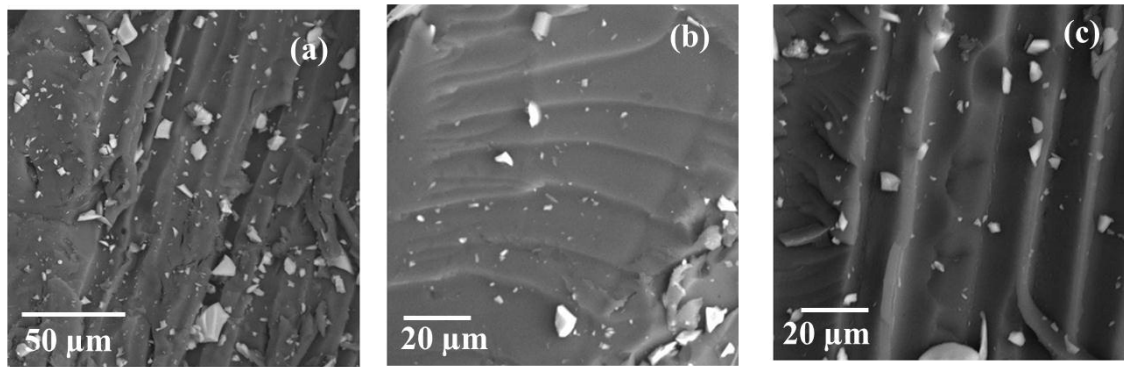


Figure 8: SEM images of the fractured surfaces after flexural testing of GE composites at 110 °C at different locations.

At temperature close to the  $T_g$ , due to induced rubbery nature in polymer, severe degradation takes place as evident from figure 8[(a), (b) and (c)] reflecting drastic reduction in strength and stiffness of the composite (figure 4). At this temperature, sharpness of the edges, river lines and fiber imprints are reduced. Induced flowability due to soft rubbery nature of the matrix can be attributed for this effect.

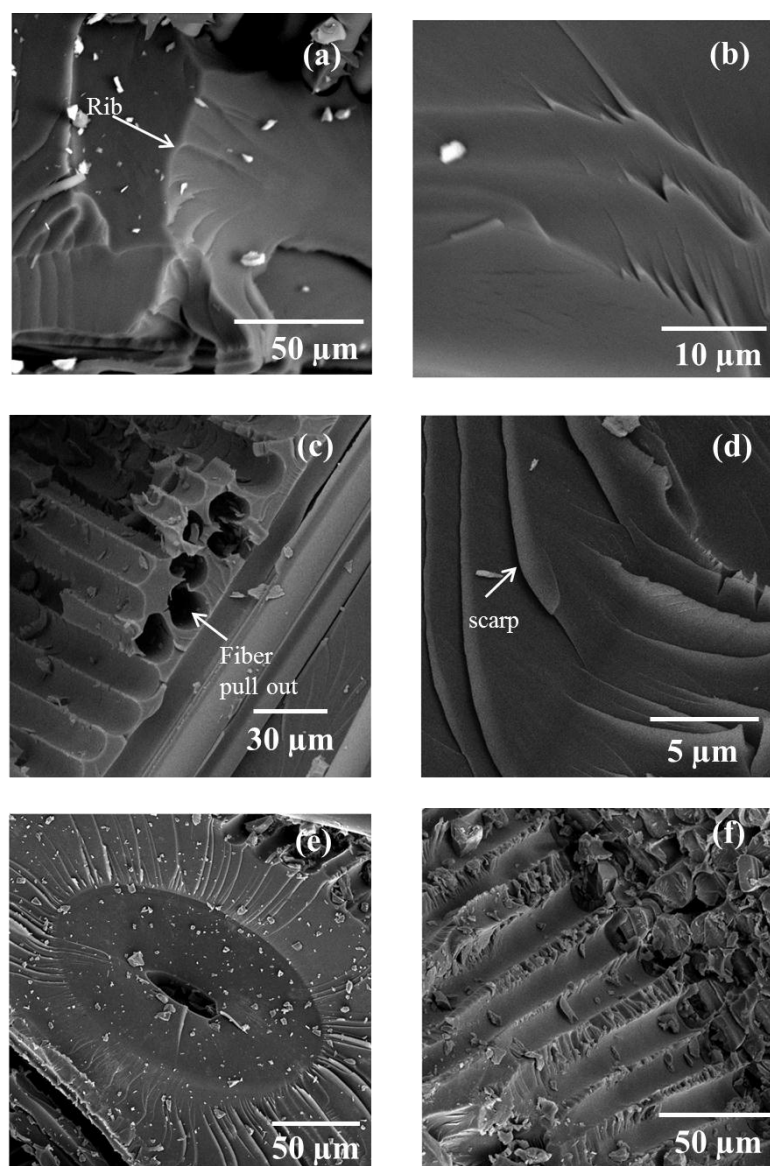


Figure 9: SEM images of the fractured surfaces of CNT-GE composites after flexural testing at (a, b)  $-80\text{ }^{\circ}\text{C}$ , (c, d) room temperature ( $20\text{ }^{\circ}\text{C}$ ) and (e, f)  $70\text{ }^{\circ}\text{C}$ .

In case of CNT-GE composites, tested at  $-80\text{ }^{\circ}\text{C}$ , formation of ribs was noticed as shown in figure 9(a). High interlaminar matrix thickness can result in formation of rib. These ribs are associated with crack arrest lines in case of bulk polymeric materials. Hence, it suggests that in case of CNT-GE composites, crack arrestment by these ribs may hold the key to enhanced strength in contrast to GE composites. Figure 9(b) also indicates river lines on the matrix surface, which is a typical signature of brittle failure. Fiber pull out and delamination as seen from figure 9(c) constitute major failure modes for CNT-GE composites at room temperature environment. Figure 9(d) represents scarps formation on the matrix, at the adjacent fibers, which then spread laterally into the inter-fiber space [19]. Void formation in the matrix is accelerated due to elevated temperature environment as can be seen from figure 9(e). Further under the action of applied load, mirror, mist and hackle zones were observed in the close vicinity of the earlier present void. Figure 9(f) indicates the in-plane shear fracture step formed at elevated temperature in CNT-GE composite.

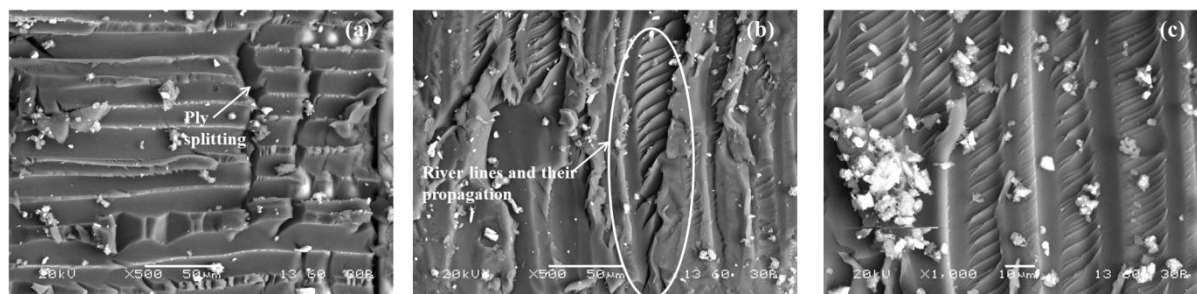


Figure 10: SEM images of the fractured surfaces after flexural testing of CNT-GE composites at 110 °C at various locations

Figure 10(a) indicates initial delamination followed by ply splitting at 110 °C temperature for CNT-GE composite. Although delamination is the primary failure mode in laminated composites, but in conjunction with delamination other micro mechanisms like ply splitting as evident from figure 10(a) can drastically decrease the strength and stiffness of the composite. At 110 °C, after delamination the matrix is unable to withstand further applied load and matrix cracking in terms of ply splitting occurs. Figure 10(b) and 10(c) indicate formation of extensive river lines even at 110 °C. These river lines, under application of load, propagate and further converge into a potential crack.

Figure 11(a) indicates the microstructure of the fractured CNT-GE composite at room temperature environment, showing uniform dispersion of the CNTs in the polymeric matrix. The diameter of the MWCNTs was determined at various locations and indicated in the same figure. As per the specification sheet provided by the MWCNT's manufacturer, the outer diameter was 6-9 nm. The discrepancy in measured and actual diameter of MWCNTs can be attributed to the coating of the polymer on the CNT surface showing good interfacial bonding [20] at this temperature. Figure 11(b) shows the matrix crack bridging by CNTs and CNT pull out phenomena, reported by various researchers [21,22] which enhances the damage tolerance of the composite.

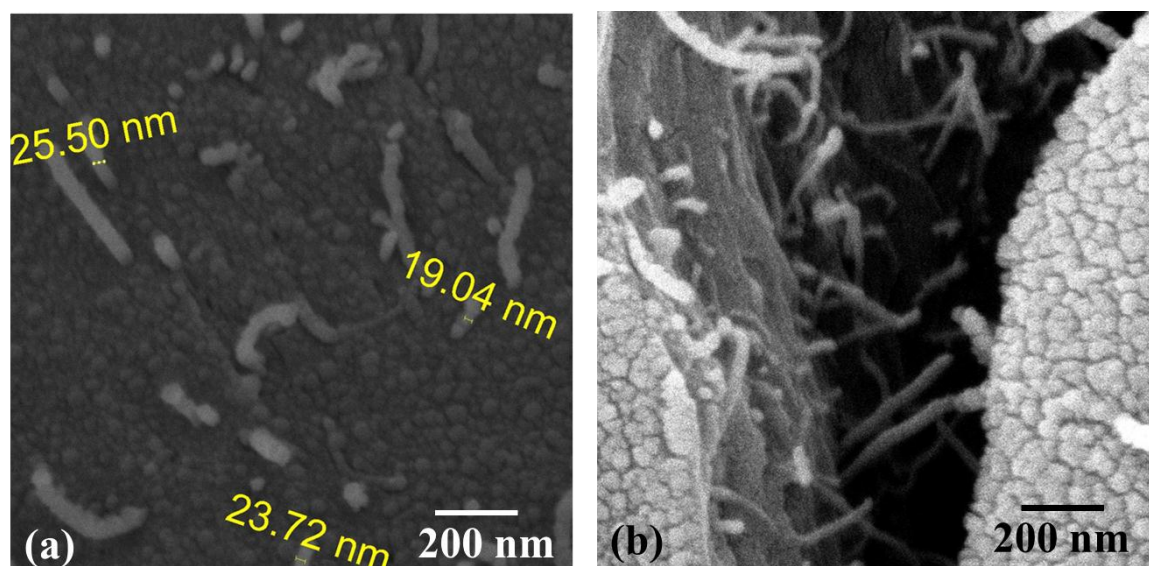


Figure 11: SEM images of the fractured surfaces of CNT-GE composites after flexural testing at room temperature ( 20 °C) showing (a) distribution of MWCNT in the polymer and (b) crack bridging by MWCNTs.

#### 4 Conclusion

The effect of in-service temperature on mechanical behaviour of CNT-GE and GE composites was evaluated. Present investigation seemingly suggests that like conventional hierarchical composites, in-service temperature has a strong impact on the mechanical performance of advanced MWCNT-modified hierarchical composites. At low in-service temperature environment, the CNT-GE composites exhibited higher elastic modulus compared to GE composites, as confirmed from DMTA within the studied range of temperature. Furthermore, addition of 0.3 wt. % MWCNT into GE composite significantly lowered the  $T_g$  by 12 °C due to hindrance in crosslink formations. The reinforcement efficiency (relative change in modulus) due to CNT incorporation in GE composite is as high as 30%, when the testing temperature was -80 °C. It further reduces to 23% when tested at room temperature (20 °C). This suggests that the rate at which both modulus and strength drops with increase in temperature is higher in CNT-GE composites than that in GE composites. At high temperature (110 °C), matrix softening, formation of microvoids and their coalescence upon application of stress caused severe degradation in the composite. This rate of degradation is significantly higher for CNT-GE composites than GE composites (reducing the modulus of CNT-GE composite by 50% than that of GE composite) due to presence of high interfacial area, providing more damage nucleation sites, causing remarkable interfacial debonding. Within the temperature range of -80 °C to 7 °C, the magnitude of Weibull scale parameter ( $\sigma_0$ ) was found to be enhanced by incorporation of 0.3 wt. % MWCNT in GE composite. At -80 °C, the relative increment in  $\sigma_0$  for CNT-GE composite was 50% with respect to controlled GE composite, which was further decreased to 20% at 70 °C. At 110 °C, reduction in  $\sigma_0$  by 50% was noticed in CNT-GE composite as compared to GE composite. The dependency of in-service temperature on mechanical properties is more pivotal in case of MWCNT modified GE composites than conventional GE

composites. This paper uncovers this dependency and widens the scope for researchers in studying environmental damage and degradation in CNT modified conventional FRP composites for its safe application.

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### References

- [1] Sethi S, Rathore DK, Ray BC. Effects of temperature and loading speed on interface-dominated strength in fibre/polymer composites: An evaluation for in-situ environment. *Mater Des* 2015;65:617–26. doi:10.1016/j.matdes.2014.09.053.
- [2] Ray BC. Temperature effect during humid ageing on interfaces of glass and carbon fibers reinforced epoxy composites. *J Colloid Interface Sci* 2006;298:111–7. doi:10.1016/j.jcis.2005.12.023.
- [3] Ray BC, Rathore D. Durability and integrity studies of environmentally conditioned interfaces in fibrous polymeric composites: Critical concepts and comments. *Adv Colloid Interface Sci* 2014;209:68–83. doi:10.1016/j.cis.2013.12.014.
- [4] Pandya KS, Veerajulu C, Naik NK. Hybrid composites made of carbon and glass woven fabrics under quasi-static loading. *Mater Des* 2011;32:4094–9. doi:10.1016/j.matdes.2011.03.003.
- [5] Dong C, Davies IJ. Flexural and tensile strengths of unidirectional hybrid epoxy composites reinforced by S-2 glass and T700S carbon fibres. *Mater Des* 2014;54:955–66. doi:10.1016/j.matdes.2013.08.087.
- [6] Upadhyay AN, Tiwari RS, Mehta N, Singh K. Enhancement of electrical, thermal and mechanical properties of carbon nanotube additive Se85Te10Ag5 glassy composites. *Mater Lett* 2014;136:445–8. doi:10.1016/j.matlet.2014.08.092.
- [7] Gojny FH, Wichmann MHG, Fiedler B, Bauhofer W, Schulte K. Influence of nano-modification on the mechanical and electrical properties of conventional fibre-reinforced composites. *Compos Part Appl Sci Manuf* 2005;36:1525–35. doi:10.1016/j.compositesa.2005.02.007.
- [8] Ajayan PM, Tour JM. Materials Science: Nanotube composites. *Nature* 2007;447:1066–8. doi:10.1038/4471066a.
- [9] Garcia EJ, Wardle BL, John Hart A, Yamamoto N. Fabrication and multifunctional properties of a hybrid laminate with aligned carbon nanotubes grown In Situ. *Compos Sci Technol* 2008;68:2034–41. doi:10.1016/j.compscitech.2008.02.028.
- [10] Li Q, Zaiser M, Blackford JR, Jeffree C, He Y, Koutsos V. Mechanical properties and microstructure of single-wall carbon nanotube/elastomeric epoxy composites with block copolymers. *Mater Lett* 2014;125:116–9. doi:10.1016/j.matlet.2014.03.096.



- [11] Sethi S, Ray BC. An assessment of mechanical behavior and fractography study of glass/epoxy composites at different temperatures and loading speeds. *Mater Des* 2014;64:160–5. doi:10.1016/j.matdes.2014.07.017.
- [12] Jiang Q, Wang X, Zhu Y, Hui D, Qiu Y. Mechanical, electrical and thermal properties of aligned carbon nanotube/polyimide composites. *Compos Part B Eng* 2014;56:408–12. doi:10.1016/j.compositesb.2013.08.064.
- [13] Chen Z-K, Yang J-P, Ni Q-Q, Fu S-Y, Huang Y-G. Reinforcement of epoxy resins with multi-walled carbon nanotubes for enhancing cryogenic mechanical properties. *Polymer* 2009;50:4753–9. doi:10.1016/j.polymer.2009.08.001.
- [14] Sethi S, Ray BC. Environmental effects on fibre reinforced polymeric composites: Evolving reasons and remarks on interfacial strength and stability. *Adv Colloid Interface Sci* n.d. doi:10.1016/j.cis.2014.12.005.
- [15] Lau K, Gu C, Hui D. A critical review on nanotube and nanotube/nanoclay related polymer composite materials. *Compos Part B Eng* 2006;37:425–36. doi:10.1016/j.compositesb.2006.02.020.
- [16] Lau K, Wong T, Leng J, Hui D, Rhee KY. Property enhancement of polymer-based composites at cryogenic environment by using tailored carbon nanotubes. *Compos Part B Eng* 2013;54:41–3. doi:10.1016/j.compositesb.2013.03.044.
- [17] Pervin F, Zhou Y, Rangari VK, Jeelani S. Testing and evaluation on the thermal and mechanical properties of carbon nano fiber reinforced SC-15 epoxy. *Mater Sci Eng A* 2005;405:246–53. doi:10.1016/j.msea.2005.06.012.
- [18] Zhou Y, Pervin F, Lewis L, Jeelani S. Fabrication and characterization of carbon/epoxy composites mixed with multi-walled carbon nanotubes. *Mater Sci Eng A* 2008;475:157–65. doi:10.1016/j.msea.2007.04.043.
- [19] Greenhalgh ES. *Failure Analysis and Fractography of Polymer Composites*. Elsevier; 2009.
- [20] Ding W, Eitan A, Fisher FT, Chen X, Dikin DA, Andrews R, et al. Direct Observation of Polymer Sheathing in Carbon Nanotube–Polycarbonate Composites. *Nano Lett* 2003;3:1593–7. doi:10.1021/nl0345973.
- [21] Hwang GL, Shieh Y-T, Hwang KC. Efficient Load Transfer to Polymer-Grafted Multiwalled Carbon Nanotubes in Polymer Composites. *Adv Funct Mater* 2004;14:487–91. doi:10.1002/adfm.200305382.
- [22] Qian D, Dickey EC, Andrews R, Rantell T. Load transfer and deformation mechanisms in carbon nanotube-polystyrene composites. *Appl Phys Lett* 2000;76:2868–70. doi:10.1063/1.126500.