

Influence of Arctic seawater exposure on the flexural behavior of woven carbon/vinyl ester composites

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Abstract

In this paper, the adverse effects of sea water environment and arctic temperatures on woven carbon fiber/vinyl ester composites are explored in the form of moisture uptake, impact on flexural modulus, strength, and structural damage. The research presented here attempts to relate failure modes to the flexural behavior of these composites exposed to three key environmental conditions: sea water, arctic temperature and combined sea water/arctic condition. Sea water saturation in general degrades the flexural strength up to $\approx 19.45\%$. Microstructures of dry and saturated samples are compared using scanning electron microscopy, where a saturated surface with distinctive hue for wet samples is observed as compared to a rough (parched) surface in the dry samples, implying large concentrations of sea water in a thin layer at the specimen boundaries. Arctic exposure and combined condition on these laminates increase the flexural strength by about 23.1% and 36.2%, respectively. However, they tend to shift the post peak behavior from progressive to brittle-type failure as compared to dry samples, which is attributed to matrix and fiber embrittlement in the material system caused by exposure to low temperature. Further, relatively large variations are observed in the flexural strength values of samples exposed to the combined condition (sea water saturated + arctic), which can be attributed to the freezing of sea water that was entrapped during sea water saturation. Variation in the quantity and location of sea water entrapped can alter the flexural strength significantly. Due to the aforementioned flexural responses and failure behavior observed in woven carbon/vinyl ester

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composites exposed to sea water arctic environment, special consideration is required while designing critical load bearing components in naval applications to avoid possible catastrophic structural failure.

Keywords

Carbon/vinyl ester composites, sea water uptake, flexural behavior, Arctic temperature

Introduction

Recent rise in demand and interest in arctic exploration calls for the understanding of mechanical behavioral aspects of naval structures in extreme conditions. Naval materials typically experience drastic changes and degradation in their macro- and microstructures when exposed to sea water and cold temperatures. Hence, it is critical to focus on the damage and types of failure that occur during service after exposed to extreme conditions, and develop methods for improving the damage resistance, tolerance, and life cycle of the naval materials being used.

Sandwich composites are preferred for naval hull structures due to their lightweight, enhanced performance and affordability. A sandwich composite consists of two thin, but stiff face sheets bonded with a lightweight and damping core between them. Commonly used materials for sandwich face sheets are carbon and glass fiber composites, while aluminum and a variety of foams are utilized for sandwich cores [1]. Naval applications expose these types of materials to adverse environments with respect to time, which can potentially cause surface alterations, internal damage, and degradation of their chemical and mechanical properties, ultimately compromising the safety of the naval structure.

In recent years, research on moisture uptake has covered different types of face sheet composites and enclosing liquids [2–4], but seldom work exists on the combination of carbon fiber composites, sea water absorption, and arctic exposure. Aviles and Aguilar [5] studied moisture absorption in foam-core sandwich composites, focusing individually on its constituents and calculating the moisture saturation content along with their corresponding diffusion coefficients. Investigations on composite laminates were conducted by Loos et al. [6], where polyester E-glass composites were submerged in different fluids, including sea water. Water absorption studies concerning foam cores can be found in the work by Siriruk et al. [7], Kanny et al. [8], and Tagliavia et al. [9].

In this paper, the focus is on the periodic flexural investigation of woven carbon/vinyl ester face sheet composites up to saturation, followed by arctic exposure. Toward that, woven carbon fiber laminates were fabricated and submerged in synthetic sea water until saturation was attained. Samples were weighed and physically tested periodically to determine the influence of moisture uptake and diffusion on their flexural stiffness and strength. A set of saturated samples were also

exposed to arctic conditions and tested in-situ to investigate the influence of sub-zero temperatures on the structural integrity. The failed regions in each case were examined to relate the failure modes to the environmental exposures. This paper is organized in the following sections: manufacturing process, which includes the material systems used and sample dimensions; experimental procedure, which includes details on sea water exposure, moisture uptake measurements, diffusivity calculations, arctic exposure and flexural test procedure. The paper will culminate with detailed discussion of results and conclusion.

Manufacturing

Carbon fiber laminates were fabricated using vacuum-assisted resin transfer molding (VARTM) process. Material system, fabrication process and sample dimensions used in this study are discussed next.

Material system

A total of 3K tow (i.e. 3000 filaments per tow) plain weave carbon fiber fabrics were used to manufacture the laminates investigated in this paper. These fabrics create lightweight and tensile stiffened structural products, and are also compatible with a variety of thermosets and thermoplastics. This type of fabric is commonly used in aerospace, marine, and automobile applications. Hetron 922 vinyl ester resin, formulated for 1.25% methyl ethyl ketone peroxide (MEKP), was the resin system utilized to impregnate the dry carbon fabric. Hetron 922 is a low viscous thermoset, which is advantageous for easy infiltration during the VARTM process. Mechanical properties of the carbon fabric and vinyl ester resin are given in Table 1.

Laminate fabrication

VARTM process was used to fabricate composite panels with 304.8 mm in length \times 304.8 mm in width. Carbon fabric along with additional textiles, such as

Table 1. Constituent material properties.

Property	Carbon fiber	Vinyl ester
Tensile strength	4.2–4.4 GPa	82.7 MPa
Tensile modulus	227.5–240.6 GPa	3.7 GPa
Elongation	1.4–1.95	4.6–7.9
Flexural strength	–	131 MPa
Flexural modulus	–	3.4 GPa
Nom. thickness	0.3048 mm	–
Barcol hardness	–	35

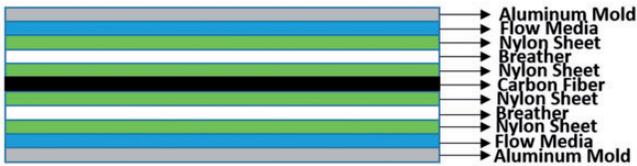


Figure 1. Layer distribution of materials used in the VARTM process.
VARTM: vacuum-assisted resin transfer molding.

flow media, breather, and nylon sheets were also cut to fit aluminum molds of the same size as the expected panels. Material layers required to fabricate a single composite panel were as follows: 2 aluminum molds, 2 flow-media sheets, 4 nylon peel plies, 2 layers of breather, and 16 layers of carbon fiber. A through-thickness schematic arrangement of these material layers is shown in Figure 1. This arrangement was then wrapped with Stretchlon[®] 800 bagging film and sealed with vacuum-sealant tape, ensuring space for both inlet and outlet connectors.

Vinyl ester resin was mixed with MEKP hardener in a container at a weight ratio of 100:1.25. The outlet was then connected to a vacuum pump until the vacuum bag achieved an airtight state of approximately 5.5 bar (gauge) pressure. The inlet of the vacuum bag was then submerged in the resin/hardener mixture allowing flow of the resin through the dry fabric layers, which was assisted by the vacuum created. It is important to note that the resin/hardener mixture was placed in a desiccator prior to infiltration in order to remove air bubbles from the mixture. Upon completion of the resin transfer process, the laminate was cured at room temperature for 24 h.

Sample dimensions

Sixteen layers of dry carbon fabric resulted in laminates with a nominal thickness of ≈ 4 mm. The layers of carbon fabric were stacked along the z-axis such that tows in each layer are oriented along the x- and y-directions as shown in Figure 2. Specimens were water-jet cut to a length of 153.6 mm and a width of 13 mm (see Figure 2). Dimensions were based on the ASTM standard D7264 [10], which require a span-to-thickness ratio of 32:1 for accurate flexural results. These specimens were also used to determine moisture saturation content and diffusivity, following the ASTM standard D5229 [11], as well as periodic flexural behavior.

Experimental approach

The motivation behind the experimental investigations was to determine the degradation of the flexural properties of carbon woven composites exposed to sea water and arctic temperatures. Toward that, the weight change percentage of carbon fiber composites caused by sea water absorption at room temperature was periodically measured and the samples were subsequently tested under flexural

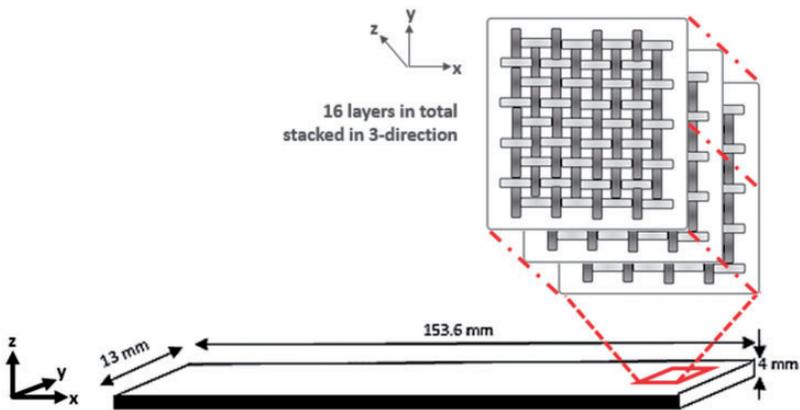


Figure 2. Dimensions of carbon fiber samples.

Table 2. Overview of test matrix.

Temperature	Environment	Test type	No. of specimen	Notes
Room (23°C)	Dry	Flexural	5	Baseline
	Sea Water	Sorption	60	Periodic weighing + diffusivity calculation
		Flexural	60	Periodic testing
Arctic (−60°C)	Dry	Flexural	5	In situ
	Sea Water	Flexural	5	In situ

loading. The diffusivity constant (D) was first determined to establish the extent of diffusion of sea water into woven carbon–vinyl ester composites. Further, a separate set of samples were exposed to -60°C for 60 days followed by flexural tests in situ at -60°C . Also, a set of sea water saturated samples were conditioned at -60°C for 48 h and tested under flexural loading at in situ -60°C temperature. An overview of the test matrix is shown in Table 2, and details of each experimental step conducted are given in the following subsections.

Sea water exposure

For sea water exposure study, the edges of the specimens were sealed with epoxy resin before exposing them to sea water to minimize infiltration through the edges. Sealing the edges with resin also considers certain real-life applications. For example, the hull of a ship composed of composite-material panels would only contact sea water from one surface, absorbing the fluid only in the through-thickness direction. Panel surfaces in real applications are typically coated, however, they can be

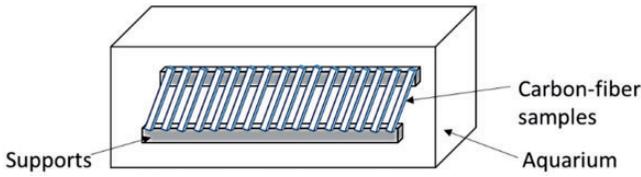


Figure 3. Experimental setup for moisture uptake study.

in contact with the environment in the case of coating wear-out. Thus, the approach adopted in this paper enables the specimens toward one-dimensional absorption. Albeit, the end use of a composite face sheet would require more detailed adaptations of real conditions, such as coatings, they are beyond the scope of this paper. The specimens were weighed and dimensions measured after the curing of the epoxy resin applied at the edges. The specimens were then placed inside an oven for moisture desorption at 80°C for a total of 6 h. A 20-gallon fish tank was used as a water bath for submerging the samples in synthetic sea water. The samples were placed on supports (shown in Figure 3) to ensure exposure of the top and bottom faces to the surrounding liquid.

Submerged specimens were removed from the sea water bath periodically for weighing. The specimens were wiped first to remove excess moisture and then weighed using an analytical balance with an accuracy to three decimals. The following expression was used to calculate the weight change percentage caused by water uptake:

$$\text{Mass change(\%)} = \left| \frac{w_t - w_b}{w_b} \right| \times 100 \quad (1)$$

where, w_t is the weight at the time of removal and w_b is the baseline weight. The samples were weighed every 24 h at the beginning of the water uptake experiment. The weighing frequency was gradually reduced until reaching a frequency of once per week. Sets of submerged samples were also removed from the water bath periodically for flexural testing.

One of the objectives of the water uptake study was to determine the diffusivity coefficient of sea water in woven carbon–vinyl ester composites. Fick's laws of diffusion have been widely used to describe moisture absorption behavior in many materials. In particular, Fick's second law is expressed in the following form:

$$\frac{dC}{dt} = D \frac{d^2C}{dz^2} \quad (2)$$

where, D is the diffusivity coefficient, C is the concentration of fluid, and z is the direction of fluid diffusion. As given in Aviles and Aguilar [5], equation (2) can be

solved to determine the overall diffusion coefficient, given by:

$$D = \pi \left(\frac{h}{4M_{sat}} \right)^2 \left(\frac{dM(t)}{d\sqrt{t}} \right)^2 \quad (3)$$

where h is the thickness of the sample, M_{sat} is the maximum saturation content, $M(t)$ is the moisture content at a given time t . The first part of equation (3) represents the saturation stages of moisture uptake of the material, while the second part represents the initial slope at the beginning of water absorption.

Arctic exposure

To the knowledge of the authors, there is no standard for arctic exposure studies on polymeric composites. Hence, a test procedure was developed in-house and followed. Specimens were initially weighed and measured followed by drying in an oven at 80°C for a period of 6 h. The samples were then placed inside a freezer at -60°C for a period of 60 days, after which they were tested in an arctic in situ environment at -60°C to determine their flexural properties. Liquid nitrogen was used in an Instron environmental chamber to attain -60°C for the in situ tests. For the purpose of examining the effects of sea water saturation at low temperatures, room temperature sea water saturated samples were first pre-conditioned in a freezer at a temperature of -60°C for 48 h, followed by in situ flexural tests at -60°C.

Flexural tests

One of the main objectives of this research was to examine the evolution of flexural properties of the woven composites exposed to sea water with respect to the extent of saturation and time. Three-point bend tests were conducted to determine the flexural modulus and strength, which were calculated from the load-displacement responses obtained from the experiments. The span length provided by the fixture was 128 mm and the diameter of support rollers was 10 mm. Specimens were tested in an INSTRON 5969 machine at a loading rate of 1 mm/min as suggested by the ASTM standard D7264. Stress, strain, and the chord flexural modulus values were calculated using the following expressions:

$$\sigma = \frac{3PL}{2bh^2}; \quad \epsilon = \frac{6\delta h}{L^2}; \quad E_f^{chord} = \frac{\Delta\sigma}{\Delta\epsilon} \quad (4)$$

where σ is the stress at the mid-span outer surface, ϵ is the maximum strain at the surface, E_f^{chord} is the flexural chord modulus of elasticity, P is the applied force, L is the support span, b is the width of the sample, h is the thickness of the sample, and δ is the mid-span deflection.

As mentioned above, arctic conditioned samples were tested within an in situ environmental chamber at a temperature of -60°C with LN_2 as the working fluid. This environmental chamber was situated with the same INSTRON 5969 machine used before. The calculation procedure described above was used for the arctic samples as well.

In order to elaborate on the impact of the environmental conditions (sea water absorption and arctic temperature) considered in the current paper, an equation that captures the influence of cross-sectional changes in an average sense on the effective modulus is given below based on damage mechanics:

$$\tilde{E} = (1 + \omega_{\text{swelling}} - \omega_{\text{shrinkage}} - \omega_{\text{degradation}})E \quad (5)$$

where \tilde{E} is the effective elastic modulus, ω_{swelling} is the influence of swelling caused by sea water absorption, $\omega_{\text{shrinkage}}$ is the influence of shrinkage caused by arctic temperature (thermal contraction), $\omega_{\text{degradation}}$ is the impact of peripheral degradation caused by sea water absorption, and E is the original (baseline) elastic modulus.

Each of the contributions through ω is defined as follows:

$$\begin{aligned} \omega_{\text{swelling}} &= \frac{S_{\text{swelling}}}{S_{\text{baseline}}}; & \omega_{\text{shrinkage}} &= \frac{S_{\text{shrinkage}}}{S_{\text{baseline}}}; \\ \omega_{\text{degradation}} &= \frac{S_{\text{degradation}}}{S_{\text{baseline}}} \end{aligned} \quad (6)$$

where S_{baseline} is the initial pristine cross-sectional area of the specimen; S_{swelling} , $S_{\text{shrinkage}}$, and $S_{\text{degradation}}$ represent the cross-sectional area changes with respect to different environmental conditions, and are illustrated in Figure 4. It is important to note that when the material is not exposed to environmental conditions, $S_{\text{swelling}}=0$, $S_{\text{shrinkage}}=0$, and $S_{\text{degradation}}=0$, resulting in $\tilde{E}=E$. Equations (5) and (6) were used later in this work to explain variations in the flexural modulus between different environments.

Results and discussion

Sea water uptake and diffusivities

Equation (1) was used to obtain sea water uptake curves with respect to time. Moisture uptake measurements revealed a moisture saturation content, M_{sat} , of approximately $1.5 \pm 0.1\%$. From earlier studies, vinyl ester composites have shown moisture uptake percentages ranging from 0.4% to 1.2%, depending on the material configuration and environmental parameters [12–14]. In Kootsookos et al. [14], vinyl ester reinforced composites exposed to sea water displayed a high moisture uptake percentage of approximately 1.2%. In Loos et al. [6], M_{sat} was observed to be 1.4% for submerged polyester E-glass composites in salt water. It is widely

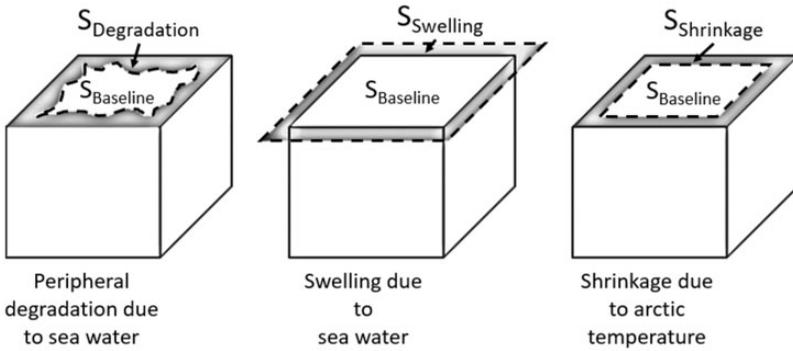


Figure 4. Cross-sections of damaged material due to different environments.

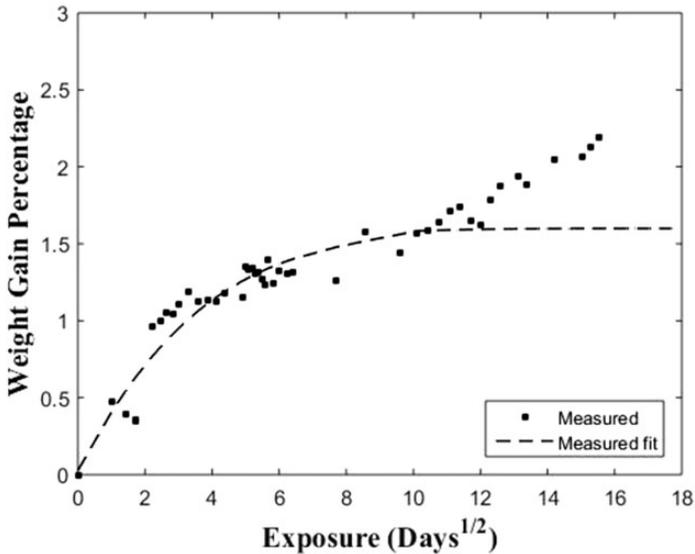


Figure 5. Sea water uptake of woven carbon/vinyl ester composites.

known that epoxy systems are less resistant to moisture uptake as compared to vinyl ester or polyester systems. A slightly higher value for moisture saturation content presented here can be attributed to manufacturing defects like void formation during in-house manufacturing of the composites, as well as to the differences in the material system, liquid type, exposure, and temperature with respect to other studies. Weight gain percentage of one specimen with respect to time is shown in Figure 5(a), which appears to follow Fickian behavior with a linearly increasing region in the initial stage and a plateau region around 120 days of exposure. It is observed from Figure 5(a) that weight gain appears to increase beyond the

Table 3. Diffusion coefficients and moisture saturation content.

Material	D (mm ² /s)	M_{sat} (%)
Laminate	1.81×10^{-6}	1.5 ± 0.1

saturated plateau region. This increase is attributed to other effects occurring in the material as explained in the next paragraph, and is not considered as a manifestation of sea water absorption.

Non-Fickian behavior in a material is said to occur, as proposed by many researchers in moisture uptake studies [15–17], when sorption data out lies the expected saturation region in the form of a decrease, increase, or a fluctuation beyond a plateau region of saturation. This change, often referred to as “non-Fickian phenomena”, beyond the apparent saturation region in the moisture absorption curve requires further investigation into several physical parameters. For example, surrounding temperature may trigger a change in the expected saturation behavior; Imaz et al. [18] showed that carbon-epoxy laminates in general obey Fickian behavior at low temperatures, but display non-Fickian behavior at high temperatures. Other physical parameters or changes may also alter the saturation process of a material, which include but are not limited to micro-cracking, resin leaching or dissolution, internal or surface cracks, pre-established voids, and surface peeling. Most of the material alterations listed above can potentially cause an overall net weight gain by trapping continuous water even in a saturated state; only surface peeling causes a net weight loss when the materials is saturated [6, 18, 19]. Thus, an increase in the weight gain percentage observed beyond the plateau region of the moisture absorption curve shown in Figure 5(a) is attributed to internal and surface cracks present in the composite.

The average of the diffusivity coefficients determined using the moisture uptake curves and equation (3) is given in Table 3. The average moisture saturation content is also given in Table 3.

Microstructural changes due to sea water uptake

Scanning electron microscopy (SEM) was performed on the cross-sectional area of dry and wet samples. Both types of specimens were diamond-saw cut through their cross-sections near the free edges perpendicular to the specimen length, as shown in Figures 6 and 7. These regions were chosen in order to avoid damage that could have been caused by the three-point bend test rollers. The cut surfaces were mechanically polished before conducting SEM as suggested by Stutzman and Clifton [20]. It is important to note that the wet samples were exposed to synthetic sea water for a period of 200 days, which corresponds to a stage of moisture saturation. As indicated by Aviles and Aguilar [5], an adverse microstructural change within a composite is expected in the form of swelling stresses, cracking, and fiber/matrix

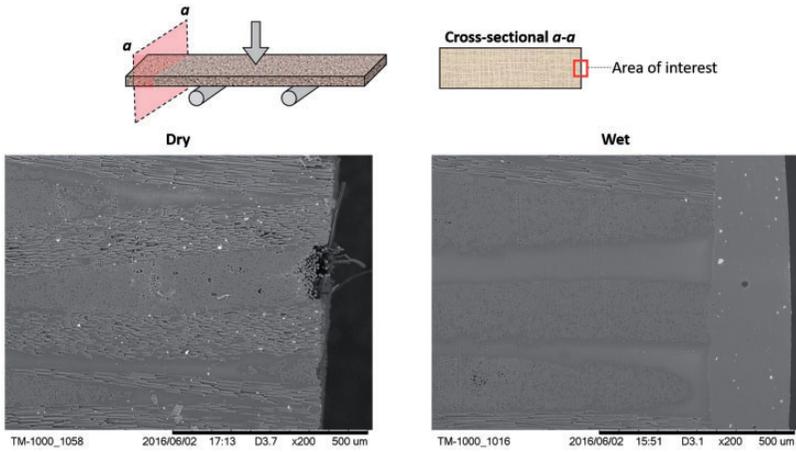


Figure 6. SEM images of the right edge on the cross-section of dry and saturated samples. SEM: scanning electron microscopy.

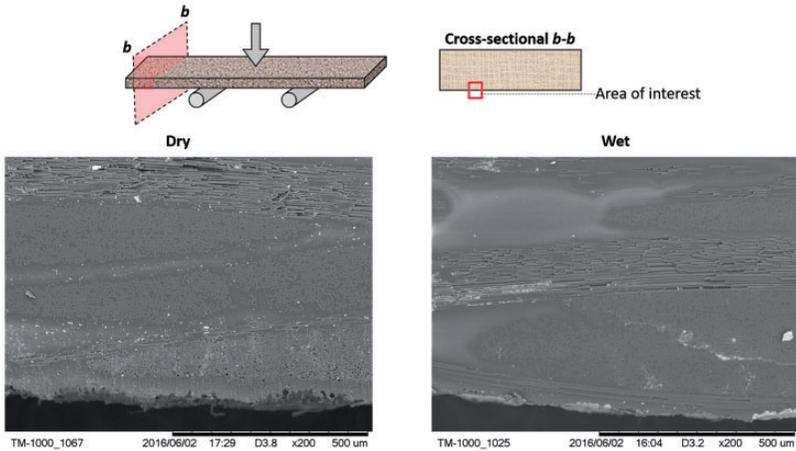


Figure 7. SEM images of the bottom edge on the cross-section of dry and saturated samples. SEM: scanning electron microscopy.

de-bonding, which is attributed to the amount of water absorbed by the material. Figures 6 and 7 show the SEM images of the cross-section of both dry and wet samples at two edges. Wet samples displayed a smoother (saturated) surface when compared to the rough (parched) surface of the dry samples. Also, a distinctive hue is detected within the edges of the wet specimens, suggesting relatively large amounts of concentration of sea water near the edges.

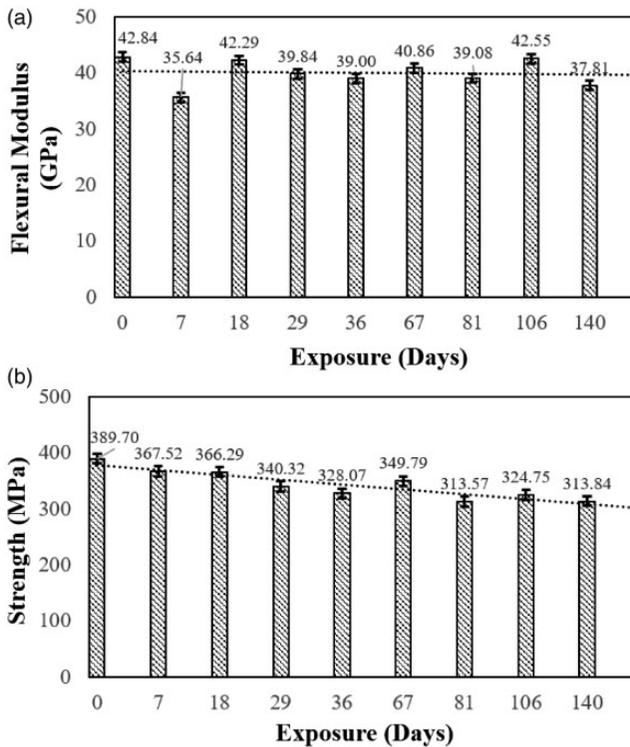


Figure 8. Variation in (a) flexural modulus and (b) flexural strength with respect to exposure time.

Effects of sea water uptake on flexural behavior

Specimens were periodically subjected to three-point bending tests to determine their flexural modulus and strength with respect to the extent of sea water uptake. Figure 8(a) and (b) shows the variation of flexural modulus and strength with respect to exposure time, respectively. The results showed that the exposure to sea water adversely affects the materials' flexural strength, resulting in an ultimate reduction of up to 19.45% for an exposure time of 140 days. This significant reduction in the values of flexural strength of the wet specimens can be attributed to the highly liquid infiltrated zones at the boundaries shown in Figures 6 and 7, causing plasticization of the vinyl ester resin matrix. On the other hand, flexural modulus values appeared to fluctuate with respect to exposure time. It was observed during the sea water exposure study that the concentration of sea water absorption was predominantly at the boundaries of the cross-section. The effective flexural modulus is a quantity influenced by the cross-section, which should show a very small downward trend with increasing degradation concentrated at the periphery of the same. However, water absorption also causes slight swelling of the

Table 4. Flexural modulus and strength values of dry and arctic conditioned samples.

Type	Flexural modulus (GPa)	Flexural strength (MPa)
Dry (23°C)	42.84 ± 0.7	389.70 ± 12
Dry (−60°C)	38.01 ± 0.9	479.65 ± 8

cross-section, which is expected to counteract the effect of peripheral degradation, as accounted in equation (5). Figure 4 shows a schematic of changes observed in the cross-sectional area due to peripheral degradation and swelling, which are competing mechanisms that cause fluctuations in the flexural modulus with sea water exposure.

Effects of Arctic conditions on flexural behavior

Arctic conditioned samples were exposed to −60°C for 60 days and then tested in situ at the same temperature to determine their flexural properties. The results were compared to that of the room temperature dry samples (see Table 4). Arctic conditioned samples manifested an 11.3% decrease in flexural modulus and 23.1% increase in flexural strength with respect to room temperature dry samples.

Failure sites of arctic and dry samples were examined using digital microscopy, where both showed similar shear band failure mode. However, it is observed from Figure 9 that the dry samples manifested significant matrix cracking along with fiber kinking as opposed arctic exposed samples where fiber kinking is dominant. That is, the failure associated with arctic samples appear to be sudden, whereas, the dry samples display a progressive failure. The effects of embrittlement due to arctic exposure is also manifested in the flexural stress–strain responses shown in Figure 10. That is, arctic exposure results in higher flexural strength, but causes catastrophic brittle-type failure in the post peak response. The decrease in flexural modulus can be ascribed to the thermal contraction or shrinkage of the initial cross-sectional area of the specimens, which was stated in Adamson [21]. Equation (5), which was previously defined, accounts for the influence of arctic shrinkage on the effective elastic modulus.

Next, a set of saturated samples obtained from the previous room temperature sorption study were exposed to −60°C for 60 days as well, and were then tested in situ at −60°C to determine their flexural properties. Flexural properties obtained were compared to that of the dry samples, and are summarized in Table 5. The saturated arctic conditioned samples manifested a 2.9% decrease in flexural modulus and an average of 36.2% increase in flexural strength with respect to room temperature dry samples. In this case, the increase in flexural strength can also be attributed to embrittlement. However, a large variation in the flexural strength values is observed in Figure 11. This variation can be attributed to the freezing

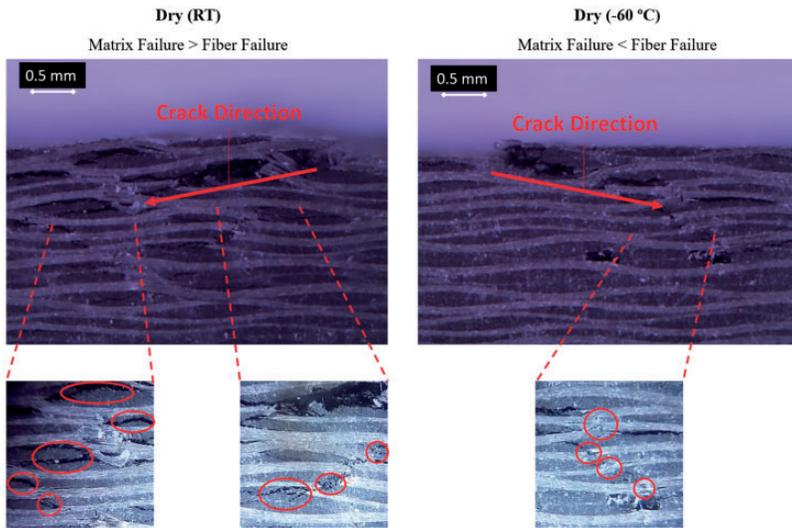


Figure 9. Digital microscopy of failed regions in dry samples tested at room (left) and arctic (right) temperatures.

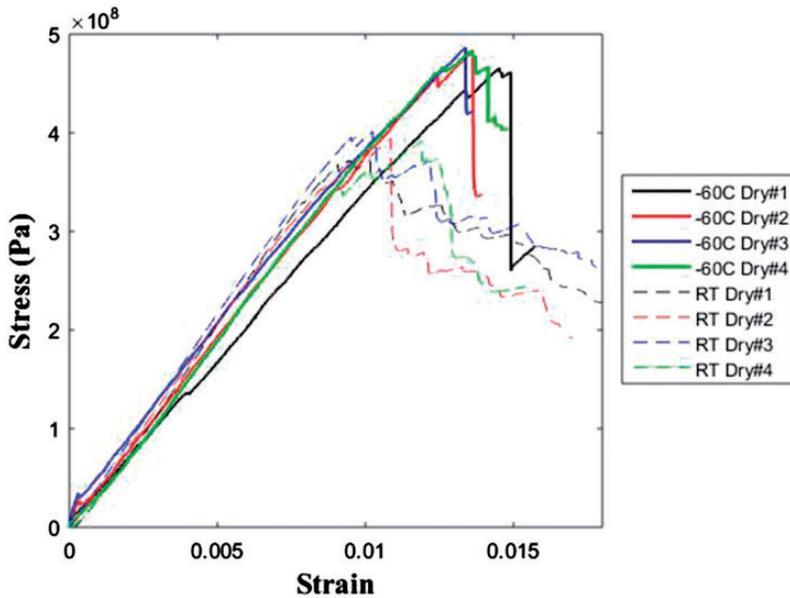


Figure 10. Flexural stress vs. strain response of dry and arctic conditioned samples.

Table 5. Flexural modulus and strength values of dry and combined sea water saturated + arctic conditioned samples.

Type	Flexural modulus (GPa)	Flexural strength (MPa)
Dry (23°C)	42.84 ± 0.7	389.70 ± 12
Saturated (-60°C)	41.58 ± 5.3	530.92 ± 54

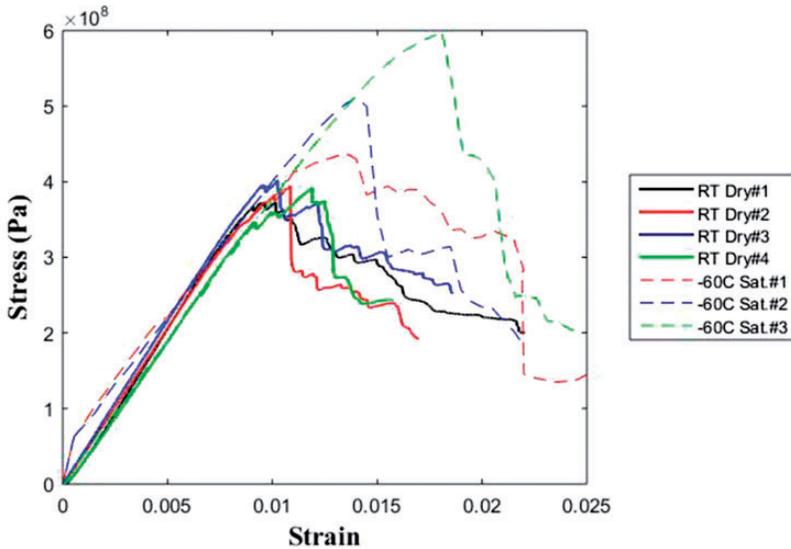


Figure 11. Flexural stress vs. strain response of dry and combined sea water saturated + arctic conditioned samples.

of sea water entrapped within the samples due to saturation. The flexural strength appears to have been affected accordingly due to variation in the amount of sea water absorbed and entrapped during saturation, and the locations within individual samples. Here, the impact of thermal shrinkage due to arctic exposure is being counteracted by the swelling of the cross-section due to sea water absorption. Along with peripheral degradation due to sea water absorption, only marginal change in the modulus was observed in arctic saturated samples as compared to relatively larger reduction as in the case of dry arctic samples. Such behavior was also annotated by Adamson [21].

Digital microscopy of the failed regions in a saturated arctic sample is shown in Figure 12. Both matrix and fiber damage is observed with kink bands occurring along two directions. It should be noted that fiber failure occurred in a catastrophic

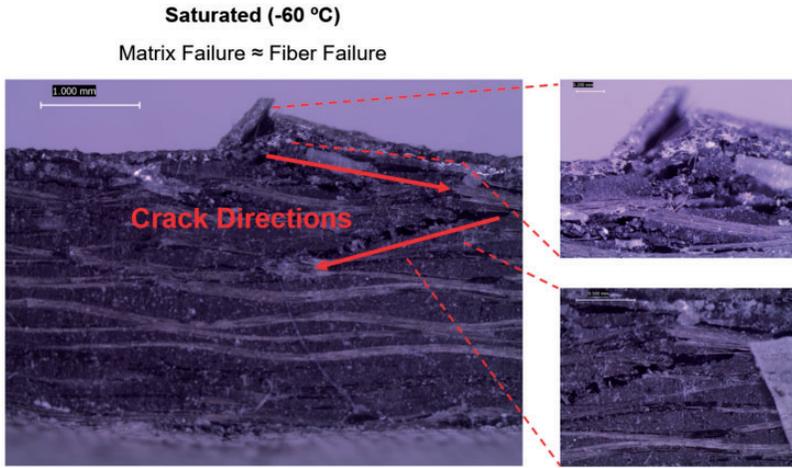


Figure 12. Digital microscopy of failed regions in sea water saturated + arctic temperature conditioned samples.

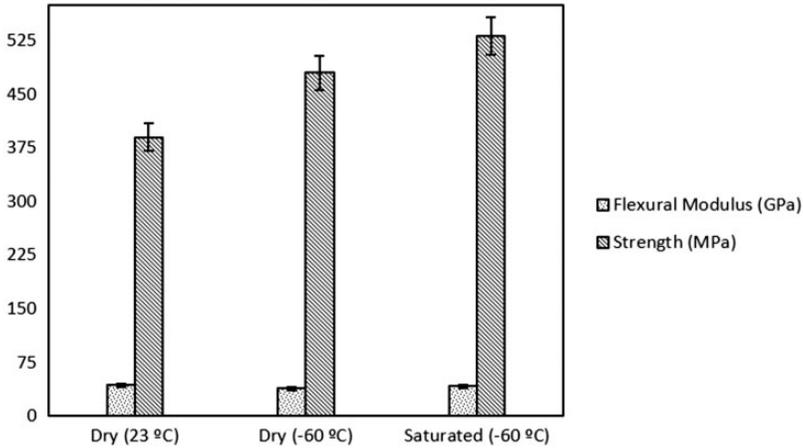


Figure 13. Summary of flexural results between arctic conditioning and room temperature.

manner, displaying a “brushed” or “broomed” effect, which was not present in the previous scenarios.

Figure 13 summarizes the flexural test results for arctic conditioning, both dry and saturated, with respect to the dry room temperature baseline case. The flexural strength has an increasing trend with decreasing environment temperature and sea water saturation. Arctic exposure results in higher flexural strength mainly caused by the embrittlement of the material as stated earlier in this section.

Conclusion

In this paper, the adverse effects of sea water and arctic temperatures on woven carbon fiber/vinyl ester composites were explored in the form of moisture uptake, impact on flexural modulus, strength and structural damage. Flexural response of dry samples at room temperature were considered as baseline behavior for comparing the same for partial and fully saturated samples, dry arctic exposed samples, and saturated arctic exposed samples. Moisture uptake study was conducted first to determine the sea water saturation content ($\approx 1.5\%$) and the diffusivity coefficient ($D \approx 1.81 \times 10^{-6} \text{ mm}^2/\text{s}$) of woven carbon fiber/vinyl ester composites. During this study, sets of samples were periodically subjected to flexural loading to determine the influence of moisture content on their flexural properties. Gradual reduction in flexural strength was observed with increasing moisture content, leading to an ultimate reduction of $\approx 19.45\%$. The flexural modulus appeared to fluctuate with moisture content. The microstructure of dry and saturated samples was compared using SEM, which exhibited a saturated surface with distinctive hue for wet samples as compared to a rough (parched) surface of the dry samples.

Dry and saturated samples were also conditioned and tested in an in situ arctic environment, where both showed an increase in flexural strength due to matrix and fiber embrittlement in the material system caused by low temperature exposure. Dry arctic exposed samples showed an increase of about 23.1% and saturated arctic exposed samples showed an increase of about 36.2% as compared to the baseline dry sample flexural strength. However, the flexural strength of sea water saturated arctic samples appears to show relatively large variations. This can be attributed to the freezing of entrapped sea water in these samples due to saturation. Variation in the quantity and location of sea water entrapped can alter the flexural strength significantly—moisture saturation content shown in this study varied $\pm 0.1\%$ between samples, and a relatively uneven distribution of water concentration was observed through SEM, which alters the failure initiating regions in a specimen. Thereby, influencing the flexural strength values. Further, the influence of sea water saturation at room and arctic temperature has insignificant influence on the flexural modulus due to confinement of sea water only at the boundary regions of the samples. However, the flexural modulus of dry arctic exposed samples showed a decrease in flexural modulus of $\approx 11.3\%$.

To summarize, the research presented in this paper attempts to relate failure modes to the flexural behavior of woven carbon/vinyl ester composites exposed to three key environmental conditions: sea water, arctic temperature and combined sea water/arctic condition. Sea water saturation in general degrades the flexural strength, whereas arctic exposure and combined condition increases the flexural strength. However, arctic and combined environment tend to shift the post peak behavior from progressive to brittle-type failure. Therefore, using these composite in arctic environment may result in catastrophic failure, and special consideration is required while designing critical load bearing components in arctic applications using woven carbon/vinyl ester composites.

Declaration of Conflicting Interests

The author(s) declared no potential conflicts of interest with respect to the research, authorship, and/or publication of this article.

Funding

The author(s) disclosed receipt of the following financial support for the research, authorship, and/or publication of this article: The authors would like to thank Dr Yapa Rajapakse and Dr Asher Rubinstein for their support through the DoD HBCU/MI Basic Research Grant (W911NF-15-1-0430) to conduct the research presented in this paper.

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