Synergistic Effects of Fatigue and Marine Environments on Carbon Fiber Vinyl-Ester Composites

Fiber-reinforced polymer (FRP) composites used in the construction of composite-based civil and military marine crafts are often exposed to aggressive elements that include ultraviolet radiation, moisture, and cyclic loadings. With time, these elements can individually and more so cooperatively degrade the mechanical properties and structural integrity of FRP composites. To assist in increasing the long-term reliability of composite marine crafts, this work experimentally investigates the cooperative damaging effects of ultraviolet (UV), moisture, and cyclic loading on the structural integrity of carbon fiber reinforced vinyl-ester marine composite. Results demonstrate that UV and moisture can synergistically interact with fatigue damage mechanisms and accelerate fatigue damage accumulation. For the considered composite, damage and S–N curve models with minimal fitting constants are proposed. The new models are derived by adapting well-known cumulative fatigue damage models to account for the ability of UV and moisture to accelerate fatigue damaging effects. [DOI: 10.1115/1.4030481]

Keywords: environmental degradation, fatigue, fiber composites, damage, marine composites

1 Introduction

FRP composites are increasingly being considered as practical alternatives to conventional metallic materials for structural load-bearing roles in new classes of naval crafts and warships (e.g., Zumwalt-Class Destroyer [1]). This move toward fiber polymer composites has been motivated by their unique and practical properties such as lower density, higher specific strength and stiffness, reduced radar signature, and better corrosion resistance as compared to metals [2]. These unique properties are projected to assist naval and military industries in their continuous efforts toward developing naval crafts with better maneuverability, fuel economy, structural, and operational performance (e.g., increased payload, stability, and stealth) as well as with decreased capital and maintenance costs [3].

Composites in naval crafts are exposed to harsh marine environmental elements such as UV-radiation, moisture, and cyclic hydrodynamic loads (e.g., due to hull slamming [4,5]). These harsh elements can individually and more so synergistically (e.g., Refs. [6–10]) deleteriously affect the structural properties and load-bearing capacity of fiber reinforced composites, which, in turn, can affect and reduce the long-term structural integrity and reliability of composite naval crafts. To shield composites in marine applications from environmental effects, protective coatings (e.g., epoxy or ester based [11,12]) are almost always utilized. However, these coatings are susceptible to wear, corrosion, environmental degradation and can become damaged due to normal use or inadvertent activities, leaving the protected composites exposed to marine environments. Therefore, even with protective coatings, to ensure the structural integrity of composite naval crafts, it is instrumental to understand and characterize the individual and cooperative deleterious effects of cyclic loadings and harsh marine environmental elements on the mechanical properties of FRP composites.

The individual effect of cyclic loading on composite materials has been extensively studied in recent decades (e.g., Refs. [13,14]) due to their importance to aerospace and infrastructure applications. These efforts provided enlightening experimental based observations and practical cumulative fatigue damage models. Experimental observations highlighted fatigue-based damage processes, mechanisms, and fracture modes. For instance they indicated that the process of fatigue in composite laminates involve the damage mechanisms of fiber breakage, matrix cracking, fiber-matrix decohesion, and delamination [13], where each of these mechanisms can interact with marine environments differently. In addition, experimental observation illustrated that the process of damage accumulation in unidirectional FRP composite laminates passes through three stages that relatively resemble the ones associated with Paris’s law for homogeneous materials. During the first stage extensive but localized matrix cracking and delamination occur. In the second stage, damage accumulates at a slow and steady rate due to stable crack growth along fiber-matrix interfaces. In the final stage, damage accumulates at an exponential and rapid rate due to fiber breakage, coalescence of the localized matrix cracks, and significant fiber-matrix debonding. The third stage culminates with failure and loss of load carrying capacity.

Various cumulative damage models representing fatigue induced damage in unidirectional laminates were proposed and they often used reduction in stiffness or strength with elapsed loading cycles to establish cumulative fatigue damage representative parameters (e.g., Refs. [15–18]). However, reduction in stiffness was found to be a more practical fundamental parameter as its measurement utilizes nondestructive tests, and therefore, a single specimen can be used to describe damage accumulation during fatigue life, which significantly reduces the number of needed tests and minimizes data scatter [18]. Developed cumulative fatigue damage models vary in terms of complexity, but fundamentally they are either defined directly in terms of monitored
residual stiffness or strength [15–18] or in terms of load level and its history [17,19]. Both types of models showed good potential for fitting experimental data, but to capture the classical three fatigue phases, fitting functions with many parameters are needed.

With respect to environmental based degradation, FRP composites have been documented to exhibit degradation in their structural properties and loss in their load carrying capacity due to exposure to UV radiation and/or moisture [20]. These elements are abundantly available in marine environments and their adverse effects can be more severe and harder to anticipate when they act in a combined manner (e.g., Refs. [21–23]).

Moisture diffused into FRP composites can, through plasticization, hydrolysis, and microcracking, damage the matrix and cause fiber-matrix debonding, which reduce the composite’s strength (e.g., for carbon fiber reinforced plastic [24,25]).

UV radiation which is always present as part of the sun spectrum was observed to cause reduction in strength [21,26], increase creep strain [21,27], increase brittleness [28], and introduces surface microcracks due to photo-oxidation and molecular chain scissions [28,29]. These microcracks can act as sources for stress concentrations and promote crack initiation and propagation [28,29]. In addition, microcracks caused by UV radiation can facilitate moisture diffusion as well as moisture induced damage can assist UV radiation in creating microcracks. As UV and moisture accelerate the adverse effects of each other, their synergistic damaging effects are worse than their individual damaging effects [21,27].

In marine environments, UV radiation and moisture can co-exist with cyclic fatigue loads. In such scenarios, the adverse effects of environmental elements and cyclic loadings can interact cooperatively to result in worse damaging effects as compared to the individual effects (e.g., Refs. [6,8,10,30,31]). For instance, it has been observed that the loss of strength and stiffness due to cyclic loadings in aqueous environments is greater than those observed in air [30,31].

Although, it is observed that fatigue loading and marine environments can cooperatively interact, producing accelerated deleterious damaging effects that can lead to unexpected drop in the load carrying capacity of FRP composites, the synergistic effects of fatigue and marine environments remain of the least characterized and understood aspects of composite degradation and failure. So far, models capable of describing the synergistic marine environment-fatigue damaging effects on FRP composites are lacking.

Therefore, this work aims to provide, through experimentation, a better understanding of the deleterious synergistic effects of fatigue and marine environments on fiber reinforced composites as well as establish models with minimal fitting parameters that can describe the loss in mechanical properties due to the considered synergistic effects.

2 Methodology

The implemented approach is purely experimental and is tailored to characterize the loss in the load carrying capacity of FRP composites due to the cooperative interaction between the damaging mechanisms of fatigue and marine environments. The methodology is designed around three components, which are: In-lab accelerated aging under simulated marine environments, cyclic loading of environmentally degraded specimens, and characterization of the loss in mechanical properties due to the combined exposure to fatigue and marine environments.

To effectively characterize the degradation due to marine environments, with or without cyclic loading, tested specimens should be exposed to well-defined, controlled, and measurable marine environmental elements. This requirement disqualifies outdoor based testing (e.g., in seas, lakes or oceans) as a viable option, since long-term outdoor exposure is associated with significant and untraceable (i.e., hard to measure and document) variations in exposure conditions, which can lead to significant data scatter and complicate quantifying the relation between exposure conditions and loss in mechanical properties. Accordingly, this work uses accelerated weathering chambers that can provide controlled exposure to UV radiation, moisture and heat to simulate marine conditions and environmental elements (moisture and UV radiation).

In real-life conditions, exposure to marine elements and cyclic loadings can occur in their simultaneous or sequential manners, where the latter occurs more frequently (e.g., due to day and night cycles, service, and weather patterns). For the damaging mechanisms associated with the different marine elements (UV and moisture) and cyclic loading conditions (load level and fluctuations, frequency) to interact and accelerate each other, the corrosive/erosive elements do not need to concurrently act on the composite. The damaging mechanisms can interact, assist, and accelerate each other even when they are activated in a sequential manner. For instance, long-term exposure of composites to UV radiation (without cyclic loading) can initiate surface microcracks which can act as nuclei from which fatigue cracks can be grown under cyclic loading. These UV initiated and fatigue grown cracks can accelerate water diffusion into the composite, leading to accelerated hydrolysis, plasticization, and loss in mechanical properties.

Composites in real-life can be subjected to endless possible scenarios of sequential exposure to marine elements and cyclic loadings that differ in terms of exposure sequences, periods, frequencies, intensities of marine elements, and load levels, etc. These scenarios result in different damage levels that can be correlated to each other once a fundamental understanding of the interaction between marine elements and cyclic loading is established. Accordingly, for practicality and as the goal of this work is to assist in understanding the fundamental aspects of marine elements-cyclic loading cooperative damaging effects, this work adopts a very basic sequential exposure scenario comprised of two phases; aging phase followed by simple cyclic loading phase. Hence, in this work, the long-term cooperative interaction between the deleterious effects of marine elements and fatigue is explored by investigating the fatigue behavior of environmentally preconditioned specimens (in accelerated weathering chambers).

The aforementioned discussion illustrates the logic behind the adopted methodology, while the following Secs. 2.1, 2.2, and 2.3 describe the methodology and experimental setup in detail.

2.1 Environmental Preconditioning (Degradation) and Marine Environments Simulation. A Q-Lab QUV/se UV Radiation/Condensation accelerated environmental chamber is used to simulate marine environments by exposing specimens to the marine elements: UV radiation and moisture/condensation. This chamber simulates sunlight UV irradiation using fluorescent UV bulbs with a wavelength of 340 nm and simulates condensation (100% humidity) by condensing water vapor on exposed specimens. Both UV exposure and condensation occur at controlled temperatures. UV radiation of 340 nm is chosen as the sun light spectrum passing ozone layers and reaching earth’s surface has UV components distributed within the 290–400 nm band and has an intensity peak around the 340 nm wavelength [32].

The used accelerated weathering chamber offers aging through exposure to UV or condensation; it does not allow for aging specimens using UV radiation and condensation simultaneously. To simulate the combined exposure to moisture and UV (as in real marine environments), specimens were cycled every 3 hrs between UV radiation and condensation exposure modes following ASTM G53 standard. This cyclic exposure method is widely approved and considered as a standard accelerated aging approach (e.g., Refs. [32,33]). Aging continued for a total of 1000 hrs. The UV irradiation cycle had an intensity of 0.6 W/m² and temperature of 60 °C while the condensation cycle had a temperature of 45 °C.

2.2 Materials. The marine composite considered in this work is carbon fiber reinforced vinyl-ester composite. Carbon-
Composite laminates were machined using a diamond wet saw. The unidirectional composite laminates had a nominal thickness of 1.4 mm and were acquired from Graphtek LLC. Specimens were cut into specific sizes of 12.5 mm × 52 mm. The fibers in the specimens were aligned with the width of the specimens (i.e., 12.5 mm long side), while for the latter (i.e., transverse) the fibers were aligned with the length of the specimens (i.e., the 152 mm long side).

2.3 Cyclic Loading and Data Reduction. The implemented cyclic loading strategy was planned considering three factors. First, fatigue phenomenon (solely due to cyclic loading) in FRP composites is, to a great extent, well-characterized, and understood. Therefore, this work aims to draw on the already established understanding and extends it to account for the acceleration in fatigue damage due to the cooperative deleterious interaction of fatigue mechanisms with marine environmental elements. Using existing fatigue models and theories for fiber composites as a platform to establish this work’s goals allows for minimizing the cyclic loading conditions (e.g., different frequencies or loading ratios) that need to be used in experiments. Second, the process of environmental degradation (aging) is time consuming and the accelerated weathering chamber has low throughput. Therefore, aging specimens to perform experiments at a wide range of cyclic loading conditions is prohibitively time expensive. Third, as the specimens representing the unidirectional carbon vinyl-ester laminate are orthotropic, they need to be tested along their two principal directions; thus duplicating the number of tests required. Based on the aforementioned factors (i.e., out of practicality and due to the nature of the objective of this work), the cyclic loading conditions utilized throughout most of this work were limited to two scenarios (loading levels of 62% and 67% ultimate tensile stress (UTS) as explained below). However, a few additional loading conditions were used to test the resulting model. Additional loadings are described in the S–N model section.

Cyclic loading was performed using uniaxial tension configuration on an Instron (1332) servo-hydraulic machine and under load control. Tests were carried out at 62% and 67% UTS loading levels with a minimum to maximum stress ratio of \( R = 0.2 \). UTS along the (0 deg) direction was measured as 1492 MPa and 1387 MPa for virgin and aged specimens, respectively. UTS along the (90 deg) was measured as 60 MPa and 51.4 MPa for virgin and aged samples, respectively. Cyclic loading followed a sine form with a frequency of 5 Hz for samples with (90 deg) fibers direction and 2 Hz for samples with (0 deg) fibers direction. These particular frequencies were selected based on practicality; they are fast enough to run tests quickly but slow enough to produce smooth waveforms and ensure that the heat generated during loading cycles is dissipated (i.e., maintain isothermal conditions). “In addition, due to the significant difference in compliance between the (0) and (90) cases, slightly different frequencies were used: 5 Hz for the (90) case and 2 Hz for the (0) case. For the (0) case, the loading system used was not able to produce smooth sinusoidal waves at 5 Hz, while it was able to produce required forms for the (90) case.” Aluminum tabs were epoxied to the surfaces of specimens at the gripping locations to prevent grips from damaging the specimens. Specimens were gripped with mechanical locking wedge grips.

During cyclic loading, strain was measured using a 25 mm gauge length extensometer, while stress in specimens was computed using the force level measured by the machine’s load cell (100 KN capacity). Stress and strain values were sampled at a rate of 100 samples per second. Cyclic stress–strain data for all tests followed the trends exemplified in Fig. 1. This figure shows the stress–strain curves for the first five cycles and last five cycles for a specimen in as received condition cyclically loaded at 62% UTS. This figure demonstrates that hysteresis is negligible, the specimen is loaded elastically, and that stiffness (slope) is reduced with cyclic loading due to fatigue damage.

Cyclic loading for each loading configuration was performed using a set of three identical specimens in terms of dimensions and pre-exposure conditions. For each specimen, stiffness was computed from the cyclic loading–unloading stress–strain curves and subsequently plotted against elapsed cycle count (fatigue life) as illustrated in Fig. 2. This figure presents the decay in the transverse stiffness with cyclic loading for the as received (i.e., virgin) specimens set tested at 62% UTS load level. Each stiffness versus cycles data set was filtered using a moving average scheme with a subset size of ten cycles. The three stiffness–cycles curves for each set (three identical specimens) were averaged to minimize data scatter, which was within the reasonable scatter range (±3%) as demonstrated in Fig. 2. In this figure, the measured stiffness for the three specimens and the averaged stiffness are very close to each other, almost overlapping. Averaging was performed by finding the arithmetic mean of the three curves at common X-axis values (i.e., number of cycles).

3 Results and Discussion

3.1 Experimental Results: Effect of Environmental Exposure on Fatigue Life. The aforementioned testing strategy (aging followed by cyclic loading) was followed to obtain a total of six sets of averaged fatigue response data (from 18 experiments) that represent the longitudinal and transverse stiffness variation with cyclic loading for aged (environmentally preconditioned) and virgin (as received) specimens. For specimens representing the longitudinal direction, virgin specimens were cyclically loaded at 62% and 67% UTS load levels, while aged specimens were cyclically loaded at 67% UTS load level. For these three cases, the decline in the longitudinal stiffness with elapsed cycles is presented in Figs. 3(a) and 3(b). On the other hand, for the specimens representing the transverse direction, virgin specimens were cyclically loaded at 62% UTS and 67% UTS load levels while aged specimens were cyclically loaded at 62% UTS load level. Transverse stiffness data for the latter three loading conditions is presented in Figs. 4(a) and 4(b).

Figures 3 and 4 demonstrate that the moduli of the laminates along both principal directions are reduced with cyclic fatigue loading, but the reduction rate and its behavior is affected by the loading direction and preconditioning (aging). For the transverse
modulus, the reduction in stiffness for all loading levels and test conditions follows a two stage behavior that commences with a steep decrease during the first few loading cycles and later transitions to follow a slow and steady decreasing rate until failure. On the other hand, the longitudinal modulus behavior due to cyclic loading exhibits three phases. In the first two phases the behavior mimics the one observed for the transverse case, where stiffness reduction commences at a steep rate and subsequently, in the second phase, transitions to a steady rate. However, in the third phase, which is not observed in the transverse case, the decrease rate in the longitudinal stiffness changes continuously in a rapidly accelerating manner until the onset of failure.

The trends observed in Figs. 3 and 4 represent the process of damage accumulation in the considered unidirectional carbon fiber vinyl-ester laminate (virgin and aged). The longitudinal moduli behaviors (Fig. 3), in terms of shape, resemble the three stages of the Paris law. The variation in the stiffness reduction rate between the different stages is attributed to the different damage mechanisms active in each stage. During the first stage extensive but localized matrix cracking and fiber-matrix delamination occur, while in the second stage damage is contributed to steady and stable growth in cracks within the matrix and along fiber-matrix interfaces. On the other hand, in the final stage, damage accumulates at an exponential and rapid rate due to fiber breakage, coalescence of the localized matrix cracks and significant fiber-matrix debonding. Most of the aforementioned damage mechanisms are also associated with damage accumulation and reduction in the stiffness of the transversely loaded specimens. However, as load is perpendicular to the fibers in the transversely loaded specimens, fibers remain intact and damage accumulation tends to be confined to the matrix and along the fiber-matrix interfaces.

Fig. 2 Stiffness (transverse modulus) versus elapsed loading cycles for three as received specimens cyclically loaded at 62% UTS, showing the reasonable level of data scatter associated with identical testing conditions.

Fig. 3 Longitudinal stiffness variation with cyclic loading for (a) virgin specimen cyclically loaded at 62% UTS and (b) virgin and aged specimens cyclically loaded at 67% UTS.

Fig. 4 Transverse stiffness variation with cyclic loading for (a) virgin specimen cyclically loaded at 67% UTS and (b) virgin and aged specimens cyclically loaded at 62% UTS.
interfaces. Consequently, unlike for the longitudinally loaded specimens, the stiffness reduction pattern of the transversely loaded specimens does not exhibit the third stage that exhibits rapidly increasing stiffness reduction rate.

Just before the onset of failure (last cycle) due to the accumulation of fatigue damage in aged and virgin specimens, Figs. 3 and 4 demonstrate that stiffness, although can be significantly reduced, does not diminish and approach zero. Based on Figs. 3 and 4, the maximum reduction in stiffness (at failure as compared to initial stiffness) does not exceed 14%. Longitudinal and transverse modulus at failure for all considered loading conditions did not cross below the 106 GPa and 7.3 GPa marks, respectively.

The effect of aging and environmental elements on the fatigue response can be seen in Figs. 3(b) and 4(b), for the longitudinally and transversely loaded specimens, respectively. Aging reduces the initial stiffness and more so the ultimate strength of the carbon fiber reinforced vinyl-ester laminates. Due to the reduced initial stiffness, the fatigue behavior of the aged specimens in both Figs. 3(b) and 4(b) appears shifted down along the stiffness axis.

For the longitudinal behavior, Fig. 3(b) demonstrates that during the first few hundreds cycles (first phase) the reduction rate in stiffness for the aged specimens is steeper than that of the virgin specimens. However, in the second stage, the stiffness reduction rate for both aged and virgin specimens is almost the same. It should be noted that although tests on virgin and aged specimens are performed at the same %UTS level, the actual applied force on the aged specimen is lower since the UTS value is lower by 7% as compared to the UTS of the virgin specimens. So even with the slightly less applied force, fatigue damage in the aged specimens is more pronounced in the first stage. Beyond the first stage, aging does not seem to cooperatively interact and amplify fatigue damage due to cyclic loading. In the last stage, Fig. 3(b) shows that virgin specimens have shorter life and exhibit a slightly steeper stiffness reduction rate as compared to the aged specimens. The latter observation is attributed to two factors. First, toward the end of fatigue life, aging damage due to preconditioning has minimal effects on fatigue behavior (as will be discussed in the following paragraphs). Second, aged specimens are loaded by a slightly smaller force (7% less) as compared to virgin specimens.

For the transversely loaded specimens, the effect of aging on the fatigue response, as seen in Fig. 4(b), follows to a great extent the trends observed for the longitudinally loaded specimens. The cooperative interaction of aging with fatigue is only pronounced during the first phase of fatigue life. In addition, as observed in the longitudinal case, toward the end of fatigue life the virgin specimens exhibit slightly higher stiffness reduction rate and shorter fatigue life than those of their aged counterparts.

The cooperative interaction between aging and fatigue behavior is observed to be limited to the first few hundred loading cycles (i.e., the early stages of the first phase). To understand the causes of this observation one needs to follow the damaging effects of aging on the physical, chemical, and mechanical properties of the composite laminates, and understand the interactions between these damaging effects with cyclic loading. Aging using UV radiation and moisture causes surface microcracks, lessens the stiffness and strength of the matrix, and negatively affects the fiber-matrix interfaces. Aging-induced microcracks and defects provide sources for stress concentrations and nuclei from which fatigue cracks can grow. Accordingly, defects due to aging facilitate fatigue damage initiation which translates to the increase in the stiffness reduction rate during the early stages of the first phase as seen in Figs. 3(b) and 4(b). However, once fatigue cracks start to grow steadily and aging induced defects cease to nucleate new fatigue cracks, the effect of the aging induced defects diminishes. Therefore, the cooperative interaction between aging and fatigue is limited to the first few hundred loading cycles (i.e., fatigue damage initiation phase).

Moreover, the combined damage due to aging and fatigue is observed to decrease both the longitudinal and transverse stiffness by no more than 10% as compared to stiffness reduction associated with fatigue only. To extract meaningful insights from this observation, one needs to first investigate the spatial distribution of damage in the aged specimens. Carbon fiber reinforced vinyl-ester laminates are considered as the preferred marine composite since they tolerate moisture and UV radiation. They still exhibit some damage due to long exposure to marine elements, but this damage does not penetrate deep into the composite. For the specimens aged in this work, aging-induced damage after 1000 hrs of exposure was observed using scanning electron microscope (SEM) microscopy to be limited to the surfaces of the aged specimens and was confined to 30–50 μm thick surface layers as seen in Fig. 5. Given that specimens are 1.4 mm thick, in all cases at least 90% (by volume) of each of the aged specimens is unaffected by aging, which explains why aging worsened the damaging effect of fatigue on the laminate property by roughly 10% (see Figs. 3(b) and 4(b)). However, considering that the laminate used has eight identical plies, the aging-induced damage proliferated through at least 30% of the exposed plies. Thus, at the ply level, the exposed plies will exhibit severe cooperative interaction between aging and fatigue.

It should be noted that the observed cooperative effect depends greatly on the polymer or matrix material. So for other composites, the cooperative damaging effects of marine elements and cyclic loadings may be significantly more pronounced.

Also, if the aging-fatigue cycle repeats as in real-life, the subsequent aging cycles will provide new microcracks and nuclei for fatigue cracks; therefore, it can be expected that the cooperative damaging effects of aging and fatigue will be much more pronounced.

3.2 Cumulative Damage Modeling. In this section, a general phenomenological model capable of capturing the fatigue behavior of the virgin and aged carbon fiber vinyl-ester laminates is developed.

Cyclic fatigue loading of composite laminates, in general, leads to damage accumulation through known major mechanisms that include matrix cracking, fiber-matrix debonding, and fiber breakage. These mechanisms simultaneously, but to different extents, affect the macroscopic structural properties (e.g., stiffness and load bearing capacity) of composite laminates. However, as the individual roles of the different fatigue damage mechanisms are hard to separate and treat on an individual basis, their effects are often treated in a combined manner by characterizing changes in the macroscopic properties of composites with cyclic loadings.

To mathematically represent the effect of fatigue mechanisms and simplify conducting comparisons, cumulative damage indices have been proposed (e.g., Refs. [8,35]). These indices translate reduction in stiffness or strength into variables that range from zero (i.e., no damage) to one (i.e., failure). This work adopts this approach to damage accumulation.
damage index approach and uses the following common stiffness based damage index as a basis for modeling cumulative damage:

$$D = 1 - \frac{E(N)}{E_0}$$ (1)

where $E_0$ denotes the initial stiffness of the composite and $E(N)$ is the stiffness at the $N$th cycle. Experimental results in Figs. 3 and 4 show that composite laminates exhibit substantial stiffness during the final loading cycles; therefore, based on Eq. (1) the cumulative damage value will be far from one at failure. To obtain a damage index that always yields a value in the range of 0–1, Eq. (1) is normalized using the stiffness measured at failure, yielding the following normalized cumulative damage index:

$$\bar{D} = 1 - \frac{E(N) - E(N_f)}{E_0 - E(N_f)}$$ (2)

where $E(N_f)$ is the modulus at the final cycle. Damage index defined by Eq. (2) implicitly represents the physically measured damage through the experimentally measured stiffness.

Damage accumulation with elapsed loading cycles ($N$), which is the core of cumulative damage models, is defined here phenomenologically. To reach to a mathematical representation that fits the fatigue behavior of carbon fiber vinyl-ester laminates, results derived from the implementation presented in Ref. [37] but modifies it to accommodate the stiffness based damage-free approach, and by pursuing a high coefficient of determination ($r^2 \geq 0.99$) fitting constants of the model can be related to loading trends similar to the ones in Fig. 6, simple power laws and exponential forms whose fitting constants can be related to loading conditions and parameters. Out of these proposed forms by Tang et al. [37], the exponential type was observed to fit the damage behavior (i.e., curves presented in Fig. 6) rather well. Accordingly, this work adopts the exponential based representation from Ref. [37] but modifies it to accommodate the stiffness based damage index (Eq. (2)) as illustrated in the following equation:

$$\frac{d\bar{D}}{d\log N} = \bar{C} e^{\bar{C}D}$$ (3)

where $n$ and $\bar{C}$ are fitting constants. The impetus behind using this rate of change in damage is the behavior of the slopes (i.e., $d\bar{D}/d\log N$) of the curves presented in Fig. 6, which at the early stages of fatigue life are small but rapidly (exponentially) increase with subsequent loading cycles. According to Ref. [37], $\bar{C}$ in Eq. (3) is a function of maximum stress, stress amplitude, frequency, temperature, and other fatigue accelerating/decelerating factors. As all experiments in this work are performed at low frequency ($f \leq 5$) and at room temperature, the effect of frequency and temperature can be ignored. Accordingly, and following Ref. [37], $\bar{C}$ can be expressed as

$$\bar{C} = \bar{C}(f, T, \sigma_{\text{max}}, \sigma_{\text{amp}}) \approx \bar{C}(\sigma_{\text{max}}, \sigma_{\text{amp}}) = \frac{C_m^{\sigma_{\text{max}}} \sigma_{\text{amp}}^{m}}{\sigma_{\text{ult}}^2}$$ (4)

where $\sigma_{\text{max}}$ is maximum stress, $\sigma_{\text{min}}$ is minimum stress, and $\sigma_{\text{amp}} = (\sigma_{\text{max}} - \sigma_{\text{min}})$. $R$ is the ratio of minimum to maximum stress, $\sigma_{\text{ult}}$ is the ultimate strength of the material, and $C$ and $m$ are constants. It should be noted that though this work adopts the basic phenomenological damage model (i.e., Eq. (3)) from Ref. [37], its implementation in this work (as follows) is very different from the implementation presented in Ref. [37]. This deviation is motivated by the difference in the objectives of this work as compared to those of Tang et al. [37]. Since here the aim is to extend the phenomenological model to establish stiffness-based $S$–$N$ curves. To proceed toward developing a stiffness-based $S$–$N$ curve for the carbon fiber vinyl-ester laminate, Eq. (3) is combined with Eq. (4) and integrated with respect to $D$ to yield

$$\frac{-1}{n} e^{-nD} = C \left(\frac{\sigma_{\text{max}}^2 (1 - R)}{\sigma_{\text{ult}}^2} \right)^m \log N + A$$ (5)

where $A$ is an integration constant and can be determined using the condition $D = 0$ and $N = 1$ (i.e., specimens are initially fatigue damage free). After determining the constant $A$, Eq. (5) can be written after manipulation as

$$D = \frac{-1}{n} \ln \left(1 - Cn \left(\frac{\sigma_{\text{max}}^2 (1 - R)}{\sigma_{\text{ult}}^2} \right)^m \log N \right)$$ (6)

such that $n$, $m$, and $C$ are constants and can be determined by fitting Eq. (6) to experimental data, such as the data presented in Fig. 6. Finally, by invoking the condition that at the onset of failure (at maximum number of cycles $N_0$) the value of $D$ approaches one, Eq. (6) can be rewritten to define the following stiffness-based $S$–$N$ relation (i.e., $S$–$N$ curve):

$$\sigma_{\text{max}} = \frac{\sigma_{\text{ult}}}{\sqrt{1 - R}} \left(\frac{1 - e^{-nD}}{Cn \log N_i} \right)^{\frac{1}{m}}$$ (7)

Above equation constitutes a simple fatigue damage model that has only three fitting parameters ($n$, $m$, $C$). Although this model did not consider the effects of parameters such as loading frequency and temperature, the effect of these parameters can be incorporated through Eq. (4).

To validate the derived $S$–$N$ model (Eq. (7)) and evaluate its ability to capture fatigue damage evolution and fatigue life of the carbon fiber vinyl-ester laminates used in this work, a two stage validation process is followed. First, the constants of the model are determined by calibrating against one of the measured fatigue behavior data sets. Subsequently, the model in conjunction with the determined fitting constants is used to predict other measured data sets at different loading conditions. This approach was applied to both virgin and aged specimens and for both longitudinal and transverse directions.

The fitting process to determine the constants of the $S$–$N$ model was performed using the nonlinear square fit built-in function in MATLAB. The quality of the fits was ensured using visual inspection and by pursuing a high coefficient of determination ($r^2 \geq 0.99$). Fitting constants of the $S$–$N$ model for the longitudinal and transverse cases, for aged and virgin specimens are included in Table 1.

![Fig. 6 Cumulative damage behavior in terms of principal moduli for the virgin specimens](Image)
To illustrate the ability of the model to capture and predict the experimentally measured fatigue evolution data, comparisons between the experimental and predicted data for the virgin specimens are included here. For the virgin specimens, the model was calibrated against the 62% UTS loading case (see Table 1 for fitting constants) and used to predict the fatigue evolution behavior at 67% loading case. Results from the predicted and measured fatigue evolution behavior are presented in Fig. 7, for both longitudinal and transverse cases. This figure is plotted in terms of damage and cycle ratio (i.e., $N/N_f$) to enable plotting multiple curves on the same plot, particularly as the fatigue life for the different loading conditions varies significantly. The comparison in this figure demonstrates that the model effectively captures the experimental result.

To verify the ability of the model ($S$–$N$ curve) to capture and predict the fatigue life for the virgin specimens, predictions from Eq. (7) is compared to experimental data representing fatigue life and failure stress ($r_{max}$ and $N_f$) in Fig. 8(a). This figure, for both longitudinal and transverse directions, confirms the efficacy and accuracy of Eq. (7) as a phenomenological $S$–$N$ curve for the carbon fiber vinyl-ester laminates used in this work. In this figure, experimental data points at 100%, 67%, and 62% UTS loading levels for longitudinal and 100%, 83%, 73%, 62%, and 59% UTS loading levels for transverse direction were included and they are inline with model predictions.

In cases where environmental conditions accelerate or decelerates fatigue damage accumulation, this model, Eq. (7), can be adapted to accommodate sensitivity to the environment through the fitting parameters, $n$, $m$, and $C$. To evaluate the effectiveness of the model in capturing the fatigue life of the aged specimens, the model was first calibrated against the fatigue evolution response of the aged specimens, measured at 67% UTS and 62% UTS for the longitudinal and transverse directions, respectively. Subsequently, the calibrated model was used to predict the fatigue life for aged specimens tested at other loading conditions (i.e., 62% UTS for longitudinal and 77%, 69%, and 64% UTS for transverse direction). Resulting fitting constants for the aged specimens are included in Table 1, while the comparison between the measured and predicted fatigue life for the aged specimens is included in Fig. 8(b). This figure demonstrates that the model is very effective in capturing the measured fatigue life of the aged specimens.

It should be noted that the aforementioned results and discussion as well as the developed fatigue damage model are only applicable to unidirectional laminates and describe the behavior under pure uniaxial loading along one of the two material principal directions. For cases, with combined loading or off-axis loading, the biaxial state of stress in the composite can result in a fatigue behavior that differs from the one discussed above.

### 4 Conclusion

The fatigue behavior of a marine composite (unidirectional carbon fiber reinforced vinyl-ester) was investigated at cyclic load
levels approaching the maximum design loads for the marine industry (i.e., when design is based on factors of safety between 1.5 and 2.0), and in conditions that mimic aging due to long-term exposure to marine environments (i.e., combined exposure to UV radiation and moisture). Based on the obtained results, the following is concluded:

- Fatigue damage evolution in terms of stiffness degradation with cyclic loading is very sensitive to loading direction. Along the longitudinal direction (i.e., fiber direction), fatigue behavior mimics Paris’s law and exhibits the common three stage fatigue behavior. Such that the damage accumulation rate commences at a steep rate but transforms into a steady rate in the second phase, while in the third phase which ends in failure damage accumulates at a rapidly (exponential) rate. On the other hand, for loading along the transverse direction (perpendicular to the fibers), fatigue damage evolution follows a two stage pattern and does not exhibit the third phase observed in the longitudinal case.

- Aging of specimens in marine environments causes a minor reduction in stiffness but a significant reduction in ultimate tensile strength (7% and 14%, for the longitudinal and transverse cases, respectively). Aging causes surface microcracks and degrades the mechanical properties in a thin surface layer 20–50 μm thick. In terms of fatigue behavior, aging and associated defects provide additional nuclei from which fatigue accumulation can start. Therefore, aging facilitates fatigue initiation and is pronounced only in the first stages of fatigue damage evolution.

In addition, a stiffness based fatigue evolution model and S–N curve are developed to describe the fatigue behavior and fatigue life of carbon fiber reinforced vinyl-ester composites. The model employs an exponential representation to capture the rate of damage accumulation and adapts it to define an S–N curve with minimal fitting parameters (three parameters). The developed model effectively captured the experimental fatigue evolution data and fatigue life for both loading directions (longitudinal and transverse) and for both test conditions (vigin and aged). Moreover, the S–N model curve followed the experimental trends closely and showed that it can be used to predict fatigue life for aged and virgin specimens. The models are strictly developed for unidirectional laminates, but they can be utilized with a wide range of composites other than carbon fiber reinforced vinyl ester composites.

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