1 Introduction

Fiber reinforced composite materials are increasingly used in high-performance lightweight applications. Due to their excellent mass specific properties, their fields of use reach from aircraft and wind energy applications all the way to the automotive industry. Their wide-spread application raises questions of safety and reliability of composite materials. Very strict airworthiness certification regulations demand a no-damage-growth design strategy for civil composite aircraft structures, i.e. damages which are below the barely visible threshold must not grow in service environmental loading conditions. Other than for metallic primary aircraft structures, the fatigue behavior is regarded very conservatively due to a lack in precise and robust design methods which is also due to the common belief that composites are not prone to fatigue. This, however, is only true for low service loadings so the current no-damage growth strategy can only be achieved by applying very strict strain limits to the structure in the design phase leading to a loss of lightweight potential.

To overcome the uncertainties regarding the fatigue behavior of composite materials and to further make use of their great lightweight potential, extensive research both experimentally and numerically is needed. In this work, the phenomenology of the fatigue behavior of fiber reinforced composites is presented in short. The preparation of experimental investigations is described presenting a specially designed test rig to perform four-point-bending fatigue tests which will be used to further investigate the damage mechanisms observed by other researchers. In addition, the various numerical approaches to model the fatigue behavior will be presented. Conclusions regarding the applicability of these models to certain problems will be drawn and a new approach to model the fatigue behavior on the micro-level will be presented. This includes an energy-based criterion for neat resin damage as well as a numerically efficient cycle jump algorithm which uses damage extrapolation for skipping cycles with low damage gradients.

2 Composite Fatigue Phenomenology on the Macro Scale

Similarly to the static load case, the fatigue behavior of fiber reinforced polymers is highly complex and driven by multiple damage mechanisms. The observable damage mechanisms include matrix cracking, delaminations, fiber-matrix debonding, and fiber breakage. The dominating damage mechanism depends on many parameters like reinforcement type, layup, and the load amplitude, i.e. a higher cyclic load can induce different damage mechanisms than a lower load [1]. Compared to metallic materials, the fatigue damage in composite materials is not isolated in a single crack but grows at multiple locations until damage coalescence leads to final failure [2]. In particular matrix damage starts very early leading to a gradual degradation in stiffness and strength, whereas metals barely suffer from degrading properties during fatigue life [1]. Usually assumed to be a matrix dominated problem, Lorenzo and Hahn [3] point out that the governing damage mechanism in unidirectional materials highly depends on the material of the constituents and on the loading amplitude, i.e. low or high cycle fatigue. For unidirectional materials, high cyclic stresses result in a fiber dominated failure whereas low cyclic stresses induce matrix dominated failure mechanisms. However, for most cases, i.e. for arbitrary layups, fatigue damage will be matrix dominated originating...
in the off-axis plies as shown by Hosoi et al. for quasi-isotropic layups [4]. Glassy state polymers like epoxy resins used as matrix materials in fiber reinforced composites have time-dependent, i.e. viscoelastic properties. Miyano et al. [5] and EPAarachchi and Clausen [6] point out that dependency on load frequency, load level, as well as temperature can be traced back to viscoelastic material properties. Also, the specimen temperature increase for different loading frequencies is mainly due to viscoelastic properties [7].

Besides the obvious influences of fatigue behavior of composites like fiber material, matrix material, layup, type of reinforcement (woven fabric, unidirectional material, etc.), also experimental testing boundary conditions have an influence.

**Frequency influence:** Composites are highly sensitive to the testing frequency [8-11] since a change in frequency influences the internal heat generation in the composite. However, no general rules like higher frequencies cause earlier fatigue failure can be drawn. Barron et al. [11] report, for instance, that a rise in frequency from 5 Hz to 10 Hz sharply reduces the life span; however a further increase to 20 Hz results in slightly increased lifetime.

**Mean stress influence:** Cyclic loading with positive stress ratios results in a phenomenon called mean stress creep [12], which is accounted for empirically in many numerical models, e.g. [13, 14].

**Stress ratio influence:** According to Kadi and Ellyin [15] as well as Mandell and Meier [10], the stress ratio of the cyclic loading plays an important role. Within the tension-tension-range of the stress ratio R, the fatigue life increases with increasing stress ratio, whereas within the compression-compression range, increasing the magnitude of R decreases the fatigue life. For simplification reasons, a stress ratio of 0 or 0.1, i.e. tension-tension loading with small or zero lower stress, is used for many experimental fatigue investigations, even though this load case is very rare in reality [1].

Other fatigue performance influences are environmental conditions like moisture absorption [16] and temperature [5], as well as sensitivity to voids and other manufacturing defects [17, 18].

### 3 Experimental Fatigue Testing

In addition to the high number of specimens needed to characterize a composite material sufficiently for the static load cases, various time- and history-dependent parameters tremendously increase the problem size. Therefore, efficient testing is a key concern in fatigue research. Since the material needs to be studied at different load levels at relatively low frequencies to avoid thermal damage, the testing procedure is quite extensive. In addition, when dealing with fiber reinforced composite specimen, deviations in manufacturing parameters and defects in the heterogeneous microstructure result in high scatter of the test results which is further compounded by the high sensitivity of fatigue damage to micro-scale damage like voids [17, 18]. For both cost and time efficient testing, a test rig has been designed which allows parallel displacement controlled testing of six four-point-bending fatigue specimens. An electric motor drives an adjustable eccentric tappet which is connected to the outer support of the specimen. Due to the construction restriction of the simple design, the load can only be applied displacement driven, not load driven as is the case with classical testing procedures used to obtain S-N-curves. Also, the displacement cannot be readjusted once the test is running, since the eccentric tappet is set prior to the test. As described above, a stress ratio of 0 (tension-tension) results in mean stress creep so that both the displacement and the load level cannot be held constant throughout the test with a classic four-point-bending support as the displacement amplitude decreases with further testing, fig. 1.
EXPERIMENTAL STUDY AND MULTISCALE NUMERICAL DESCRIPTION OF THE FATIGUE BEHAVIOR OF FIBER REINFORCED POLYMERS

4 Numerical Description of the Fatigue Behavior

Many researchers have proposed numerical models to describe the various effects of cyclic loading in composite materials [19-24]. Degrieck and Van Paepegem [1] categorize three major groups of numerical fatigue models: Fatigue life models allow for a rather rough life-time prediction. Usually based on S-N-data, failure criteria similar to the static load cases are utilized to make global life time predictions. The advantages of these models are their simplicity and efficiency both numerically and experimentally, as some of them need very little experimental data. However, similar to static failure criteria, progressive and developing damage is not considered so the observed stiffness degradation over the lifetime cannot be predicted. The indication of “failure” at a material point also indicates local failure only; a global failure prediction is difficult to obtain with these models.

The second category of fatigue models are residual stiffness/strength models which are usually empirical descriptions of the mechanical property degradation during fatigue life. This allows for a continuous degradation of properties throughout the analysis. In combination with simple failure criteria, the effect of fatigue loading can be simulated quite accurately and efficiently given that the experimental data are sufficient. Due to their empirical nature the extension to different material configurations and layups is difficult and related to quite expensive experimental effort. Also the underlying failure mechanisms are not captured, only their effect on the mechanical properties is modeled. If different failure mechanisms are activated by a change in boundary conditions, the model might unintentionally be used beyond the validity boundaries leading to erroneous results.

Progressive damage models are physically based and describe the actual failure mechanisms and often focus on one particular mechanism to determine the effect of, for instance, matrix cracks on the transverse stiffness of the ply or determine the growth of the damage itself, for instance delamination growth. Their physical basis makes them more expandable to other layups or materials than purely empirical models. Even though these models capture certain damage mechanisms in detail, damage initiation and damage interaction cannot be accurately captured on the macro scale.

From the fatigue phenomenology it can be summarized that damage under cyclic loading:

- Starts very early in the loading history and at the micro scale
• Is matrix dominated for most problems
• Leads to continuous degradation of laminate properties (stiffness and strength) due to micro damage and matrix degradation
• Is sensitive to interfaces like fibers and voids

Since all reported damage mechanisms and fatigue damage observations arise at the micro-scale, it can be concluded that an accurate description of the constituents in the micro-scale yields fundamental and accurate results for the macro-scale and thus of the whole composite. Furthermore, the high number of failure criteria and fatigue models for composite materials developed in the past indicate that the macro-scale is limited in accuracy so that the best macro-scale material model cannot fully capture the range and complexity of active damage mechanisms governing the micro-scale driven fatigue behavior. Clearly a micro-scale problem, damage initiation in the composite is the weak point in many fatigue models so accurate evaluation of it demands for a detailed micro-model.

4.1 Modeling the Micro-Scale

Heterogeneous, periodic micro structures like fiber reinforced composites can be modeled using representative volume elements (RVE) [25-27]. The prerequisite for this is the validity of scale separation, i.e. the length scale of the RVE is small enough to represent a point on the macro scale. However, the RVE must be large enough to contain enough information to be representative in order to yield accurate effective properties. Trias et al. [28] quantify this condition with the RVE size minimum being at least 4 times the size of the largest inclusion, i.e. the fiber diameter, for a carbon fiber reinforced composite. The effective stresses and strains are formulated in terms of volume averages over the entire RVE volume. Furthermore, the Hill-Mandel condition must be fulfilled which demands the consistency of energies calculated from effective measures with the energies calculated from micro scale measure. Using periodic boundary conditions, this principle is fulfilled a priori.

McCarthy and Vaughan [25] present a tool to create a 2D statistically RVE of a fiber reinforced composite using the nearest neighbor algorithm. On this basis, a tool for ABAQUS has been developed which allows the definition of an extended 3D model. Similarly to McCarthy’s and Vaughan’s model, the fibers are placed periodically in the RVE. In contrast to the original model, though, the distribution is artificially random to be able to produce RVE with varying fiber volume fractions. Different algorithms are utilized for that including the hard core method and the nearest neighbor algorithm with each being more suitable for low and high fiber volume fractions, respectively. The fiber volume fraction can be predefined and is usually matched within a ± 0.25 % tolerance.

Cohesive COH3D8-Elements are used to model the fiber-matrix interface since fiber-matrix debonding is an important damage initiation mechanism. Also, since the influence of internal interfaces is to be investigated, using an imperfect interface modeling technique is sensible.

The tool imposes periodic boundary constraints which are used to introduce the full 3D macro scale displacement gradient

$$H = \frac{\partial u(X, t)}{\partial X}$$  \hspace{1cm} (1)

where \(X\) denotes the position vector of the current material point, \(t\) the time and \(u\) the displacement field.

For improved usability, a graphical user interface (GUI) has been developed which is completely embedded in the software using ABAQUS’ Python scripting interface capabilities. Within the GUI, the distribution algorithm, material parameters, element sizes and boundary conditions can be defined in tabs. Fig. 3 shows the GUI as well as an example of a resulting 3D RVE. The GUI also allows the definition of cyclic loading which is explained in further detail in the next section.

The material models used for the three constituents are different for static and cyclic loading. For static load cases, the matrix material is modeled elastoplastic using Mohr-Coulomb plasticity similar to the approach of McCarthy and Vaughan [25]. The fiber matrix interface uses the bilinear traction-separation law implemented in ABAQUS with the according material parameters. For both static and cyclic load cases, the fiber material is modeled ideally elastic without degradation. As for the cyclic load case, the matrix material is modeled
EXPERIMENTAL STUDY AND MULTISCALE NUMERICAL DESCRIPTION OF THE FATIGUE BEHAVIOR OF FIBER REINFORCED POLYMERS

Table 1: Material parameters used for the static load case (with * GPa, ** MPa, † J/m², ‡ GPa/m)

<table>
<thead>
<tr>
<th>Fiber</th>
<th>Matrix</th>
<th>Interface</th>
</tr>
</thead>
<tbody>
<tr>
<td>$E_{11}$</td>
<td>238.0*</td>
<td>$E$</td>
</tr>
<tr>
<td>$E_{22} = E_{33}$</td>
<td>28.0*</td>
<td>$ν$</td>
</tr>
<tr>
<td>$ν_{12}$</td>
<td>0.23</td>
<td>$Φ_f$</td>
</tr>
<tr>
<td>$ν_{13}$</td>
<td>0.03</td>
<td>$Ψ$</td>
</tr>
<tr>
<td>$ν_{23}$</td>
<td>0.33</td>
<td></td>
</tr>
<tr>
<td>$G_{12} = G_{13}$</td>
<td>24.0*</td>
<td></td>
</tr>
<tr>
<td>$G_{23}$</td>
<td>7.2*</td>
<td></td>
</tr>
</tbody>
</table>

calculated from the interface thickness $t$ and the Young’s modulus $E$ of the matrix via

$$K = \frac{E}{t} = \frac{3.63 \text{ GPa}}{0.165 \mu m} = 22 \cdot 10^6 \text{ GPa/m} \quad (2)$$

The interface thickness is chosen arbitrarily and is equal to 2.5 % of the fiber diameter of 6.5 µm which is used here.

4.2 Continuous Degradation Procedure

The continuous degradation material model used here is a residual stiffness model: Note that the residual stiffness models mentioned in section 2 deal with composite materials on the macro-scale. The model used here is used for the isotropic matrix and interface material on the micro-scale. For low cycle fatigue in isotropic materials, ABAQUS features an approach based on the accumulated elastoplastic hysteresis energy. To accumulate this type of energy, the material has to be loaded beyond the yield criterion and also unloaded again beyond the yield criterion. The degradation is carried out using an isotropic $(1 − D)$ damage approach where $D$ is the damage variable. The damage initiation criterion is

$$N > N_0 = c_1 Δw^{c_2} \quad (3)$$

with $N$ being the current cycle number, $Δw$ the accumulated elastoplastic hysteresis energy, and $c_i$ material parameters. The damage evolution follows a Paris law [29] like equation

(a) Python-based GUI embedded in ABAQUS

(b) Example of a resulting RVE-model

Fig 3: GUI and resulting 3D RVE

ideally elastic, however, in this case the properties are degraded with ongoing cyclic loading based on an energy criterion as explained further in the next section. Also for the cyclic loading, the traction-separation approach of the cohesive elements is replaced by a continuum mechanics approach similar to the matrix material.

Tab. 1 lists the assumed material parameters used for the static load cases with $Φ_f$ and $Ψ$ denoting the friction and the dilation angle used in the Mohr-Coulomb-model, respectively, and $t_n$, $τ_s$ denoting the onset of Mode I and II degradation with $U_{frac}$ being the total fracture energy of the interface. The stiffness of the interface was chosen to be equal to the matrix stiffness. The cohesive stiffness $K$ can be calculated from the interface thickness $t$ and the Young’s modulus $E$ of the matrix via

$$K = \frac{E}{t} = \frac{3.63 \text{ GPa}}{0.165 \mu m} = 22 \cdot 10^6 \text{ GPa/m} \quad (2)$$

The interface thickness is chosen arbitrarily and is equal to 2.5 % of the fiber diameter of 6.5 µm which is used here.

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with $N$ being the current cycle number, $Δw$ the accumulated elastoplastic hysteresis energy, and $c_i$ material parameters. The damage evolution follows a Paris law [29] like equation
\[
\frac{dD}{dN} = \frac{c_3 \cdot \Delta w^{c_4}}{L} \tag{4}
\]
again with \(c_i\) being material parameters and \(L\) denoting the characteristic element length. This degradation procedure is very rough and well suited for low cycle fatigue with highly plasticizing load cases. However, for high cycle fatigue with low or no plastic deformation introduced, the accumulated elastoplastic energy is not an applicable energy measure. Instead for linear elastic materials, three energy measures are eligible to replace the elastoplastic hysteresis energy: the dilatational (volumetric), the distortional, or the total strain energy density can be postulated to accumulate and thus to initiate damage in the matrix according to equation (4). The study by Asp et al. [30] among others shows that the dilatational (volumetric) stress state around the fiber is the critical driving force for initiating fiber-matrix-debonding. For this reason, the dilatational strain energy density given by

\[
\Delta w = \frac{1 - 2\nu}{6E_D} \left( \sigma_{11} + \sigma_{22} + \sigma_{33} \right)^2 \tag{5}
\]

with \(E_D\) denoting the damaged Young’s modulus, and \(\sigma_{11}, \sigma_{22}, \sigma_{33}\) the normal stresses, is taken as the accumulation energy measure to assess the long-term behavior on a micromechanical level.

Following equation (4), the damage variable increase \(\Delta D\) for a given cycle jump \(\Delta N\) can be expressed within the time discretization as

\[
\Delta D = c_3 \cdot \Delta w^{c_4} \Delta N \tag{6}
\]

Note that the characteristic element length is omitted here since the accumulated energy is formulated in form of energy densities. Mesh dependency was found to be non-existent since no localization of the damage occurred with further mesh refinement on a benchmark single hole tension specimen.

The key to a numerically efficient analysis is the proper choice of the cycle jump size \(\Delta N\) in the above equation. Too large values result in inaccurate damage propagation whereas too little values increase the cost of the calculation. In this work, a constant value of 1000 cycles was used as a compromise of accuracy and efficiency. In a tension-tension load case, schematically displayed in Fig. 4, the framework is such that at maximum loading condition (point 1), the energy measure is evaluated within a user material subroutine (UMAT), the damage initiation criterion (3) is checked and if met, the damage variable \(D\) is calculated at every integration point according to equation (6). It is saved as a status variable which is being read before the next loading stage at point 2 so the stiffness is degraded individually at every integration point.

The cycles are displacement driven and implemented by the Python-tool using the amplitude feature in ABAQUS. The user can define the \(R\) ratio as well as the “upper” displacement gradient allowing for multiaxial fatigue load cases.

As mentioned above, the energy based degradation described above is then suitable for both the matrix material and the interface since the traction-separation behavior is replaced by a continuum mechanics approach. The effect of the interface can be accounted for by adjusting the degradation parameters separately for the matrix and the interface material. In this work, the interface properties are set equally stiff but more vulnerable in terms of damage initiation and propagation to account for the observance of early interface damage prior to the actual crack evolution, tab. 2. These parameters were fitted based on data provided by Sauer and Richardson [31] who investigate the fatigue behavior of polymers. To match the typical, qualitative shape of the stiffness degradation curve, the degradation procedure has been modified. While the damage variable \(D\) is still calculated according to equation (6), the actual stiffness degradation is carried out using the following expression:
Table 2: Material parameters used for the cyclic load case (in case of the resin material: qualitatively fitted using data by Sauer and Richardson [31])

<table>
<thead>
<tr>
<th>Parameter</th>
<th>Value</th>
<th>Interface</th>
</tr>
</thead>
<tbody>
<tr>
<td>(c_1) [MPa] (^{1.44})</td>
<td>1000</td>
<td>1</td>
</tr>
<tr>
<td>(c_2) [-]</td>
<td>-1.44</td>
<td>-1.44</td>
</tr>
<tr>
<td>(c_3) [MPa] (^{-1.5})</td>
<td>0.008</td>
<td>0.01</td>
</tr>
<tr>
<td>(c_4) [-]</td>
<td>1.5</td>
<td>1.5</td>
</tr>
</tbody>
</table>

\[ E_D = (1 - D^{2/3})E_0 \] 

where \(E_D\) is the degraded Young’s modulus, and \(E_0\) is the undamaged Young’s modulus.

To qualitatively verify the degradation procedure for the resin material, a neat resin RVE was investigated under tension-tension loading, fig. 5. While the parameters \(c_1\) and \(c_2\) control the damage initiation, i.e. the number of cycles until the onset of degradation, the parameters \(c_3\) and \(c_4\) control the shape of the degradation curve.

**4.3 Results and Discussion**

**4.3.1 Static loads**

In contrast to the statistically representative volume element used by McCarthy and Vaughan [25], the RVE used in this work creates completely random fiber distributions with predefined fiber volume fractions. The advantage is the universality of the tool as no detailed information of the microstructure except the well determinable fiber volume fraction is needed to create the micro model. However, the lack of statistical representativeness might lead to deviations in the damage behavior which are not negligible. To assure the validity of the model, a transverse loading was imposed on an RVE with roughly the same fiber volume fraction used in the numerical works by Vaughan and McCarthy [32] and the experimental works by O’Higgins et al. [33], fig. 6. As can be seen, the stress-strain relation...
obtained with the random fiber distribution matches very well. Deviations come from the natural scatter when using statistical methods to place the fibers in the surrounding volume. However, the slight non-linearity as well as the failure stress and strain can be reproduced very accurately. Also the damage initiation near but not at the location of the closest distance of the fibers match the observing of Asp et al. [30] as well as Vaughan’s [32] micro model. The final failure then occurs when the properties of the interface elements are completely degraded. The debonding locations introduce high local plastifying stresses in the matrix in-between the debonds so the developing matrix yield bands connect the fiber-matrix-debonds to form a transverse crack through the RVE.

The similarity between Vaughan’s statistically representative RVE and the completely random fiber distribution model here indicates that the damage behavior is not primarily characteristic of the statistical distribution but depends more on the fiber volume fraction which is the only input-parameter used here. In other words, the random fiber distribution with the same fiber volume fraction creates a characteristic fiber pattern similar to a statistically representative distribution. The lacking statistical aspect of the model used here is therefore of negligible importance for the damage behavior of the RVE.

4.3.2 Cyclic loads

The cyclic analysis has been carried out with a cyclic tension-tension (\( R = 0 \)) transverse strain loading simulating uniaxial transverse tension. The transverse strain amplitude was kept constant at 0.45 %. Since the cohesive elements have been replaced with ordinary solid elements using the energy criterion described above, the location of damage initiation must be checked for plausibility. In a first calculation, a deliberately small RVE was chosen which was meshed very finely to accurately study the crack initiation and growth behavior before calculating a larger, more representative RVE. Gamstedt and Sjögren [34] have investigated the damage initiation sites of glass fiber reinforced epoxy on a micro mechanic scale under tension-tension loading. Similarly to the static load case
investigated by Asp et al. [30], the damage initiation mechanism is a fiber matrix debonding near but not at the location of the closest distance of the fibers. This initiation and growth of fiber-matrix-debonding is very well reproducible with the RVE shown in fig. 7a.

With further cycling, the debonds spread through the matrix to form a dominant transverse crack which can be seen in the RVE calculation as well, fig. 7b. Gamsted and Sjörgen [34] point out that the actual debond locations are widespread across the observed cross-section while the actual transverse crack formation is somewhat random and sparse. In the calculation shown, all interfaces get damaged while only one transverse crack develops.

It can be observed experimentally that the debonding hot-spots are located east and west of the fibers spreading to the north and the south without growing completely around the fiber, fig 7c. Note that the experimental observations in fig. 7c were found upon static loading assuming that the damage mechanism is the same in case of cyclic loading.

Checking the RVE calculation, the same debonding behavior can be observed, highlighted in fig. 7b as well. No fiber-matrix-debond growths completely around the fiber which is due to the different stress states around the fiber which contribute more (west and east) or less (north and south) to the dilatational strain energy density. It can be concluded that the initiation of interface damage is mainly due to the dilatational stress state which also develops around a totally isolated fiber. This means that one fiber in an RVE also experiences interface damage, namely at the east and west of the fiber. Taking into account the fiber-fiber-interaction, the dilatational stress state is shifted from the east and west towards the location of the closest distance of the fiber and is exaggerated so that a higher amount of dilatational energy is observed compared to the isolated fiber. But since quasi isolated fibers damage as well, basically all fiber-matrix interfaces experience damage with ongoing cyclic loading so the observation of multiple fiber-matrix-debond locations reported by Gamsted and Sjörgen [34] is plausible. The crack coalescence, however, is mainly due to the fiber-fiber-interaction and highly sensitive to the location of the fibers to each other. A crack will form along an especially suitable path which can hardly be predicted beforehand. This explains the observation that damage initiation occurs at multiple locations while transverse crack formation is relatively sparse which is observed with the proposed model used here, and is consistent with the literature sources.

The actual transverse crack growth as connection of the fiber-matrix debonds at sparse locations as reported in [34] can be seen even better using an RVE which contains more fibers as shown in fig. 8. While multiple debond sites can be observed, the transverse crack behavior is very similar to the experimental observations shown in fig. 7c.

Fig. 8: Damage initiation sites (a), transverse crack formation for an RVE with 65.2 % fiber volume fraction
5. Conclusions and Outlook

In this contribution, general considerations on the fatigue behavior of composites are presented. The conceptual design of an cost-efficient test rig is presented. A categorization of the various numerical composite fatigue models is presented and evaluated in terms of applicability for different use cases. The authors come to the conclusion that the micro-scale plays an important role in fatigue damage initiation and propagation which cannot be considered homogenized only when the fundamentals of fatigue mechanisms are to be investigated numerically. A tool is presented to create RVE models with random fiber distributions capable of imposing static and cyclic load cases on a micro-scale. The characteristic damage behavior in static loading conditions is investigated and found to match the statistical representative volume element (SRVE) used by McCarthy and Vaughan [25]. The conclusions drawn are that the random fiber distribution creates similar fiber patterns which result in the same characteristic damage behavior as the SRVE.

For the cyclic load cases, a continuum damage mechanics model with energy based degradation is presented for the matrix and interface material. The material parameters for the matrix material are fitted based on typical stiffness degradation behavior of polymers available in the literature [31]. The energy measure used is the dilatational strain energy density which provides a profound damage initiation criterion for static loads [30]. The characteristic damage behavior in the cyclic load case matches the literature findings quite well. Since viscoelastic polymer properties, which are the cause for frequency dependence and temperature development during fatigue testing, are neglected in this model, the energy measure used in the current model might not be sufficient to accurately capture all fatigue effects at the micro-scale. The energy accumulated by the damping component of the material might be a more suitable energy measure for including temperature and frequency dependencies for instance.

The present work uses a cycle jump algorithm which is essentially a damage extrapolation procedure with constant values for $\Delta N$. The advantage is that not every cycle needs to be considered in the calculation saving massive computational cost at loss in accuracy. This algorithm needs to be refined so that an adaptive cycle jump can be performed, i.e. in highly developing life phases the cycle jump would be smaller, whereas in slowly developing damage phases the cycle jump could be larger. Also the effect of the accuracy needs to be investigated in detail.

The most effort for future work includes the experimental testing of the neat resin specimen with and without a roving included to validate the material models used here.

References


