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Multiscale high cycle fatigue models for neat and short fiber reinforced thermoplastic polymers

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ABSTRACT

Two multiscale modeling approaches are proposed to predict the high cycle fatigue (HCF) failure of neat thermoplastic polymers and of short glass fiber reinforced thermoplastics. The modeling is based on the concept of so-called weak spots within the polymer matrix, whose behavior couples viscoelasticity, viscoplasticity and continuum damage mechanics. Scale transition is achieved with nonlinear mean-field homogenization. The predictions are evaluated against experimental results.

1. Introduction

Short fiber reinforced thermoplastic polymers (SFRTPs) and neat thermoplastic homogeneous polymers (TPs) are materials that have been employed for several decades as an alternative for metals for various applications in automotive, sporting goods, consumer electronics, and many other domains, because of their interesting physical properties and the relative ease of their manufacturing. They have been used to accomplish various objectives, among them designing light-weight structures with controlled cost in order to reduce carbon dioxide (CO_2) emission, which is a major issue in almost all industries. For instance, European automakers are obliged to ensure a reduction of CO_2 emission from their new cars and light commercial vehicles of 30% by 2020, compared to the emission in 2011 [1]. As a result of their important expansion, produced parts are more and more likely to be subjected to extreme operating conditions such as thermo-mechanical fatigue. Effects of fatigue loadings are generally evaluated by costly and time-consuming experiments. On the other hand, numerical modeling has shown an increasing efficiency and allowed to limit recourse to experiments.

In order to predict fatigue life of unreinforced or neat TPs, several failure scenarios and experimental interpretations were employed in the literature to motivate modeling approaches. The most common scenario is based on defining the TPs fatigue failure by two stages, a first one of progressive degradation of material mechanical properties until crack initiation, and a second stage when the initiated crack propagates up to the final fracture of the work-part [2]. Cracks initiated at preexisting defects or inhomogeneities within the material or low density domains formed at nanometric scale as proposed by Mourglia-Seignobos et al. [3] for fatigue failure of PA66. According to Sauer and Chen [4] and Lesser [5] the crack initiation stage takes the majority of the fatigue lifetime for several polymers. Similarly, Janssen et al. [6] showed that based on fatigue tests on samples of polycarbonate (PC), polypropylene (PP), and PMMA, the initiation stage almost covers the entire fatigue life (>99%).

Failure scenarios were also proposed for SFRTPs under fatigue loadings, e.g. [7–11]. Horst and Spoormaker [7,8] studied the fracture surface, and they noticed the important role of stress concentration at fiber tips. Horst and Spoormaker [7,8] proposed a damage mechanism that consists of the following steps: damage begins with void formation, mainly at fiber ends; these voids coalesce into small cracks, which will propagate into the material and give the final failure. Other authors later confirmed the proposed

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mechanism and proposed similar ones. However two stages are almost always respected, which are crack initiation and its propagation. The initiation stage remains an open research field [12]. Recently, Klimkeit et al. [13] investigated the fatigue damage mechanisms of a short glass-reinforced PolyButylene Terephthalate and Polyethylene Terephthalate with fiber volume fraction of 30% (PBT+PET GF30) and different fiber orientations at loading ratio $R = 0.1$, using multiple techniques: *replica technique*, *Infrared imaging and scanning electron microscopy*. They confirmed the same mechanisms, however they added that damage observed along the fiber sides is related to spatial distribution of fibers rather than stress distribution around one single fiber, and they confirmed that the duration of the propagation phase of macrocrack is small compared to the lifetime of the material. More recently, Arif et al. [14,11] investigated the effect of relative humidity on the SGRPA6.6 damage mechanism. Their investigations confirmed the observations of Horst and Spoormaker [7,8] and their proposed damage mechanism. In conclusion, the damage mechanisms are complex, and they depend mainly on the TPs matrix behavior, the interface treatment between the fibers and the matrix. However the phenomena can be summarized in an idealized way in order to have a failure scenario which is nevertheless realistic enough to justify modeling approaches. The matrix behavior plays an important role [e.g. 15], and damage is mainly initiated at the fiber tips. The interface between fiber and matrix is also important, while fibers failure seems to be much less likely to happen under fatigue loading since the stress state is generally low enough compared to the fibers failure limit state.

In summary for TPs and SFRTPs, failure is a result of two processes: the initiation of a crack, and its propagation. Crack propagation is usually modeled using fracture mechanics (FM). Approaches based on FM are the most classical. They assume pre-existing defects, however they generally ignore the initiation stage of failure. Based on the work of Griffith [16], the well-known model of Paris and Erdogan [17] is used, which was developed initially for metals. It gives the rate of crack propagation per cycle as a function of the so-called stress intensity factor variation. Examples of references applying the Paris–Erdogan equation for both polymers and polymer composites are [18,2], which make the fundamental assumption that the fatigue lifetime is determined by the duration of the crack propagation phase. However, as already mentioned the initiation stage is the most important for many TPS and SFRTPs. Thus it should be taken into account in the modeling.

Generally, the initiation stage for TPs is modeled using fatigue failure criteria, designed to estimate a limit state representing failure, similarly to those employed under monotonic loadings. The formulation of those criteria is usually based on using invariants of stress or strain tensors. The basic idea is to separate the stress space into two regions, safe and unsafe. The operation is made using a closed surface containing the origin and bounding the safe region [19]. Among the most used criteria applied to TPs, one can cite those of Sines and Crossland. Several authors employed the existing failure criteria with or without modification in order to predict the *S-N* curves mainly of testing samples of TPs. An example of those works is that of Berrehili et al. [20,21] on the fatigue of HDPE. The authors showed that the previous criteria are not able to capture the effect of multiaxial loadings and the sensitivity to the mean stress. In order to capture those effects, they proposed a new criterion. The failure criteria based approaches may estimate the limit state for the material under a given laboratory fatigue loading, but they can't handle realistic fatigue loadings directly. Moreover, they don't allow to model progressive failure until crack initiation within the material. For this purpose, sophisticated material models such as Klompen et al. [22] that was used by Janssen et al. [6] for fatigue, are proposed in order to model the different progressive phenomena which may be responsible for the initiation of a crack such as softening and physical aging. The authors assumed similarities in deformation and failure kinetics under creep and cyclic loadings. The model was employed to predict the behavior of polycarbonate (PC). The material failure is defined as the moment at which a macroscopic strain reaches a critical value found experimentally. The main limitation of these models is that the considered material failure mode is mainly ductile and related to macroscopic geometric effects such as necking. In the case of HCF loading, considered generally to be mechanically dominated, the material failure may also be brittle and sudden without important macroscopic deformation.

The main existing modeling approaches of SFRTPs under fatigue loading, that are based on crack initiation prediction, may also be classified into two categories: failure criteria based approaches and progressive failure based approaches. Several approaches based on failure criteria are proposed in the literature, mainly also inspired from the existing criteria for metals. The material anisotropy plays an important role in SFRTPs fatigue failure, so classical criteria used for metals assumed to be homogenous and isotropic are not predictive. Hence some authors tried to include the anisotropy in a phenomenological way in order to be able to use such criteria. The main criteria employed for fatigue of SFRTPs, are principal stress and von Mises criteria [12,23], Crossland and Sines criteria [12], modified Tsai-Hill criteria [e.g. 24–26,13,12] and Manson-Coffin criteria [12]. Although the failure criteria based approaches can lead to satisfactory results in terms of prediction of the *S-N* curves of laboratory specimens, their use is still complicated for industrial purposes, on real workpieces. Their major limitations are that they can't handle a complex history of fatigue loading, and they can't model the observed progressive failure of the material, commonly called progressive damage. In order to measure progressive failure, several variables were employed. The most common one is the hysteresis loops. The change of their slope is usually considered as an indicator of progressive failure. Sophisticated modeling approaches, based on damage mechanics, were also used for SFRTPs progressive failure of the material, such as the model of Nouri et al. [27] which takes the material anisotropy in a macroscopic way. Since the microstructure represents a layered-like configuration, the approach of [28] developed for laminate continuous fiber composites within the framework of continuum damage mechanics, was employed with a modification of the damage evolution law. In the model of Nouri et al. [27], the material behavior is assumed to be elastic and anisotropic and the damage is introduced through five internal state variables. It was used to reproduce the experimental damage of short glass fiber reinforced polyamide 6 (SGFRPA.6) [27,29]. The major limitation of this type of approach is the insensitivity to microstructure change and to the loading rate, and it is difficult to be used in the case of real complex loading. Another criticism is about the measure of damage during loading, since the very low thermal conductivity of the polymer material together with the viscous part of its behavior, make it very sensitive to changes in temperature. The change of slope can be a result of self heating which can be important and may introduce a loss of stiffness of the material, but which is not necessarily completely irreversible. However the loss of stiffness due to mechanical fatigue is irreversible since it is related to permanent deformations. Therefore, measuring the damage based on the change of the hysteresis loops slope is not a reliable measure.

In conclusion, there is a need to develop a multiscale approach in order to reproduce the observed damage mechanisms, to take into account the complex microstructure and handle complex loadings, which will make the modeling of real workpieces and their design easier. Such a multiscale approach based on several simplifying assumptions (e.g., linear elastic response up to fatigue failure) has been proposed recently by Bidaine et al. [30], Krairi and