Abstract: ...

Keywords: micromechanics, fatigue, microcracking, oxidation
Durability of CFRP laminates under thermomechanical loading: a micromechanics-based damage mesomodel

G. Lubineau a, P. Ladevèze a, D. Violeau a

aLMT-Cachan (E.N.S. de Cachan / Université Paris 6 / C.N.R.S.)
61 Avenue du Président Wilson / 94235 Cachan CEDEX
Tel: 33 - 1.47.40.24.02 Fax: 33 - 1.47.40.27.85

Introduction

Due to the importance of their specific properties, carbon/epoxy laminated composites have acquired a major place in aeronautics and space applications. This interest has given rise to numerous research works on their damage behavior on all scales (especially in micromechanics: Parvisi and Bailey, 78 [1]; Highsmith and Reifsnider, 82 [2]; Tan and Nuismer, 89 [3]; ... and mesomechanics: Ladevèze, 89 [4], Herakovich, 98 [5]; ...). Today, as a result, we have a good understanding of their behavior under classical loading, which enables us to use them in the design of low-temperature, non-vital applications, such as tailcones and some parts of the fuselage of subsonic airplanes.

A new challenge proposed by the aeronautics community is now to extend the use of these materials to more critical applications, one of these being a future supersonic civil airplane to be developed in the framework of the French “Aeronautical Supersonic program” supported by the Ministry of Research and the Ministry of Equipment, Transport and Housing. In this particular application, the material will be subjected to intensive thermal cycles (four-hour takeoff/landing cycles). In addition to classical fatigue damage, because of the relatively high temperature (approximately 130°C at Mach 2.0) and the presence of oxygen, degradation due to the progressive oxidation of the material will also occur. In fact, the degradation process is a much more complex combination of static degradation, fatigue damage and oxidation-induced degradations, resulting in the rapid development of elaborate patterns of degradation, especially microcracking (McManus et al., 96 [6]; Kobayashi et al., 01 [7]; Lafarie-Frenot and Rouquie, 04 [13]).

Due to the long anticipated lifetime of the application (20,000 flights, i.e. 80,000 hours), its design requires the development of appropriate accelerated tests. In order to design these accelerated experimental conditions, one must be able to rely on a model with sound physical content, which is sufficiently accurate for extrapolation. The aim of this paper is to propose such an approach for predicting the degradation of aeronautics structures under these conditions.

The paper is divided into four parts.

In the first part, the micromechanical approach used to describe degradation under static loading (especially transverse microcracking) is reviewed. This model, which was developed in previous works (Ladevèze, 04 [9]; Lubineau and Ladevèze, 05 [10]), is a general micromechanical approach which can be used in association with a suitable computational strategy for structural calculations. In fact, it is an hybrid model. In part, it is discrete in that cracks are described through the a priori introduction of cracking surfaces. However, continuum damage mechanics is used for progressive degradation mechanisms, such as fiber/matrix debonding. A simplified microcracking law developed for this model is presented at this point and will be used for identification purposes in Parts two and three.

In the second part, the modeling of fatigue damage is presented. In our approach, even under cyclic loading, microcracking is still assumed to be governed by the static evolution law (Ladevèze and Lubineau, 03 [11]; Lubineau and Ladevèze, 05 [10]). The only consequence of cycling is a reduction of critical microcracking tenacities. Thus, microcracks appear simply due to the weakening of the intercrack materials. Whereas classical micromechanical approaches introduce specific laws for fatigue microcracking, this approach completely unifies the descriptions of microcracking under static loading and under...
fatigue loading, since the important concept of critical energy release rate is preserved. Here, the loss of critical tenacity due to fatigue is quantified thanks to a new internal variable. Some examples of the typical range of this variable for classical experimental results taken from the literature (Wang et al., 84 [12]) are presented and an associated evolution law is proposed.

The third part describes the introduction of the environment, especially an oxidizing atmosphere at high temperature, into the model. Here again, critical tenacities and microcracking are in the foreground. These are defined as functions of the temperature and of the partial oxygen pressure. Several simplifications are proposed and an estimation of this oxidation-induced reduction of tenacities for several environmental conditions is given.

Finally, the basic modeling aspects described in the first three parts using a simplified approach are introduced into the complete computational micromechanical approach. The last part presents first calculations based on experimental observations (Lafarie-Frenot and Rouquie, 04 [13]; Lafarie-Frenot and Ho, 05 [8]). These lead to a computational model based entirely on quantities which have a strong material interpretation. We hope that such a micromechanics-based model will make predictive extrapolations accessible in the near future.

### 1. A short review of micromodeling under static loading

#### 1.1. The damage mechanisms on the microscale

We assume that any complex state of degradation of a laminated composite results from the accumulation followed by the localization of four elementary mechanisms which are clearly identified on the microscale (Figure 1).

![Figure 1. The mechanisms of degradation on the microscale](image)

1. transverse microcracking, 2. local delamination, 3. diffuse damage, 4. diffuse delamination

The first two mechanisms ("transverse microcracking" and induced "local delamination") are usual and have been studied by the micromechanics community in numerous works, mostly in the framework of finite fracture mechanics as these mechanisms are "discrete" on the ply’s level (see Nairn, 00 [14]; Berthelot, 03 [15] for reviews).

The last two mechanisms ("diffuse damage" associated with fiber-matrix debonding, and "diffuse delamination" associated with microvoids within the interlaminar interface) are usually not introduced on the microscale, but have been introduced on coarser scales for a long time, e.g. in the damage mesomodel for laminates (Ladevèze and Le Dantec, 92 [16]; Ladevèze et al., 04 [17]).

#### 1.2. The computational damage micromodel for laminates

Although the micromechanics of laminates has been studied extensively, few of the available models enable the designer to carry out actual calculations on engineering structures. The formalism of most of the proposed approaches applies only in very specific cases.

Recently, a computational micromodel was proposed for the description of microcracking, local delamination, (visco)plasticity and diffuse damage in engineering structures (Ladevèze, 04 [9]). This is, in fact,
a hybrid approach (Figure 2).

![Figure 2. The computational damage micromodel: minimum cracking surfaces](image)

This model is, in part (transverse cracking and local delamination), discrete and minimum cracking surfaces are introduced *a priori* within the discretization of the structure. The failure criterion for each elementary cracking area is defined through well-known micromechanical considerations (the distinction between the initiation and propagation of transverse microcracking and the induced thickness effect: Dvorak and Laws, 87 [18]; Boniface *et al.*, 97 [19], ...). One can say that at least for that part the model belongs in finite fracture mechanics.

For the rest, diffuse damage and (visco)plasticity are described through continuum mechanics models. Diffuse damage is quantified by two damage indicators, $d'$ in the transverse direction and $d$ in the shear direction, which are assumed to be homogeneous within each elementary volume between potential cracks. A classical progressive evolution law is assumed (Ladevèze and Le Dantec, 92 [16]).

The general framework of the model takes into account all the mechanisms described in Section 1.1. In this paper, for the sake of simplicity, we will consider diffuse damage and transverse microcracking only.

### 1.3. Simplified law for transverse cracking and diffuse damage

Since our purpose is not to detail the complete computational damage micromodel, we choose to describe the evolution of transverse microcracking through a simplified homogenized law which can be obtained directly by probabilistic simulation using the previous hybrid model. Whereas in the above model transverse cracking is described by discrete cracked surfaces, here we use the classical microcracking rate $\rho$ (defined by $\rho = \frac{H}{L}$ where $H$ is the ply’s thickness and $L$ the mean spacing between cracks) to describe the level of transverse microcracking along a characteristic length of the edge of the ply (Figure 3).

![Figure 3. Approximate periodic description of the local microcracking pattern near the edge](image)

Thus, in the specific framework of this paper, the level of degradation in the vicinity of a point is
completely defined by an “equivalent” microcracking rate and by the level of diffuse damage.

Let us denote \( G_{I_{ini}}, G_{II_{ini}} \) and \( G_{III_{ini}} \) the energy release rates of each cracking initiation mode. For the initiation of cracks, we choose the following, classical micromechanical evolution law:

\[
\left[ \left( \frac{G_{I_{ini}}}{G_c} \right)^{\alpha} + \left( \frac{G_{II_{ini}}}{G_c} \right)^{\alpha} + \left( \frac{G_{III_{ini}}}{G_c} \right)^{\alpha} \right]^{\frac{1}{\alpha}} \geq 1 \Rightarrow \dot{\rho} \geq 0
\]  

(1)

where \( G_{I_c}, G_{II_c} \) and \( G_{III_c} \), which from here on will be referred to as \( \{ G_c \} \) for simplicity’s sake, are the current critical energy release rates. For static loading, these tenacities are almost constant throughout the cracking process, except for very small microcracking rates where the apparent tenacities are smaller because of the presence of large initial defects. In fact, for static loading, \( \{ G_c \} \) appears to take the following form:

\[
\{ G_c \} = \{ \tilde{G}_c \} = \left[ G_{I_c|0}, G_{II_c|0}, G_{III_c|0} \right] \cdot \psi(\rho)
\]

(2)

Simple forms of the energy release rates \( G_{I_{ini}}, G_{II_{ini}} \) and \( G_{III_{ini}} \) can be found. For example, for pure Mode-I loading, \( G_{I_{ini}} \) can be expressed (Ladevèze and Lubineau, 01 [20]) as:

\[
G_{I_{ini}} = \frac{\mu}{2} \cdot \tilde{\sigma}_{22}^2 \cdot f(\rho)
\]

(3)

where:

- \( \tilde{\sigma}_{22} \) is called the ”effective transverse stress” (Ladevèze and Lubineau, 03 [21]). Of course, in the case of initiation near the edge, this stress must include free edge effects. For the laminates studied in Parts 2 and 3, in which the edge effects are small, this stress will be estimated using classical laminate theory.
- \( \mu \) represents the thickness effect during initiation, which is experimentally well-known. In practice, for a ply of thickness \( H ), \mu \) is governed by the following law:

\[
H > H_c \Rightarrow \mu = H_c \quad ; \quad H \leq H_c \Rightarrow \mu = H
\]

(4)

where \( H_c \) is a critical thickness defined for each material. As a consequence, for thin plies, the result is an energy criterion, whereas for thick plies, the result is an apparent stress criterion.
- \( \tilde{E}_2 \) is the transverse Young’s modulus of the material within which microcracking develops. Therefore, in practice, it is the transverse modulus of the healthy material modified by the current level of diffuse damage.
- \( f(\rho) \) is an intrinsic homogenized function which is accessible through probabilistic simulations using the general model and can also be approximated analytically.

In summary, the cracking curve for the simple case of a ply under transverse tension can be represented as:

\[
\tilde{\sigma}_{22}^2 \cdot f(\rho) = \frac{2 \cdot G_{I_c}}{\mu}
\]

(5)

2. Modeling of the degradation under cyclic loading

2.1. Microphenomenology

The first consequence of cyclic loading is the development of diffuse damage. Classically, this is modeled as the superposition of a static part and a fatigue part such that, at any time:

\[
(\tilde{d}, \tilde{d}') = (\tilde{d}_s, \tilde{d}'_s) + (\tilde{d}_f, \tilde{d}'_f)
\]

(6)

A power law of the associated damage forces appears to be appropriate for describing the evolution of the fatigue part \( (\tilde{d}_f, \tilde{d}'_f) \) (Payan and Hochard, 02 [22]).
Point 1: Our approach to microcracking, which was introduced in (Ladevèze and Lubineau, 03 [11]), is that its development is still governed by the static law, even under cyclic loading. However, the microcracking tenacities are assumed to decrease with each cycle, leading to a natural increase of the microcracking level, which follows Equation (1).

Let us denote \( \{ \tilde{G}_c \} \) these “damaged” tenacities. Of course, these evolve during the whole cyclic process; thus, in a general sense:

\[
\{ \tilde{G}_c \} = \{ \tilde{G}_c(N, C_{cyc}) \} \tag{7}
\]

where \( N \) is the number of cycles and \( C_{cyc} \) denotes the cyclic loading conditions (maximum stress level, loading ratio, and so on).

In previous works (Lubineau and Ladevèze, 05 [10]), these “damaged” tenacities were linked directly to the current level of diffuse fatigue damage. However, this damage, although it exists, appears to be very small. Under pure transverse cyclic loading, such a level is not even measurable. Therefore, diffuse damage, whose definition is based on the loss of stiffness, may not be an appropriate variable for measuring the cyclic weakening of the material. We prefer to introduce a new internal variable \( \eta_f \) such that, for cyclic loading:

\[
\{ G_c \} = \{ \tilde{G}_c \} = \{ \bar{G}_c \} (1 - \eta_f) \tag{8}
\]

Remark: Thus, \( \eta_f \) can be viewed as a damage variable related to the critical quantities. It compounds the influence on the critical quantities of all the homogeneous mechanisms which depend on the cyclic loading.

2.2. Decrease of tenacities during cyclic loading

Let us now illustrate the variation of \( \eta_f \) during classical fatigue experiments. Experimental observations on fatigue tests are usually described in terms of the microcracking rate as a function of the number of cycles for prescribed fatigue conditions. Extensive experimental works have been published: Boniface and Ogin, 89 [23]; Berthelot et al., 01 [24]; Yokoseki et al., 02 [25]; Sun et al., 03 [26]; . . . For illustration purposes, Figure 4 shows the mean values of the experimental observations by Wang et al., 84 [12] for mechanical fatigue on cross-ply carbon/epoxy laminates.

If one assumes that microcracking always appears during the maximum stress state (\( \tilde{\sigma}_{22} \mid_{\text{max}} \)) of each cycle, Relations (5) and (8) can be used to calculate the evolution of \( \eta_f \) from such classical information. For example, under pure Mode I:

\[
\eta_f(N) = 1 - \frac{\mu}{2G_I} \cdot \frac{\sigma_{22} \mid_{\text{max}}}{E_2} \cdot f(\rho) \tag{9}
\]

Thus, the evolution of \( \eta_f \) can be identified and is illustrated in Figure 5 for the experiments by Wang et al., 84 [12]. For each laminate and each stress level, \([\eta_f \leftrightarrow \log(N)]\) is found to be a linear law. This seems to be a very general conclusion, since it was also verified on several other materials.

Nevertheless, further mechanical interpretation of the results of Figure 5 is difficult due to the fact that in this drawing a large number of different microcracking states and stress levels are mixed at each point. (For example, Points A, B, C and D of the four samples have approximately the same level of fatigue damage \( \eta_f \), but the actual experimental degradation states, such as microcracking rate, of all these points are very different.) Therefore, we now need to build an actual evolution law taking into account all these differences.

2.3. The evolution law for \( \eta_f \)

In accordance with the phenomenology described in Section 2.1, \( \eta_f \) quantifies the progressive weakening of the intercrack material.

Point 2: The rate of evolution of \( \eta_f \) within a finite intercrack volume \( \partial \omega \) (Figure 6) is a function of the local energy stored within that volume. Therefore, we propose to adopt for the evolution law the general form:

\[
\frac{\partial \eta_f}{\partial N} = \mathcal{F}(\ll e \gg \partial \omega, \eta_f) \tag{10}
\]
Figure 4. Experimental microcracking rate vs. the number of mechanical cycles (mean values of fatigue tests at room temperature, from Wang et al., 84 [12])

Figure 5. Evolution of $\eta_f$ during cycling for the experiments of Figure 4

where $\ll \cdot \gg$ designates the mean value.

An example of such a law for the experimental results presented in Figure 5 is shown in Figure 7. In the case where the transverse stress alone is responsible for microcracking, an intrinsic evolution law turns out to be:

$$log\left[\frac{\partial \eta_f}{\partial N}\right] = f_1 \cdot e^{\eta_f} + f_2$$

(11)
Figure 6. Finite volumes of homogeneous material between existing cracks

where \( f_1 \) and \( f_2 \) are two identified scalar material quantities, and \( e_{\eta_f} \) is:

\[
e_{\eta_f} = (1 - \eta_f) \cdot \sqrt{\frac{\sigma_{22}^2}{2E_2}}
\]

For the approximate approach presented here, \( \sigma_{22} \) is defined as the mean stress at the midpoint between existing cracks.

Figure 7. Experimental identification of \( \frac{\partial \eta_f}{\partial N} \) for the experimental results by Wang et al., 84 [12]

In order to complete the description of the damage evolution, it is logical to take into account the existence of a fatigue threshold \( e_{th} \), below which no fatigue damage occurs.

3. Modeling of the oxidizing atmosphere

3.1. Oxidation as a reduction of tenacities

**Point 3:** Concerning oxidation (or, more generally, any environmental condition), our approach assumes that only critical quantities are involved. The current tenacities of the oxidized material at a point \( M \) of a structure are written as:

\[
\{ G_c \} = \{ G_c^{\text{ox}} \} = \{ \bar{G}_c \} \ast (1 - \eta_{\text{ox}}) \]

where \( \eta_{\text{ox}} \) quantifies the relative weakening of the material associated with oxidation.
Since oxidation is a reaction-diffusion mechanism controlled mainly by the temperature ($T$) and the partial oxygen pressure ($P_{o_2}$), a very general expression of $\eta_{ox}$ is:

$$\eta_{ox} = \eta_{ox}(T; P_{o_2}; M; t)$$ (14)

Much effort has been spent by the physics and chemistry communities in order to describe the concentration of the products of oxidation within polymers (Colin et al., 02 [27]). In the future, we expect that it will be possible to implement such laws directly into our model in order to take the relevant associated material quantities into account directly in the structural calculation. Unfortunately, today these laws are valid only for pure resin and only begin to address the case of unidirectional plies (Colin et al., 05 [28]). Moreover, there is no link yet between these works on the chemical composition and the apparent residual material properties which are the essential data for structural calculations.

Therefore, what we propose here is a simplified description of oxidation which we believe to be sufficient for practical applications.

### 3.2. Approximation of the evolution of tenacities

The general expression (14) is, for the time being, too complicated to be applicable to the calculation of complex structures. As a simplification, we propose to assume that from a mechanical point of view the effect of oxidation is instantaneous as soon as the material is in contact with the oxidizing atmosphere. This supposes that the characteristic time for oxidation is small in comparison with the characteristic time of the fatigue experiments.

In order to illustrate this point, let us consider the experimental results by Sun et al., 03 [26]. The authors carried out constant-stress sinusoidal loading cycles on unidirectional [90$^\circ$] IM7/977-3 laminates at a frequency of 4 Hertz. The stress ratio was 0.5 and several maximum stress levels $S$ between 0.6 $F_{2t}$ and 0.9$F_{2t}$ were tested, where $F_{2t}$ denotes the static strength of the sample at the temperature of the test. Two temperatures were tested ($T = 24^\circ$C and $T = 149^\circ$C) and the resulting stress-life curves given by the authors are shown Figure 8.

![Stress-life curves of [90$^\circ$]$_8$ laminates at 24$^\circ$C (75$^\circ$F) and 149$^\circ$C (300$^\circ$F) (from Sun et al., 03 [26])](image)

Let us now assume that the failure of unidirectional samples resulting from the unstable propagation of a crack starting near the edge is representative, at least qualitatively, of the failure of the first crack. In this case, according to Equation (5), the static strength of the sample at $T = 149^\circ$C is given by:
During cyclic loading at for a given maximum load $S$, the sample fails when:

$$S^2 = \frac{2\tilde{\gamma}_c|_{149}}{\mu} \cdot \tilde{E}_2|_{149} \cdot \tilde{f}(0)$$  \hspace{1cm} (15)$$

where:

$$\tilde{\gamma}_c|_{149} = \frac{\tilde{\gamma}_c}{G_c|_{149}} \cdot (1 - \eta_f) \cdot (1 - \eta_{\text{ox}}(T, P_{O_2}; t))$$  \hspace{1cm} (17)$$

Thus, the stress-life curve at 149°C is defined by Equations (15) and (16) such that:

$$\frac{S^2}{F^2_{2t}|_{149}} = \frac{G_{c|149}^{\text{ox}}}{G_c|_{149}} \cdot (1 - \eta_f)$$  \hspace{1cm} (18)$$

$$S^2 = \frac{F^2_{2t}|_{149}}{F^2_{2t}|_{24}} = (1 - \eta_f)$$ \hspace{1cm} (19)$$

Finally, considering that the evolutions of $\eta$ in the two experiments are very close, one gets:

$$\frac{G_{c|149}^{\text{ox}}}{G_c|_{149}} = (1 - \eta_{\text{ox}}(T, P_{O_2}; t)) \approx \frac{S^2}{F^2_{2t}|_{149}} \cdot \frac{F^2_{2t}|_{149}}{F^2_{2t}|_{24}}$$ \hspace{1cm} (20)$$

Relation (20) was plotted as a function of time in Figure 9. Our first conclusion is that the effect of oxidation on the critical quantities is far from negligible (here, it corresponds to an approximate twelve per cent reduction for $T = 149^\circ C$ and $P_{O_2} = 0.2$ bar).

Figure 9. Evolution of $\eta_{\text{ox}}$ with time estimated using the results of Figure 8

Nevertheless, the reduction of $G_{c|149}^{\text{ox}}$ appears to occur very quickly. After approximately thirty minutes, $G_{c|149}^{\text{ox}}$ remains constant throughout the rest of the test. This last conclusion leads us to a simplified model:
Point 4: We assume that the effect of oxidation is immediate, at least in a thin layer surrounding the active surfaces (i.e. the surfaces denoted $S_{ox}$ which are in contact with the oxidizing atmosphere). Therefore, we assume that along the active surfaces the critical material quantities get immediately degraded. At points which are not connected with an active surface, the material is assumed to remain healthy. Consequently, one has:

$$\{G_c\} = \{G_{c}^{ox}\} = \{\bar{G}_c\} \ast (1 - \eta_{ox}(T, P_{O_2})) \quad \forall M \in S_{ox}$$ (21)

3.3. Dependence of $\eta_{ox}$ on $(T, P_{O_2})$

The effect of oxidation plays an important role in thermomechanical fatigue. To prove this point, (Lafarie-Frenot and Rouquie, 04 [13]) and (Lafarie-Frenot and Ho, 05 [8]) conducted experiments on IM7/977-2 [$0_3/90_6/0_3$] and [$-45_3/45_6/-45_3$] laminates under different atmospheres (nitrogen, air and oxygen) and at different maximum temperatures. Figure 10 summarized their main results for the cross-ply laminate. The results for [$-45_3/45_6/-45_3$] were quite different due to different edge effects.

![Figure 10. Density of transverse edge crack vs. the number of thermal cycles for \([0_3/90_6/0_3]\) IM7/977-2 under different atmospheres and maximum temperatures. \([A,B,C]\): thermal cycles \([-50/180]^{\circ}C\); D: thermal cycles \([-50/150]^{\circ}C\)]

The phenomenology described in Figure 11 for progressive cracking under thermomechanical fatigue and oxidizing atmosphere is assumed.

In summary, oxidation takes place at the highest temperature of the cycle ($T_{max}$), leading to a first reduction of the tenacities quantified by $\eta_{ox}(T_{max}, P_{O_2})$. Furthermore, the tenacities are reduced during the cyclic loading due to the development of $\eta_{f}$ with each cycle, an evolution which is governed by the maximum stress. Thus, at any time, the complete relative reduction of the critical quantities can be written as:

$$(1 - \eta) = (1 - \eta_{f}) \cdot (1 - \eta_{ox}(T_{max}, P_{O_2})) \quad (22)$$

Conversely, microcracks are assumed to develop when the stress state is maximum, i.e. at the minimum temperature ($T_{min}$). Thus, at any time, microcracking is governed by the reduced tenacity at low temperature, which can be written as:
where $\bar{G}_I^{c|T_{\text{min}}}$ denotes the initial tenacity at low temperature.

For each experimental condition, $\eta$ can be expressed as a function of the number of cycles as:

$$1 - \eta(N) = \frac{\mu}{2\bar{G}_I^{c|T_{\text{min}}}} \cdot \frac{\tilde{\sigma}_{22}^{s|T_{\text{min}}} \cdot \tilde{E}_{2} \cdot f(\rho)}{E_2}$$

This evolution is represented for each experimental condition in Figure 12.

Now, the values of $\eta_{\text{ox}}(T_{\text{max}}, P_{O_2})$ for each case can be extracted from these results. Indeed, from Equation (22):

$$\log(1 - \eta) = \log(1 - \eta_F) + \log(1 - \eta_{\text{ox}}(T_{\text{max}}, P_{O_2}))$$

Let us make two remarks:
• First, in all the experiments, the maximum effective stresses are very close to one another and, therefore, the relation \([\log(1 - \eta_F) \leftrightarrow \log(N)]\) is the same in all cases.

• Of course, \(\eta_{ox}\) is zero for a nitrogen atmosphere. Therefore, the common reference curve \([\log(1 - \eta_F) \leftrightarrow \log(N)]\) is known.

The results associated with each experimental condition \((T_{\text{max}}, P_{O_2})\) must be obtained by translation of \(\log(1 - \eta_{ox}(T_{\text{max}}, P_{O_2}))\) from this common reference curve in a logarithmic diagram (Figure 13).

![Figure 13. Evolution of the complete reductions \((1 - \eta)\) of the tenacities](image)

Applying that procedure, we obtain the following values for \(\eta_{ox}\):

<table>
<thead>
<tr>
<th>(T_{\text{max}})</th>
<th>(P_{O_2})</th>
<th>(\eta_{ox})</th>
</tr>
</thead>
<tbody>
<tr>
<td>150</td>
<td>1</td>
<td>0.58</td>
</tr>
<tr>
<td>180</td>
<td>0.2</td>
<td>0.82</td>
</tr>
<tr>
<td>180</td>
<td>1</td>
<td>0.87</td>
</tr>
</tbody>
</table>

One can observe that the dominant effect is that of the temperature, whereas the effect of the partial oxygen pressure is only secondary. Indeed, the identified values of \(\eta_{ox}\) for oxygen and air at 180°C are very similar, and much higher than the identified value for pure oxygen at 150°C.

One can also observe that taking \(\eta_{ox}\) as a constant is only a first approximation which may need to be improved for short exposure times. Nevertheless, for studying the behavior for very long durations, we believe that this approximation, which simplifies the calculations considerably, is reasonable.

4. First results and discussion

The approach presented above for fatigue and oxidation was integrated into the computational damage micromodel for laminates reviewed in Section 1.2. Here, we present the first illustrations of the expected results for pure thermal fatigue of \([0_3/90_3]_s\) IM7/977-2 laminates, based on the experimental procedures and results described in (Lafarie-Frentot and Rouquie, 04 [13]) and (Lafarie-Frentot and Ho, 05 [8]).

The discretization used is shown in Figure 14. Using symmetry considerations, only one-eighth of the sample was analyzed. According to the computational damage micromodel, each ply was a priori
Figure 14. The studied geometry. A: complete $[0_3/90_3]_s$ sample. B: discretization of one-eighth of the sample and detail of the elementary cell used in the calculation.

divided into homogeneous elementary volumes (in which diffuse damage could develop) and interfaces, which, depending on their orientation, represented either transverse microcracking or local delamination. Each elementary volume was meshed in 3D, leading to a large number of degrees of freedom. Thus, the approach proposed is relatively simple from the “modeling” point of view, but requires powerful computational capabilities. Therefore, multiscale computational strategies must be developed and used for such problems. The resolution performed here, which will be detailed in further publications, is based on the approach proposed at the LMT-Cachan (Ladevèze et al., 01 [29]; Ladevèze and Nouy, 03 [30]).

Figure 15. Computed microcracking pattern as a function of the atmosphere (pure thermal cyclic loading $[-50/180]^{\circ}C$)

- Since the edge effects are taken into account, the microcracks are initiated near the edge, especially for the central ply. Then, the cracks propagate in the sample from the edge to the core according to an initiation/propagation scheme included in the computational micromodel.
• For pure cycling under nitrogen, the degradation begins very late (after 380 cycles, which is close to the experimental value, see Figure 10).

• The effect of oxidation is catastrophic, since after a few tens of cycles, under both air and oxygen atmospheres, many microcracks appear at the edge of the central ply. The central ply of the sample exposed to pure oxygen is saturated very quickly at the edge, which is consistent with the experiments (see Figure 10).

• Skin plies are not degraded under a nitrogen atmosphere, but show very high cracking densities under air and oxygen. Again, this is consistent with the experiments. The explanation is that under nitrogen, due to a minor edge effect in skin plies, microcracking develops very late. On the contrary, under an oxidizing atmosphere, the reduction of the critical tenacities throughout the skin makes the initiation of transverse cracking possible from any point of the surface.

5. Conclusion

We proposed a pragmatic approach for the description of the degradation of laminated composites under cyclic loading, including the effect of oxidation. This approach is based on a clear classification of the degradation mechanisms on the microscale. Continuous damage mechanics is used for progressive mechanisms on the ply’s scale, whereas finite fracture mechanics is used for “discrete” mechanisms, through minimum cracking surfaces introduced \textit{a priori}.

The effects of cyclic loading and of the environment are taken into account through the consecutive decrease in critical quantities.

Concerning fatigue, since the important concept of critical energy release rate is preserved, this approach unifies the descriptions of microcracking under static loading and cyclic loading. The fatigue evolution is not described by a phenomenological relation between the cracking density and the number of cycles, but the rate of decrease in the critical quantities is related to the energy density within the “fiber-matrix” intercrack material, which, in our opinion, makes more sense physically. Moreover, this makes it possible to describe both crack initiation at the edge and crack propagation towards the core using a single fatigue law. This point, which will be clarified and discussed in subsequent publications, differs from classical micromechanics analysis, which associates separate laws with these two modes of evolution.

The effect of the environment seems to be well-described, despite the coarse approximation used for the evolution of the critical quantities. Several orders of magnitude were identified using experimental results taken from the literature. The maximum temperature of the cycle is the main parameter. Further research remains necessary for the model to become completely quantitative and predictive, but the results obtained are promising.

Further research is also required, especially in computational mechanics, to turn this approach into an engineering tool.

REFERENCES


