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Modeling of Micro - Damage of E-Glass/Epoxy Composite

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***Abstract*— The paper deals with progressive and fatigue damage of E-glass/epoxy composites. The objective of the research is to experimentally investigate progressive failure modes, stiffness degradation and corresponding transverse matrix crack density of long fiber E-glass epoxy composite loaded in quasi-static tension and repeated one direction tension with constant load amplitude. Stress controlled fatigue tests of plain $[\pm 60]_S$; $[\pm 30]_S$; $[0]_8$ and $[0/90_2/\pm 45/90]_S$ samples were carried out in order to investigate residual stiffness degradation, corresponding fatigue progressive damage and failure modes of the laminate of different layups. Sudden onset of the transverse matrix cracks has been observed in off-axis plies. The crack density increases with increasing number of cycles as new transverse matrix cracks instantaneously form. It has been concluded that progressive/fatigue damage of the laminate is not a continuous homogenous process but the series of discrete sudden events emerging at ply level. Composite residual stiffness and corresponding transverse matrix crack density were correlated to the number of cycles and loading level. Progressive and fatigue damage was simulated employing multi-scale FEA modeling. Micro-model was used to simulate first ply failure (FPF) and meso-model was used to model gradual formation of transverse matrix cracks and the correlation to laminate stiffness.**

Keywords*— Composite; E-glass/epoxy; Progressive damage; Fatigue; Transverse matrix cracking; Stiffness degradation; Multi-scale modeling

I. INTRODUCTION

Even though composite material can sustain certain amount of fatigue damage - it is damage tolerant - its mechanical properties decline as a result of fatigue loading. The objective of this paper is to investigate stiffness degradation of long fiber E-glass/epoxy composite loaded in cyclic tension with constant load amplitude and to study progressive failure modes, transverse matrix crack density and consequent stiffness decline.

In contrast to conventional materials such as steel, more phenomena are involved in the process of composite fatigue. Fatigue of fiber, matrix and the interface takes place and interact with each other. This results in several interacting failure modes such as matrix micro-cracking, fiber-matrix debonding, fiber rupture, delamination, fiber kinking, fiber pullout, etc. There is usually no dominant fatigue crack as in case of metals.

Micro-damage of the constituents as well as changes in mechanical properties of a particular constituent affects global mechanical properties such as stiffness or strength. In order to predict residual mechanical properties in the most general case it is essential to analyze and understand the progressive damage failure modes.

The objective of the research is to experimentally investigate progressive failure modes, stiffness degradation and corresponding micro damage of long fiber E-glass/epoxy composite loaded in static tension and cyclic load controlled tension. Once real progressive failure modes and mechanisms are understood, it can be computationally simulated. Multi-scale FEA model capable of modeling real failure mode, its evolution and its influence on the mechanical response of the material was employed. The homogenization of dispersed damage usually done by employing complicated material models was avoided. This will allow general use of the model for an arbitrary layup and stacking sequence and should significantly reduce the scope of experimental testing needed to obtain input data. Matrix micro-cracking is of particular interest because it is believed to be the main cause of delamination that can lead to fatal failure of a structure. Delamination itself and the phase of composite final failure are outside the scope of the research.

II. STATE OF THE ART

It is relatively simple to predict the initial failure of the composite ply – first ply failure (FPF). Existing failure criteria such as maximal strain or Tsai-Wu can be used. However, it is not an easy issue to detect FPF experimentally. Moreover, it is very complicated to study progressive and fatigue damage evolution after FPF and composite final failure. Stress redistribution due to existing damage takes place and must be taken into account in that case.

First attempts to study fatigue damage of composites were based on traditional approach of S-N curves. However, this method can hardly be used for general laminates because of great number of combinations of fiber and matrix materials, orientations of fibers, stacking sequences, etc. Montesano et. al. [1] distinguishes empirical or semi-empirical models based on extensive fatigue testing, models based on residual stiffness and strength and progressive failure models based on real failure modes. Mao and Mahadevan [2] use their method based on residual stiffness. Damage parameter is defined by initial, residual and fracture stiffness. Damage increment per cycle is given by function *f* dependent on stress range, stress ratio, damage, etc. Talreja [3] uses continuum damage mechanics to homogenize micro-damage. Muc [4] defines residual stiffness as a function of initial stiffness, fictive time, stress range, stress ratio and a measurable damage parameter such as matrix crack density. However, also these methods rely mostly on custom made experimental testing. These complicated models requiring large experimental data sets for implementation does not necessarily pay back in terms of accurate predictions [5].

Some methods are based on iterative FEA modelling in combination with failure criteria for different failure modes. These methods take into account damage induced stress redistribution but still need an experimental fitting parameter for the particular composite layup [6,7]. Methods based on micromechanical modeling [8,9] seem to be promising for general use. It uses representative volume element assuming periodicity of stresses and damage. However, it has been observed that once first ply failure initiates the structure is not periodic any more. To overcome this problems some methods model real progressive failure modes and its density. These methods are focused mostly on matrix cracking [10,11].

III. MATERIAL AND EXPERIMENTAL SETUP

The aim of the experimental testing was to investigate stress-strain behavior, stiffness degradation and corresponding progressive damage failure modes of long fiber E-glass epoxy composite samples loaded in quasi-static tension and cyclic tension with constant load amplitude. The testing followed the principles of ASTM D3479 [12] and ASTM D3039 [13] standards.

A. Samples preparation

The material is E-glass/epoxy laminate certified for aeronautics (fiber Interglas 92145, epoxy MGS LR385, hardener MGS LH 385/386). Fiber volume fraction is 40-41 %. Thickness of one laminate ply is 0.25mm. The laminate was fabricated using standard vacuum surface infusion method. Laminate plates were hardened for 24 hours on air and finish hardened in the furnace. Aluminum pads with chamfer angle of 90° were attached and the laminate plate was cut using water jet cutter. Dimensions of rectangular samples were 140x20mm.Surfaces of the samples were smooth to facilitate in-situ optical monitoring of micro-damage. Four sets of laminate layup were used: $[\pm60]_S$; $[\pm30]_S$; $[0]_8$ and $[0/902/\pm45/90]_S$. Layup of the sets was chosen to represent various combinations of normal and shear straining within individual plies when loaded by axial force.

B. Experimental setup

Universal testing machine MTS 810 equipped with hydraulic clamps was used for fatigue tension-tension loading. The testing machine is controlled by MTS 458.20 controller capable of processing load cell, extensometer and strain gauge data. The acquired data is stored in the memory of controlling PC. Progressive failure modes – especially transverse matrix cracks and its density – were recorded using in-situ micro-camera (200x magnification). Experimental setup together with detailed snapshot of transverse micro-cracks in $[\pm60]_S$ laminate is depicted in Fig. 1. Self-compensating foil strain gauges Omega SGD-7/350-LY43 were used to measure strain. Post-mortem investigation of the fracture surface was performed employing optical microscope Nikon Eclipse LV100.

C. Testing procedure

In case of quasi-static testing the samples were loaded in monotonous tension with constant displacement rate 0.5mm/s. Stress-strain response and corresponding progressive failure modes were recorded using load cell, strain gauges and in-situ micro-camera. In fatigue, samples were loaded in repeated one direction tension with constant load amplitude. Stress ratio was close to 0.1. Frequency of the loading was 5 Hz. No autogenous heating was observed at this frequency. All samples were tested at room temperature. At least 6 samples of every layup were tested in fatigue. Initial and residual stiffness of the specimens was recorded. Corresponding progressive failure modes were recorded employing in-situ micro-camera and processed using direct optical observation. This method is analogical to X-ray NDT method. However, E-glass epoxy is transparent and transverse matrix cracks can be observed straight away. Therefore, radiation-resistant penetrant that can influence fatigue properties of the material does not need to be used. Moreover, the method of direct observation leads to significant time savings.

Whilst peak load remained constant throughout cyclic testing (load controlled loading), peak strain continuously increased due to softening of the samples. The relation between such load and corresponding strain was identified as residual stiffness. Afterwards, residual stiffness was normalized with respect to its initial value to form normalized residual stiffness. More details can be found in [14,15].

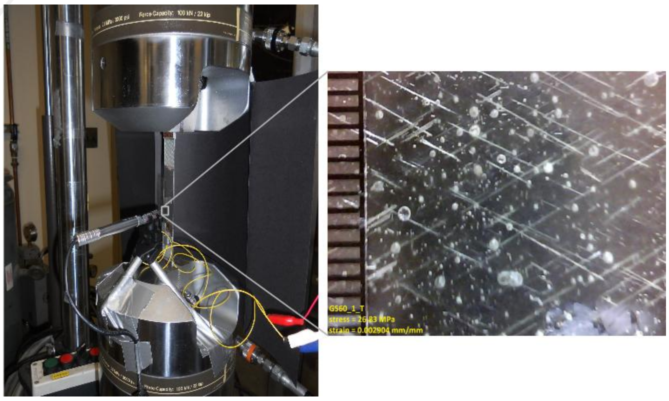


Fig. 1. Experimental Setup: E-glass/epoxy sample attached in hydraulic clamps with in-situ micro-camera to monitor micro-damage. Internal micro-cracks in $[\pm60]_S$ sample in detail at right. One scale division represents 0.5 mm

IV. EXPERIMENTAL RESULTS AND DISCUSSION

A. Static testing

Stress-strain response of several samples of all layups is shown in Fig. 2. Tangent modulus E normalized to its initial value E_{init} was evaluated in Fig. 3. Transverse matrix crack density of 90°, 60° and 45° plies as a function of strain in loading direction is shown in Fig. 3. Stress-strain response of $[\pm60]_S$ and $[0/902/\pm45/90]_S$ laminate is linear until first ply failure (first matrix crack) occurs. Response of $[0]_8$ samples is linear until final failure. No matrix cracking was observed in case of $[\pm30]_S$ and $[0]_8$ samples. It has been observed, especially for 90_2°, 60°and 45° plies, that the progressive damage is not a continuous gradual process. Contrariwise, it is a series of discrete events at ply level – abrupt formations of transverse matrix cracks.

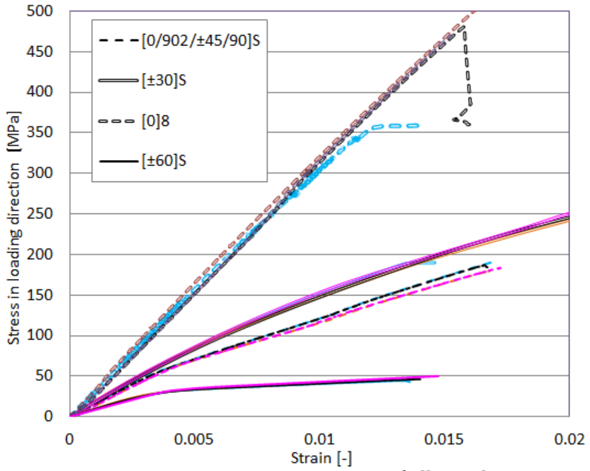


Fig. 2. Stress strain response of all samples

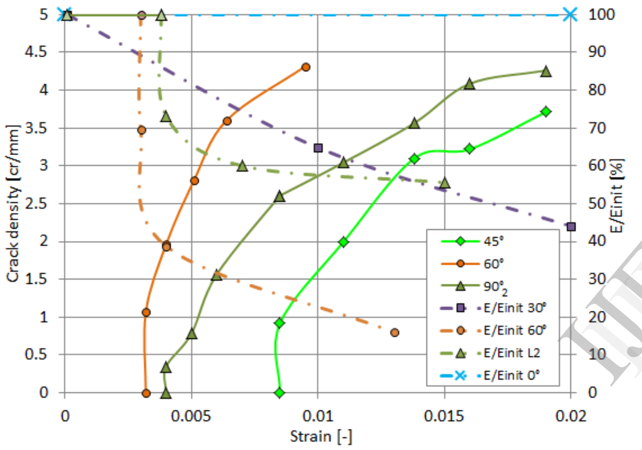


Fig. 3. Transverse matrix crack density as a function of strain in the loading direction (left axis), normalized tangent modulus as a function of strain in the loading direction (right axis)

B. *Fatigue testing*

The example of fatigue testing results for $[\pm60]_S$ samples are shown in Fig. 4 and Fig. 5 as the relationship between number of loading cycles and residual normalized stiffness for various loading levels related to static tensile laminate strength R_{mt}. Initial steep decrease followed by gradual decline of the residual stiffness was observed as expected. Extensive matrix cracking can be observed using in-situ micro-camera in case of $[\pm60]_S$ and $[0/90_2/\pm45/90]_S$ samples. The principles of progressive damage development are analogous to those described in the previous section. The matrix micro-cracks propagate instantaneously across the entire thickness of the ply and stop at the ply interface where it reaches equilibrium. Since the cracked ply is constrained at its boundaries by the neighboring plies the crack relieves the stress in the adjacent region. With increasing number of cycles fatigue properties of matrix degrade and new matrix cracks are formed. The phenomenon of residual stiffness saturation can be observed. The actual progressive damage state corresponding to the number of cycles is depicted in Fig. 6 for $[\pm60]_S$ layup.

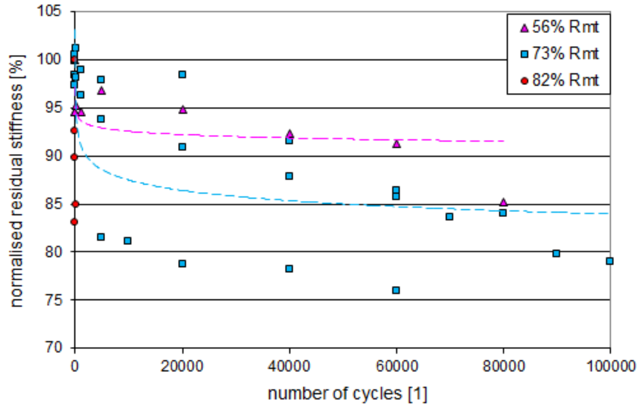


Fig. 4. Relationship between number of loading cycles and normalized residual stiffness for various loading levels related to static tensile laminate strength R_{mt} for $[\pm60]_S$ samples

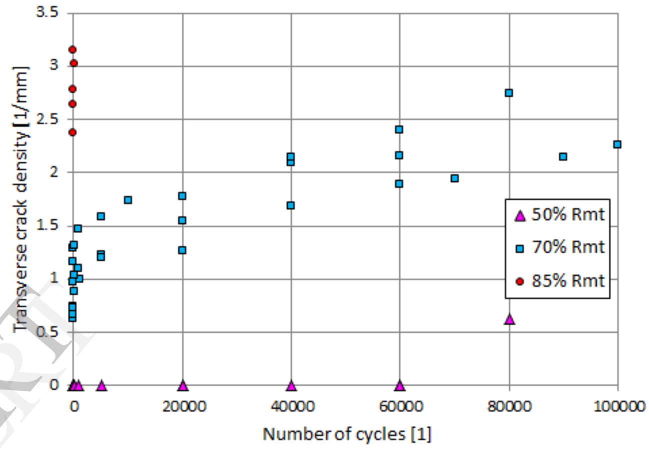


Fig. 5. Relationship between number of loading cycles and matrix crack density for $[\pm60]_S$ samples

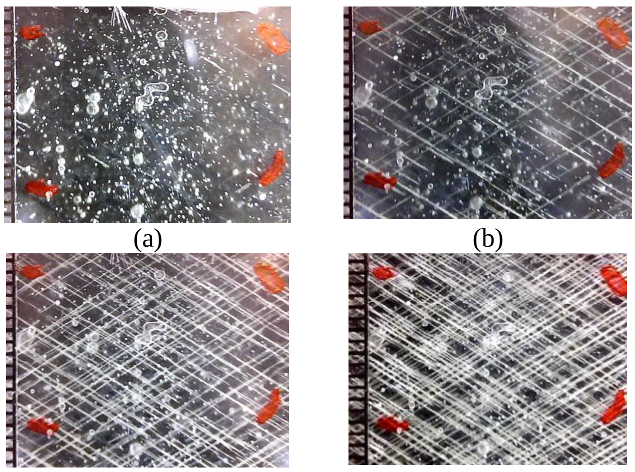


Fig. 6. Snapshots of composite micro-damage (transverse matrix cracks) obtained using in-situ camera for $[\pm60]_S$ sample. Peak stress 70% R_{mt}. (a) N = 0, (b) N = 1, (c) N = 1000, (d) N = 100.000

Fatigue loading itself is stress-controlled. However, as matrix degrades, new cracks are formed in spite of the fact that both stress amplitude and stress ratio remain constant. Therefore, both peak strain of the loading cycle and matrix crack density increase. Fig. 7 shows the relationship between axial strain of $[\pm60]_S$ samples and corresponding matrix crack density for both static (line) and fatigue loading for increasing number of cycles (dots). It can be observed that with increasing number of cycles, matrix crack density increases. Strain increases as well due to crack-induced softening. It can also be concluded, that matrix properties degrade in fatigue because matrix cracks are initiated even for lower strain than the strain needed for FPF in case of static loading. Due to degradation, lower strain is needed in fatigue for cracks to be formed.

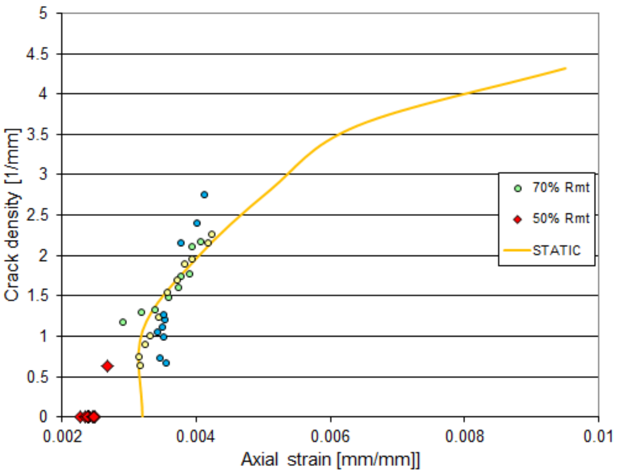


Fig. 7. Relationship between axial strain of [±60]$_S$ samples and corresponding matrix crack density for both static (line) and fatigue loading for increasing number of cycles 1-100,000 (dots) for two loading levels.

V. NUMERICAL MODELING

The procedure of numerical modeling is shown for [±60]$_S$ only. Therefore, transverse matrix cracking failure mode is considered the only failure mode. If the laminate contained 0° or close to 0° plies progressive fiber failure model would have to be employed. Delamination and final failure of the laminate are outside the scope of this research.

A. First ply failure – constituent level

The goal of micro-mechanical modeling is to determine the stress and strain field and any possible damage at micro level, i.e. fiber-matrix level. As a result, failure mode can be determined straight away. Because of the complexity of the composite structure only representative volume element (RVE) is modeled - see Fig. 8. If proper periodic boundary conditions are applied the RVE represents the response of the whole structure due to its periodicity.

$$\varepsilon = M_\varepsilon \varepsilon^{mac} \tag{1}$$

$$\sigma = M_\sigma \sigma^{mac} \tag{2}$$

Strain and stress tensors at micro level are defined by equations (1) and (2) [8], where ε^{mac} and σ^{mac} denotes far field strains and stresses applied to the lamina and M_ε and M_σ are strain and stress amplification factors for macro strain and macro stress, respectively. Matrices M_ε and M_σ can be calculated for any point of RVE using FEA. Maximum vonMises failure criterion was used for matrix cracking in this case.

First, finite element model of RVE was built in Ansys APDL commercial software. Square array configuration was chosen. Linear elastic material model was assumed for both constituents because the primary goal of the analysis is not obtaining effective material properties. Appropriate symmetry and periodicity boundary conditions were applied for normal and longitudinal shear loading. The interface was not modeled separately. It is assumed that matrix strength equals to the strength of the interface when using vonMises failure criterion.

First, on-axis stiffness matrix of the lamina is computed using RVE. Off-axis stiffness matrix of *k-th* ply is calculated using transformation matrix. Relation between laminate mid-plane strain and axial loading force is found using classical laminate theory.

Once the relation between axial loading force N_x and laminate mid-plane strain ε^0 is found, strain ε_{FPF} corresponding to loading force N_{FPF} causing FPF can be computed. N_{FPF} is determined experimentally from tensile tests employing in-situ micro-camera. Alternatively, classical strength criterion for laminae such as Hashin or maximal strain can be used. However, the failure mode of the lamina is not known in that case.

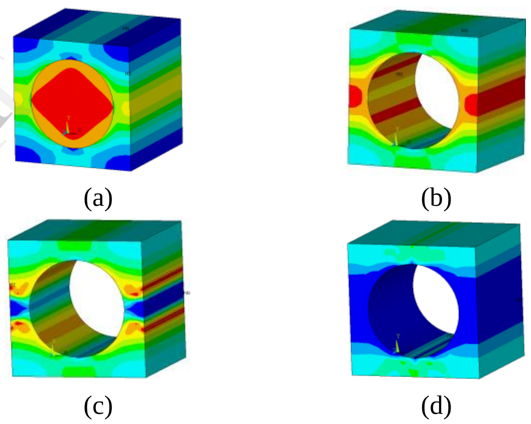
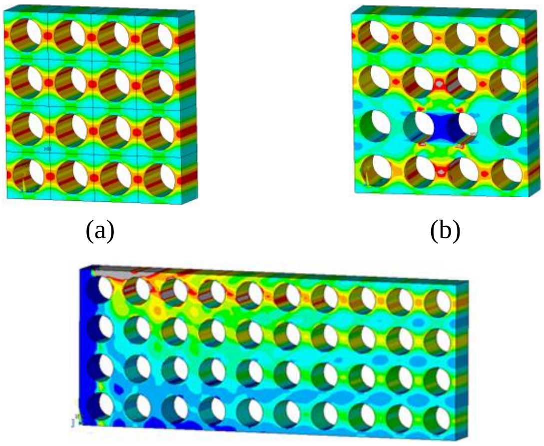


Fig. 8. vonMises stress distribution within the RVE. (a) both matrix and fiber visible (b) only matrix visible (c) initial matrix crack (blue area) (d) propagated matrix crack (blue area)

B. First ply failure – ply level

After initial failure develops stress redistributes within the RVE and subsequent damage grows within the RVE – see Fig. 8. From this moment, the periodicity of the structure is disrupted and stress redistribution in the neighborhood of the cracked RVE must be taken into account. Therefore, RVE does not represent response of the whole structure anymore! The initial failure of the weakest RVE leads to stress concentration in the nearest neighboring RVEs and subsequent damage growth even at constant static load. Therefore, matrix crack instantly propagates through the whole thickness of the ply in the direction parallel to fibers. This crack is called transverse matrix crack because usually it occurs transversally to the loading direction of a particular ply. Since the ply is constrained by the neighboring plies the crack usually stops at the ply interface and the load is redistributed to the adjacent plies. Gradual growth of the transverse ply crack is depicted in Fig. 9. Crack is depicted in blue color. FEA model consisting of several RVEs was used to simulate this process.



(a) (b) (c)

Fig. 9. Gradual growth of the transverse ply crack (depicted in blue color) (a) uncracked matrix (b) first ply failure that leads to stress concentration in adjacent RVEs (c) transverse matrix crack over the whole thickness of the ply

C. Progressive failure – ply level

After the transverse matrix crack in the *k-th* ply form load is redistributed to the adjacent plies. Ply stress in the *k-th* ply decreases in the neighborhood of the crack. However, far enough from the crack stress rises again and new crack is formed. Therefore, right after FPF more cracks in the *k-th* ply form at a certain density. When increasing ply strain, additional strain energy is added to the system and new cracks can form in between existing cracks. The objective of this part of the research was to determine ply strain needed to form new matrix cracks, i.e. to increase matrix crack density. Fracture mechanics approach (FMA) was employed to facilitate such analysis. Strain energy release rate (SERR) is defined by equation (3) if we assume crack formation as a discreet process, where U is strain energy density and A is the area needed to form a new crack. Strain energy density is defined by equation (4) and for linear material model it can be expressed in terms of strain and modulus of elasticity E for uniaxial tension.

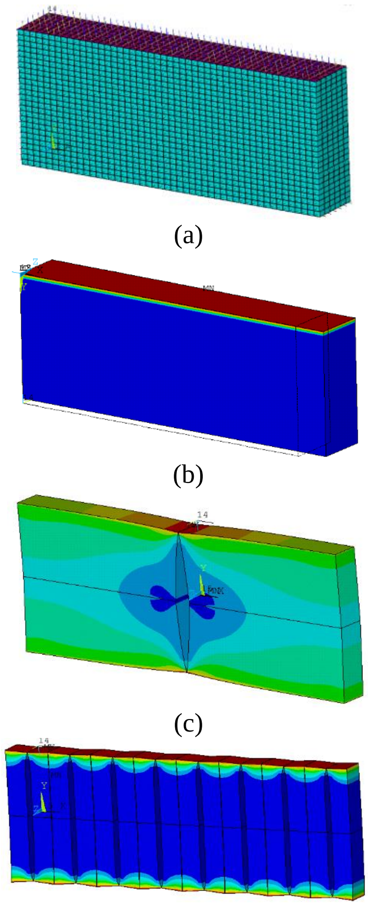
$$G_m = -\frac{\partial U}{\partial A} = -\frac{\Delta U}{\Delta A} \quad (3)$$

$$U = \frac{1}{2}\int_V \varepsilon.\sigma dV = \frac{1}{2}E\varepsilon^2 V \quad (4)$$

It was assumed that the laminate is loaded by in-plane loading. Therefore, the lamina is loaded by longitudinal tensile strain ε_L, transverse tensile strain ε_T and in-plane shear strain $\varepsilon_{LT.}$ Two main assumptions were that longitudinal tensile strain ε_L, does not contribute to transverse crack initiation and that the strain energy release rate remains constant during progressive loading of the lamina. Then, the expression for transverse strain needed to form new matrix cracks can be derived as a function of shear and tensile moduli E, G of a lamina before and after matrix cracking (5).

Periodic FEA meso-model was developed to determine shear and tensile moduli of a ply constrained by the adjacent plies. The model is depicted in Fig. 10. Numerically predicted transverse matrix crack density in comparison with experimental results for $[\pm 60]_S$ laminate is shown in Fig. 11.

$$\varepsilon_{Ti}^2 = 2\varepsilon_{Ti-1}^2 \frac{\left[(E_{i-1} - E_i) + \frac{\varepsilon_{LT}}{\varepsilon_T}(G_{i-1} - G_i)\right]}{\left[(E_i - E_{i+1}) + \frac{\varepsilon_{LT}}{\varepsilon_T}(G_i - G_{i+1})\right]} \quad (5)$$



(a) (b) (c) (d)

Fig. 10. Periodic FEA meso-model of the *k-th* ply constrained by the adjacent plies (a) FEA mesh (b) stress distribution without a crack (c) crack density 0.8 cr/mm (d) crack density 6.4 cr/mm.

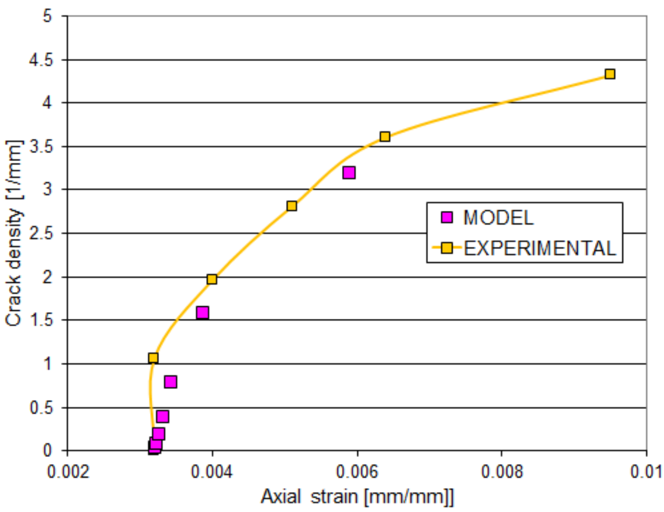


Fig. 11. Numerically predicted transverse matrix crack density in comparison with experimental results for $[\pm 60]_S$ laminate

Once matrix crack density is known as a function of strain, stress-strain curve of the quasi-static test can be constructed employing similar FEA model – see Fig. 12. The nonlinear response of the model is given only by modeling cracks in the meso-model. Material model remains linear elastic. The homogenization of damage or fitting material properties to nonlinear experimental data was avoided at all.

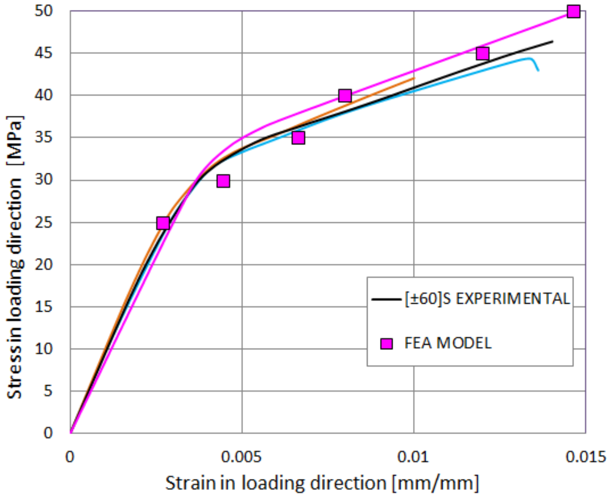


Fig. 12. Stress-strain curve of the quasi-static tensile test constructed employing FEA model in comparison with experimental data

D. Fatigue failure – ply level

In the case of fatigue loading at constant load amplitude and stress ratio, the initial damage develops in the first cycle or during first few cycles. Then, the internal equilibrium is reached. This state of damage is called Characteristic Damage State (CDS) and is characterized by certain density of matrix cracks. Because of the fact, that the ply containing CDS is restrained by the neighboring plies local ply micro-stresses in the neighborhood of damage location decrease. Therefore, no new cracks should develop. However, during cyclic loading matrix fatigue and fracture properties degrade. Its failure envelope shrinks and damage initiates even at smaller stresses. This can be observed e.g. in Fig. 7. This leads to crack density increase and subsequent residual stiffness decline. Same numerical model as for progressive damage is used to predict number of matrix cracks after certain number of loading cycles. Just equation (5) is modified by the degradation factor *D* as SERR of the matrix decreases with increasing number of cycles. Degradation factor *D* is fitted to experimental fatigue data as shown e.g. in Fig. 7.

VI. CONCLUSION

Both quasi-static and load-controlled fatigue testing of E-glass/epoxy samples of various layups have been performed. Acquired data have been used as the input data into the multi-scale numerical model. It has been observed, especially for 90_2°, 60°and 45° plies, that the progressive damage is not a continuous gradual process. Contrariwise, it is a series of discrete events at ply level – abrupt formations of transverse matrix cracks. In fatigue, the initial steep decrease followed by gradual decline of the residual stiffness was observed as expected. Extensive matrix cracking has been observed using in-situ micro-camera in case of $[\pm 60]_S$ and $[0/902/\pm 45/90]_S$ samples. The matrix micro-cracks propagate instantaneously across the entire thickness of the ply and stop at the ply interface where it reaches equilibrium.

First ply failure of the laminate was analyzed employing micro-mechanics FEA model. Both matrix crack density and the effect of crack formation on global mechanical response of the laminate were simulated using FEA meso-model. Real failure modes were modeled. Avoiding the homogenization of progressive damage usually done by employing complicated material models will allow general use of the model for an arbitrary layup and stacking sequence and should significantly reduce the scope of experimental testing needed to obtain input data. FEA multi-scale modeling also allows using nonlinear material model for a matrix or a fiber. It has been concluded that the stress redistribution after FPF must be taken into account to obtain reliable prediction of damage evolution.

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