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Effects of resin-rich areas on failure initiation in carbon-epoxy composites using finite element analysis

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Abstract

Resin-rich areas in composite laminates can occur as inter-laminar resin 'layers' between plies or as intra-laminar resin 'pockets' within a single layer. In this work, numerical methods are used to study the effects of resin pockets on the transverse stiffness and failure initiation of carbon-epoxy composites. Random, or non-uniform, representative volume elements (RVE) with and without embedded resin pockets were studied. Three different types of samples with predefined volume fractions (V_f) were analyzed, and data relating to the influence of resin pockets on homogenized stiffness and the strain at which failure initiates was collected and reported. Based on a control sample for each volume fraction, two methods were used to create RVE samples with resin pockets. In one, the distances between fibers were maintained and fibres removed to create the resin pocket, with a corresponding decrease in (V_f). In the second method, the V_f was maintained and fibers were moved to create the resin pocket, with a corresponding reduction in the distance between fibers. It is shown that intra-laminar resin pockets can reduce both the stiffness and the failure strain of composite materials. Stiffness was reduced in samples where the resin pocket resulted in a reduced volume fraction. For samples with the same volume fraction, particularly for high V_f composites (e. g. 60%), the failure initiation strain in the matrix was, on average, 20% lower for samples with resin pockets compared to samples without resin pockets.

Keywords:

resin-rich areas, micromechanics, representative volume element (RVE), finite element, defects, failure initiation

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1. Introduction

The study of stress concentrations at the microstructural level in composite materials can provide valuable information about the failure behavior of these materials at larger scales. The microstructural level is focused on the behavior within a lamina and is associated with individual fibers and the surrounding matrix. The necessity to study composites at this level arises from the multi-phase nature of the material, where the inhomogeneous morphology and the interactions between phases have a significant influence on the structural properties of the material at both lamina and laminate levels. The current study considers two phases consisting of carbon fibers and epoxy resin, and assumes perfect bonding between phases with a random, non-uniform distribution of fibers within the resin. The bonding between phases of composites play an important role in crack propagation of composites; however, for failure initiation it is more dictated by microstructure than bonding. Structural design allowables for composite materials are developed based on standardized testing at the coupon level, and a significant challenge to the establishment of consistent and reliable values is the variability of test results across samples. While variations of stiffness are often insignificant, experimental values obtained for the strength and failure behavior of composites can be inconsistent. Some of these inconsistencies can be traced back to variations in microstructural morphologies and non-uniformities at the micro scale[1, 2, 3]. Other contributing factors include manufacturing conditions and material variations, and the study of all such factors contributes to the goal of uncertainty quantification which in turn paves the way towards more reliable and robust design. Efforts such as the World-Wide Failure Exercise (WWFE) and its subsequent recommendations [4] are examples of the robust design of composite materials for failure.

Composite materials can be designed with specific mechanical properties to suit particular applications. Targeted properties include stiffness-to-weight ratio, mechanical strength, fatigue and durability, energy absorption, or even the currently fashionable look of carbon fabrics. Design approaches where strength and failure load are the driving criteria can be complicated and are often conservative because of the variations in material properties obtained during characterization testing. Factors that contribute to variations in strength include manufacturing parameters, geometric imperfections, interlayer bonds, fiber/matrix bonding behavior, non-uniform distribution of fibers in the cross section, and variations in the diameter and mechanical properties of individual fibers. Some of these factors that contribute to variations in strength are at the microscale and some at the fiber/matrix microstructural level.

Early microstructural studies of composite materials utilized a unit cell analysis con-

sidering only a uniform distribution of the inclusion phase [5]. These models have now evolved and current studies consider increasingly realistic models incorporating geometric non-uniformities [6, 7]. A realistic microstructural model also allows for a thorough study of mechanical behavior under any combination of loads, and can be used to facilitate design without extensive experimental procedures [8]. The objective of the current work is to quantify some of the uncertainties associated with these non-uniform morphologies, specifically intralaminar resin-rich areas and their effect on the initiation of failure.

The current work studies failure initiation in composites by analyzing 100 different computer-simulated microstructural samples in order to gather information on the sensitivity of failure initiation to changes in the random microstructure of a unit cell. In general, failure in fiber-reinforced composites initiates in the matrix phase with matrix-dominated modes of failure [9, 10]. Thus, for the analysis of the onset of failure in composite materials, the fiber can be assumed to remain in the elastic regime, and the matrix phase (epoxy) can be assumed to be elasto-plastic with damage-law behavior [8, 11]. The current study focuses on failure in the transverse direction (perpendicular to the fibres), and examines the effects of resin rich areas on failure initiation in the matrix phase. It has been shown that a realistic representation of the non-uniform distribution of fibers within the matrix of fiber-reinforced composites can have a significant impact on the magnitude and distribution of stress concentrations in finite element models of the unit cell [12]. Similarly, one could expect that this non-uniform microstructural morphology will also affect failure initiation and progression. The initialization of matrix failure under transverse load has been studied by Fiedler et al. [13], who provided a basis for the use of material modeling in the study of failure and failure initiation in composites. However, the micromechanical models used in their work were limited to cylindrical, hexagonal, and composite (similar to square) RVEs. More realistic constitutive models have since been developed to address the matrix constitutive behavior [8] and geometrically non-uniform RVE microstructures [6, 7, 11, 14, 15]. A distinct feature of carbon reinforced composites that has not been widely studied is the resin rich area, or resin pocket. In the current work, resin pockets are added to computer generated models to create a microstructure with both non-uniform microstructural morphology and resin pockets typical of those found in laboratory samples. The effect of the resin pockets on the mechanical properties of the microstructure is quantified.

Composite materials can contain a number of different defects, where the type of defect is predominantly influenced by the manufacturing method. Defects can occur in the composite material itself as well as in components manufactured from composite materials. Defects

can be challenging to detect and their effect on structural properties difficult to predict. The tolerance for defects usually depends on the structural application. For example, the tolerance for defects in aerospace composites is generally much lower than for composite applications in other industries. Defect sensitivity also depends on the design of the structure and the loading conditions. For example, a homogeneous structure designed for homogeneous load (i. e. cylindrical shell under axial load) is extremely sensitive to imperfections [16] and, as a result, these types of structures are subjected to extensive inspection for the detection of defects.

There are several types of defects at the scale of the fiber and matrix (microscale) including fiber/matrix debonding, fiber waviness, dry fibers, and resin rich areas. Fiber waviness is a common defect that directly reduces the stiffness and strength of composites [17, 18, 19]. Unlike the fiber waviness defect, studies on the effect of resin pockets in microstructures are not widely reported in the literature. Sanei et al. proposed a method for generating synthetic microstructures that include defects such as non-uniform fiber distribution and resin seams [20]. In another work [21], the same authors used a multiscale approach to predict failure initiation and progression. They emphasized the development of a stochastic response for reliability analysis of failure rather than the more commonly employed average-based failure envelopes. Their study used 100 different computer-simulated microstructures, which is the same number used in the current study.

Another parameter that is influenced by non-uniform fiber distribution and resin pockets at the microscale is the distance between fibers. The effects of inter-fiber spacing on residual stress and failure was studied in [22]. Yang et al. deduced that, at higher volume fractions, because of the decrease in minimum inter-fiber distance, the residual stress increases which in turn reduces the stress and strain at failure initiation [22]. The same study also concluded that because residual stresses generally maximize at the loci where inter-fiber spacing is minimal, and because residual stresses contribute significantly to the initial failure in the matrix, inter-fiber spacing both directly and indirectly affects failure initiation. Bulsara et al. [1] studied damage initiation in non-uniform ceramic composites for different types of loads including tensile and thermal, and they demonstrated that when RVEs are subjected to off-axis loads (normal to the fibers), failure initiation is relatively insensitive to the radial distribution function of the fibers. The radial distribution function is a measure of the probability of the distance of the fibers from a reference fiber. They speculated that the low mismatch between elastic moduli of fiber and matrix in ceramic composites could be responsible for this small sensitivity. Hojo et al. studied the effects of irregularities in the

microstructure on the failure initiation and interfacial normal stresses [15]. They studied mechanical and thermal loadings on both uniform and non-uniform microstructures. For the case of mechanical loading they found out that the maximum stress occurs where fibers are close together and the fiber pair is aligned with the loading direction. However, they also concluded that the maximum stress is more affected by inter-fiber distance than by angle. In the current work it is shown that the failure criteria is sensitive to both the distance between pairs of fibers as well as the angular distribution with respect to the load direction, and that the failure initiation is *more sensitive* to the angular relationship between pairs of fibers and the loading direction than it is to the distance between the fibers.

2. Finite element analysis

Random, or non-uniform, microstructures were created using an algorithm similar to [23] that has been shown to be representative of the actual microstructure of composites [24]. The algorithm was modified to include one or more resin pockets of varying sizes and shapes. Two methods were used to create that: in one, a fibre volume fraction was kept as the original, and in the other parts of the original microstructure were kept. The details are further explained in Sec. 3.1. Once the microstructures are created, they are reproduced in Abaqus [25] for finite element and failure analysis. Because failure in composite materials is expected to initiate in the matrix phase, the fiber phase is considered elastic with perfect bonding between the two phases, and failure in the matrix phase is detected using damage models.

In microscale analysis, choosing the appropriate size of Representative Volume Element (RVE) is essential to an accurate and representative analysis. For the case of fiber-reinforced composites, the RVE size is normally represented as the ratio of the length of the RVE to the fiber radius ($\delta = l/r$). An acceptable size of RVE is one that is statistically representative of both the morphology and behavior of the material [26, 27, 28, 29]. For the purpose of this study the ratio of $\delta = 40$ was chosen, which has been shown to be representative of carbon epoxy composite materials [12, 28].

Two sets of boundary conditions were applied to the RVE, one for the application of a deformation and the other to establish a set of periodic boundary conditions (PBC). A detailed description of PBCs for composite materials can be found in [30]. PBCs are applied on the edges of the RVE as equation boundary condition according to the following,

$$\begin{aligned} u(0, y) - u(l_1, y) &= \varepsilon_x l_1, \\ v(x, 0) - v(x, l_2) &= \varepsilon_y l_2 \end{aligned} \tag{1}$$

where u is deformation in the x -direction, and v is the deformation in the y -direction, and l_1 and l_2 are the length along the x - and y -direction respectively. Applying such boundary conditions ensures that the distribution of stress (or strain) is periodic in the composite, simulating the overall behavior of the composite material at a larger scale. The study used many samples in order to gather statistical histogram data representative of the geometric differences between RVEs both in terms of fiber distributions and the presence of resin pockets.

The non-linear behavior of the epoxy is modelled using a paraboloidal yield criterion [13] defined by the following yield function:

$$\Phi(\sigma) = 6J_2 + 2I_1(\sigma_c - \sigma_t) - 2\sigma_c\sigma_t \tag{2}$$

where J_2 is the second invariant of the deviatoric stress tensor, I_1 is the first invariant of the stress tensor, σ_c and σ_t are respectively the compressive and tensile yield strength of the epoxy. As a uni-directional composite is homogenous along the fiber direction, the material has a uniform deformation along this direction. Therefore, it is logical to assume a 2D dimensional representation (i.e. generalized plane strain) of composites for analysis of stress concentrations and failure initiation. For the case of plane strain where $\sigma_{33} = \nu(\sigma_{11} + \sigma_{22})$, the tensors of J_2 and I_1 reduce to:

$$J_2 = \frac{1}{3} (\sigma_{11}^2 + \sigma_{22}^2 - \sigma_{11}\sigma_{22} + \nu^2(\sigma_{11} + \sigma_{22})^2(\nu - 1) + 2\sigma_{12}^2) \tag{3}$$

$$I_1 = (1 + \nu)(\sigma_{11} + \sigma_{22}) \tag{4}$$

The values of σ_c and σ_t for epoxy resin are 114.5 and 47.0 MPa, respectively, and are taken from Fiedler et al.[13]. At small strains the value of the paraboloidal yield criterion, Φ , is an approximately linear function that is negative-valued and increases with increasing applied strain, becoming zero-valued at locations in the matrix where failure initiates. Analyses were performed at two different strain levels for each RVE. Based on the maximum value of Φ in the matrix and the associated value of applied strain, a linear extrapolation or interpolation was performed to find the strain at which Φ turns to zero, and that value was considered the strain at which failure initiates in the matrix phase. It is worth noting where "strain

at failure initiation” stated, it is the macro-strain applied on the RVE samples where the matrix phase starts to fail (yield).

3. Results

A large variety of computer-generated microstructural RVE samples representative of the actual microscopic morphology of composite materials were used in this work. The RVEs were studied with the objective of quantifying the effects of resin pockets on the failure initiation in the composite material. A large number of samples provides a spectrum of data that includes morphological changes in fiber distributions and resin pocket geometry. The results provide information on how local variations in volume fraction changes the failure behavior of the composite. In addition, the study demonstrates the sensitivity of the failure criterion to distances and angles between neighboring fibers.

3.1. Resin pockets

The resin-rich area is a phenomenon (or arguably a defect) that occurs at different scales in composite materials. At the laminate level, resin-rich area occurs at T-joints, ply-drops, under curved yarns of textile composites, and at curved concave surfaces. Another form of resin-rich area, often identified as a ‘resin pocket’, occurs at the intra-laminar fiber/matrix level (microscale). These resin-rich areas are small pockets of resin that form during the manufacture of the composite prepreg material itself. When the resin is added to the fibers, the resin flow interacts with the non-uniform structure of the fibers to create small pockets of resin. Figure 1 shows two microscopic images in which one has a relatively consistent distribution of fibers and the other contains resin-rich areas. Resin pockets result in inconsistent fiber microstructure and create regions vulnerable to stress propagation. Ideally, manufacturers aim for consistent microstructures in which the local fiber volume fractions at different loci are the same as the global fiber volume fraction. Because resin pockets are a non-uniform and probabilistic phenomenon, their effect must be studied using a stochastic approach. In the current study, failure initiation analysis was performed using 100 different computer-simulated microstructures with geometrically varying resin-rich areas in order to quantify the effect on failure initiation.

Two approaches were used to create a resin-rich area in each of the microstructural samples. In the first approach, a sample with a random microstructure but a predefined volume fraction (V_f), identified as type A, (Figure 2A) is used as a baseline. The type A RVE is modified to a type B by removing a number of adjoining fibers (marked by a cross in

the Figure 2A) to create a resin pocket (Figure 2B). As a consequence, the type B samples have a lower V_f and lower overall stiffness compared to the type A samples. The second type of resin pocket sample, type C, are RVEs with the same V_f as the type A samples but having a 'resin pocket' region in which no fibers were added. Finite element analyses were performed on type A, B and C microstructures at volume fractions of 60, 50, and 40%, where 100 different microstructures were analyzed for each case.

The rationale for using the two types of resin pocket samples was to provide a basis for comparing the effect of the resin pocket in isolation without any change in fiber distribution to the effect of the pocket when the volume fraction is maintained but the RVE morphology modified. Type B samples have the same microstructural morphology as the type A random samples except in the resin pocket area, which provides a basis for comparing the effects of resin pockets only, and removes effects such as minimum fiber distance from the failure initiation analysis. In type C samples, the volume fraction is kept the same as the predefined V_f of the type A samples. When the V_f is kept the same and a resin pocket added, the fibers are forced closer to each other, creating fiber-rich areas and increasing the probability of failure at lower strains.

The method used to create the resin pockets provides for fibers to be removed in the area of two overlapping circles defined using randomly selected distances, radii and locations as shown in Figure 3. This method creates resin pockets that are similar in shape to actual resin pockets observed in composite micrographs. The size and orientation of the resin pockets are varied from one sample to the next. It can be seen in the type B samples (Figure 3B) that the von Mises stress distribution is changed compared to the type A randomly distributed samples.

Figure 4 shows the histogram of the strain at which failure initiates for 100 RVEs of each sample type at three different volume fractions. The three graphs in Figure 4 show that, for samples without resin pockets (type A) and the same samples where fibers have been eliminated to form a resin pocket (type B), there is no significant change in the failure strain. The type B samples have significantly lower stiffness than the type A as a result of the fiber removal, and yet their failure initiation strain is similar. This suggests that failure initiation is a local phenomena that is dependent on the local morphology of fibers and matrix rather than volume fraction.

When the samples without resin pockets (type A) are compared to the samples with added resin pocket and the same V_f (type C) as shown in Figure 4A, there is a notable decrease in the strain at failure initiation for the type C samples. This difference in failure

strains between type A and type C samples is less significant for lower volume fractions because at lower V_f the resin pockets influence the local fiber configuration to a lesser extent.

The values for failure strain, average stiffness and coefficients of variation are provided in Table 1. The failure strain values for all three V_f s show that the failure behaviour of type A and B samples do not differ significantly even though their stiffness values are not similar (sample type A are about 8% stiffer on average compared to sample type B). Comparing type A and type C samples at V_f s of 60 and 50%, the difference in strain at failure initiation is close to 20% but their stiffness values are similar because they both have the same number of fibers. This demonstrates that resin pockets in composites with high volume fractions (over 50%) results in inconsistent microstructures, and that the resin pockets create fiber-rich areas causing failure initiation at lower strains. At the lower V_f of 40% the resin pockets do not necessarily create fiber-rich areas, and the existence of the resin pocket does not contribute to any significant difference in the failure behavior for all three types of samples (A, B, and C). Moreover, null hypothesis for different V_f are tested and results are reported in Table 2. As it can be seen, the p -values with confidence of 95% (p -values is smaller than 0.05) for all the comparisons are rejected except one comparison. The only distributions that we failed to reject the null hypothesis is where we compare V_f of 60% and distribution of samples without resin pocket, and distribution of samples with resin pocket (eliminated fibers). The result show that for distributions that the V_f s are similar the difference is very significant (the null hypothesis is rejected), leading to very small p -values.

Figure 5 shows the stiffness as a function of failure strain for the three predefined V_f . A single data point for the equivalent hexagonal unit cell (non-random configuration) is included in each figure for comparison. For all volume fractions, the hexagonal unit cell has a higher failure strain than the non-uniform samples because the fibers in the non-random configuration do not get as close to each other. Stress concentrations are created when fibers are close to each other, resulting in failure initiation at lower strains. In addition, the transverse stiffness values for samples with eliminated fibers (sample type B) are lower than those of the other two types.

3.2. Resin pocket size

The size of the resin pocket was also found to have an effect on failure strain. In this study, resin pocket size was measured using a Delaunay triangulation of the fiber center points as shown in Figure 6. A threshold value was chosen that identified the largest triangles, neighboring large triangles were bundled to form a bigger polygon, and another threshold

was then applied to remove smaller polygons. When the two thresholds are chosen correctly, the selected polygons match the resin pocket areas as shown in Figure 6B.

Figure 7 shows the size of resin pocket in the samples as a function of failure strain for both type B and type C resin pocket samples. The results show that for type B samples with eliminated fibers and associated reduced V_f (green data points in Figure 7), there is no evident correlation between the size of resin pockets and failure strain. This agrees with the finding that the failure strain of samples with eliminated fibers showed little or no change compared to random samples. However, for type C samples with added resin pockets and having V_f s similar to the type A random samples (red data points in Figure 7), there is a weak correlation between the size of the resin pocket and the failure strain. For the three volume fractions shown, the failure strain values increase with decreasing resin pocket size, but the data correlation is reduced for lower volume fractions. One of the reasons for this is that at higher volume fractions there is less room for resin pockets and the maximum size of resin pocket is smaller and these samples cannot cover the entire spectrum of resin pocket sizes. As a result, the failure strain in high volume fraction materials is not only more sensitive to the presence of resin pockets in general, but also to smaller resin pockets, and that even small resin-rich areas are critical to the expected failure strain for these types of materials.

3.3. Fiber distance and direction

To better understand the role of microstructural morphology and fiber distribution on failure strain and stress concentration in the matrix phase, a small strain was applied to each of one hundred random samples. For each sample, the two closest pairs of fibers were identified and paraboloidal yield criterion in the matrix phase between the two fibers was applied. The data for the two hundred closest pairs in the one hundred different random samples was gathered and is shown in Figure 8 where the distribution of yield function is shown as a function of the angle between the loading direction and the direction of pairs of fibers. In Figure 8 the colormap indicates the distance between the pairs of fibers, and the paraboloidal yield criterion values are negative as a result of the very small value of applied strain (the negative values mean that the epoxy has not failed). As the value of the angle increases, indicating that the pairs of fibers are less aligned with the applied load, the paraboloidal yield criterion values become increasingly negative, indicating an increase in failure strain. The distance between the fiber pairs, indicated by the colour map in Figure 8, has a lesser effect than the angle with the loading direction and is most significant when the fibers are perpendicular to the loading direction. This is evident in the trend in distribution

of distance from lowest distances to highest distances in 0° (lowest distances are shown with blue and highest ones with red).

Figure 8 shows that matrix regions between fiber pairs that are close together and aligned closely with the load direction are the first candidates for failure initiation. The figure also shows that, for failure initiation, the direction of fiber pairs relative to the load is more significant than the distance between fibers.

4. Concluding Remarks

In this work, a numerical approach is used to investigate the effect of resin pockets on transverse stiffness and failure initiation in composite microstructures. An epoxy matrix material with yield properties introduced in [8] was used. Random, or non-uniform, microstructures with and without embedded resin pockets were analyzed to compare the differences in stiffness and failure strain between samples. Two approaches were used to create resin pockets. The first approach utilized a sample with a predefined V_f and eliminated fibers to create a resin pocket so as not to vary the distance between fibers. In the second approach, the fibers were moved to create a resin pocket with the same V_f as the original sample without a pocket but with a smaller average inter-fiber distance.

Results showed that samples with removed fibers fail at the same or similar strains as the samples without resin pockets despite the fact that they have lower V_f and lower stiffness. Also, samples with resin pockets but the same volume fraction failed at lower strains than samples with no resin pockets.

As the failure initiates at those locations in the matrix that fibers are close to each other, it is worth studying the effects of interface region on failure initiation as well. Specifically, for studying progressive failure in composites it is crucial to include interface region as a third phase.

Comparing hexagonal unit cells with random RVEs, although the stiffness values are comparable, the failure strains obtained are overestimated, demonstrating that structured unit cells are not representative of local phenomena such as matrix phase failure initiation.

The analysis of 100 different samples of three different microstructures with and without resin rich areas provided data that quantifies the effects of this common and probabilistic microstructural feature. This type of study and resulting data could be used in probabilistic and multi-scale analyses of composite materials. A similar methodology of statistical analyses of many algorithmically-developed samples for different manufacturing defects can be developed in the future for composite materials. As defects are probabilistic phenomena

such statistical and probabilistic analysis help to understand and scrutinize their effects on the overall behavior of composite materials and structures.

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Table 1: Average and Coefficient of Variance of strain that failure initiate and homogenized stiffness.

Predefined V_f	60%			50%			40%		
Sample Type*	A	B	C	A	B	C	A	B	C
AVG. failure strain (%)	0.11	0.11	0.09	0.13	0.13	0.11	0.14	0.15	0.14
CV failure strain	4.55	5.87	8.32	4.99	5.60	9.35	6.71	8.17	6.28
AVG. stiffness (MPa)	13.76	12.57	14.00	11.51	10.71	11.71	9.87	9.36	9.91
CV stiffness	0.47	2.58	0.98	0.46	1.83	1.10	0.41	1.65	0.60

*Sample type A: without resin pockets, Sample type B: with resin pocket (eliminated fibers), Sample type C: with resin pocket (added resin pocket).

Table 2: p -values for the null hypothesis (two-sample t-test).

Predefined V_f	60%		50%		40%	
Sample Type*	B	C	B	C	B	C
A	0.33	7.6e-29	0.048	1.16e-25	0.51e-3	0.51e-4
B	—	7.8e-28	—	7.44e-29	—	8.33e-9

*Sample type A: without resin pockets, Sample type B: with resin pocket (eliminated fibers), Sample type C: with resin pocket (added resin pocket).

(A)

(B)

Figure 1: Microscopic images of carbon/epoxy composite. (A) is a sample without any significant resin pocket. (B) is a sample with resin pocket.

Figure 2: Three types of samples analyzed in this study. (A) Samples with random microstructure. The red crosses mark the fibers that were chosen for elimination to create sample type B. (B) Samples with eliminated fibers to create a resin pocket. The rest of microstructure is the same as the sample type A. (C) Samples with added resin pocket and the V_f similar to type A samples.

Figure 3: Eliminating fibers to create resin pockets. The location of p_1 , the distance d , the angle α and the radii of the two circles were chosen randomly from a range. The same type of resin pockets were created where fibers were moved around to keep the number of fibers (and V_f) the same. The lower figures are contour of von Mises stress.

(A) Predefined $V_f = 60\%$

(B) Predefined $V_f = 50\%$

(C) Predefined $V_f = 40\%$

Figure 4: Histogram of strains that failure starts for three different types, and three different predefined V_f . Each histogram represents 100 sample of the specific type.

(A) Predefined $V_f = 60\%$

(B) Predefined $V_f = 50\%$

(C) Predefined $V_f = 40\%$

Figure 5: Change of stiffness versus failure strain for three different predefined V_f .

(A) sample with eliminated fibers

(B) Delaunay triangulation of fiber centers

(C) Paraboloidal yield criterion of resin

Figure 6: (A) Random samples with eliminated fibers to create resin pockets (B) Delaunay triangulation of fiber centers to detect resin pockets, and their sizes (C) paraboloidal yield criterion contour of epoxy resin.

(A) Predefined $V_f = 60\%$

(B) Predefined $V_f = 50\%$

(C) Predefined $V_f = 40\%$

Figure 7: Size of resin pocket area compared to the failure strain for three predefined V_f .

Figure 8: Distribution of paraboloidal failure criteria when the samples are subjected to small value of strain. y -direction values are Paraboloidal failure criteria in the middle of two fibers center points, the x -direction values are the angle between pairs of fibers and direction of load, the colormap is the distance between two pairs of fibers.

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