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Abstract

An improved understanding of the mechanics of stress rupture has been developed by recognizing that the process of damage accumulation and failure is largely the same for both burst failures produced under quasistatic loading conditions and under constant hold pressure stress rupture. Both failures are tensile failure of the carbon fibers in a localized region such that the remaining fibers are overloaded and can no longer support the applied stress. The key parameter leading to failure in both cases is fiber stress. For both quasistatic burst testing and stress rupture testing, fiber stress is initially accumulated in the same manner because of increasing quasistatic pressure. The remaining increase in fiber stress in a quasistatic burst test is simply the result of increasing pressure, while for stress rupture under conditions of constant pressure, the increase in fiber stress is a result of redistribution of stress from the matrix into the fibers and from damage mechanisms such as broken fibers. This simplistic model explains phenomena observed in stress rupture tests where some vessels fail at short times, others at later times, and others not at all. It also explains why stress rupture failures are more likely to occur shortly after reaching the hold pressure, with decreasing failures as a function of time after reaching the hold pressure. Micromechanics modeling is used to show how the stress is transferred from the matrix to the fibers during a hold and is used to estimate the maximum amount of stress that can accumulate in a unidirectional laminate. Further, the modeling is used to show the relationship between measured composite strain and fiber stress. Applying this new understanding, composite strain measurements can provide insight into the fiber stress, and the increase that occurs under constant pressure conditions as the result of viscoelastic effects. Fiber stress increases based on strain measurements for composite overwrapped pressure vessels (COPVs) with multi-directional lamina are provided and compared against the previously estimated maximum fiber stress increase for a unidirectional laminate. Historically, due to a poor understanding of stress rupture physical mechanisms, accelerated life testing has been relied on to estimate stress rupture reliability of carbon fiber COPVs. However, these methods have recently been shown to have significant limitations as applied to carbon fiber COPVs. The improved physical understanding of the mechanics of stress rupture presented here supports proof test screening as an alternate approach to mitigating stress rupture that overcomes the limitations of accelerated life testing.

1.0 Introduction

Stress rupture, also known perhaps more appropriately as creep rupture, is a failure mode that is exhibited in structures comprised of viscoelastic materials. Stress rupture failures occur at periods of time after a structure has been held under constant load/stress. As such, the structure fails at stresses that are lower than the nominal expected strength of the structure. Additionally, a specimen fails in stress rupture at a stress below the strength that specific specimen would have achieved under quasistatic loading until failure (i.e., in a strength or burst type test profile). In addition to resulting in failures at lower-than-expected stresses, stress rupture failures are stochastic in time after reaching hold. When testing multiple nominally identical test articles in conditions designed to induce stress rupture failures, some test articles fail at short times, while others fail at later times, and some test articles do not fail at all. Stress rupture has been of particular concern for composite overwrapped pressure vessels (COPVs) because these vessels are used at extremely high pressures. As a result, COPVs have stored potential energy that is equivalent to pounds of explosive materials. COPVs are widely used in aerospace launch vehicles and spacecraft and burst failures of these vessels are expected to result in catastrophic loss of the vehicle. While there have been COPV failures in operational space vehicles with catastrophic consequences, none have been documented to be solely a result of stress rupture failure.

There is significant literature on mechanistic models developed to predict the deformation (i.e., creep) response of fiber-reinforced composites subjected to conditions of constant stress loading [1]. However, there are few models available to describe the mechanics of how this deformation progresses to stress rupture failure of composites. One exception is the so-called fiber-breakage model [2]. The fiber-breakage model provides a description of the process of stress redistribution that results when fibers break in the composite and the remaining fibers are overloaded in a region resulting in catastrophic failure. While this is a contributing mechanism of the mechanics of stress rupture failure, as described in the following, it is not the complete story. Further, the fiber-breakage model has not been demonstrated to be useful for predicting stress rupture failure or assessing COPV reliability.

This article presents new insights on the mechanics of the stress rupture failure process. It is postulated that the underlying mechanics of a burst failure of a fiber-reinforced composite is a result of quasistatic increasing stress is the same as that which occurs in a stress rupture failure. In both, the stress in the fibers is increased resulting in the failure of weaker fibers until such point that the remaining fibers are overloaded in a region resulting in catastrophic failure. In the simple burst test profile, the increasing stress in the fibers results from increasing applied stress in the composite (per a simple law of mixtures relationship) as well as from redistribution of stress from broken fibers (and other microscale damage mechanisms such as matrix cracking and fiber matrix debonding). In the case of stress rupture, the majority of stress in the fiber leading to failure results from the same mode of loading (i.e., the quasistatic stress increases up to the stress rupture test hold stress). This results in the identical fiber stress increases from overall composite stresses and stress redistribution from microscale damage. The only difference for a stress rupture failure is that once the overall stress is held constant, the remaining increase in fiber stress that leads to failure is driven by stress redistribution from the matrix to the fibers due to viscoelastic relaxation of the matrix and any additional stress redistribution from microscale damage (e.g., fiber breakage). This mechanism of stress redistribution from the matrix to the fibers due to viscoelasticity and its role as a driving mechanism for creep damage in ceramic matrix composites at high temperature was previously mentioned in Reference 3 but not described in detail. As the viscoelastic process is time dependent, the resulting stress rupture failures are also time dependent. If the hold stress is very close to the point at which the fiber stress state will result in overall failure due to quasistatic loading (burst), then the specimen will fail after a short time during hold. If the hold stress is far away in magnitude causing the fiber stress state to reach a point of failure, then it will take longer for a larger stress redistribution to result in failure, or it may not fail at all. Lastly, if the hold stress is very much less than the fiber stress state needed to cause failure, then failure will not occur even after very long holds as the entirety of viscoelastic relaxation occurs and sufficient additional fiber stress is not increased to cause failure.

Simple micromechanical modeling is presented to demonstrate this proposed stress rupture failure phenomenon. The modeling describes a unidirectional composite specimen configuration in which the matrix is viscoelastic and the fiber is not (e.g., as in a carbon fiber epoxy matrix composite). Quantitative results are provided for the model for typical carbon fiber and epoxy matrix properties which predict the maximum additional fiber stress and strain that might be expected for this viscoelastic relaxation process. Additionally, it is shown that the percentage increase in fiber stress as a result of viscoelasticity in the matrix during a stress rupture hold is the same as the percent increase in overall composite strain that can be measured with strain gauges on the specimen. Further, it is described how this same approach can be applied to multidirectional lamina composites such as COPVs, and experimental results comparing COPV strains during stress rupture are discussed. Additional aspects of this approach to modeling stress rupture and how they compare with previous experimental results are also discussed. One of note is how the stress increase in fibers from the viscoelastic effects are at their maximum rate when the hold stress is first reached and then the rate decreases exponentially with time toward an asymptotic stress level. As such, it is expected that the rate of stress rupture failures will be highest as the stress rupture hold stress is reached and then decreasing to a point where eventually no failures are detected, which is consistent with observations in such tests in both composite fiber strand specimens as well as subscale COPVs.

Lastly, it is asserted that if burst and stress rupture failures are manifestations of the same underlying mechanical phenomena, then the distribution of burst strengths measured from burst testing can provide insight into stress rupture failure. In fact, the observed stochastic behavior of stress rupture failure times (or no failure) is simply a reflection of the variability (i.e., the distribution) in burst failure strengths of the population of COPVs.

2.0 Mechanics of Stress Rupture Failure

The conventional approach to assessing the risk associated with stress rupture of COPVs involves accelerated testing by exposing the vessels to much higher than operational pressures (i.e., above proof pressures) for time periods much shorter than the planned operational lifetimes. Failures under these conditions are recorded as a function of time at a constant hold pressure, which allows the reliability (i.e., 1 – the probability of failure) to be estimated. By trending the failure times at multiple hold pressures, these reliability estimates are extrapolated down in pressure to the operational pressure levels and out in time to the planned mission duration. Several issues with this approach have been recently identified [4]. One rationale for why this approach has been previously used is that the physical mechanism of stress rupture failure was not well understood. However, it is asserted in this article that the underlying physical mechanism for stress rupture failure is the same as that for a nominal tensile burst failure. Namely, they are both tensile failure mechanisms occurring when the tensile stress in the fibers reaches a point in a region of the composite such that the remaining fibers are overloaded beyond their tensile strength resulting in catastrophic failure. The only difference between a nominal burst failure as a result of quasistatic increasing pressure and a stress rupture failure is how the final increment of increasing fiber stress is achieved. In a burst test profile, increasing pressure increases fiber stress. When holding at a constant pressure in stress rupture test profile, there is a redistribution of stress to the fiber as the matrix relaxes viscoelastically.

The mechanics leading to tensile failure of a fiber-reinforced composite as a result of quasistatic increasing stress are reasonably well understood. As the applied stress is increased on the composite, there are correspondingly increasing stresses in both the fiber and the matrix. However, micromechanics analysis demonstrates that since the stiffness of the fiber is much higher than that of the matrix, the stress in the fiber correspondingly increases much higher than that in the matrix. As designed, the strength of the fiber is much higher than the matrix, and so, the fibers provide the dominant load carrying and strengthening component of the composite system. As the overall stresses in the composite increase, and the corresponding stresses in the fiber and matrix constituents increase, microdamage mechanisms occur. These include localized matrix cracking, fiber-matrix debonding, and fiber failure for lower strength fibers. As these damage mechanisms occur, the strain energy stored in these components is also released resulting in localized increased stress in nearby fibers. Eventually, the fiber stress will reach a point in some larger region where the remaining unbroken fibers are unable to support the stress (i.e., the tensile strength of the fibers in that region is exceeded), which results in overall or catastrophic failure of the composite. The driving force for the increased fiber stress is the increasing applied composite stress in concert with the smaller additional localized stress redistribution from localized damage mechanisms.

The mechanics of stress rupture failure are nearly identical, except for what happens during the hold at constant stress or pressure. Prior to initiating the hold, a specimen undergoing a tensile strength test and one undergoing a stress rupture test experience the exact same loading conditions. Thus, if it were possible to test two exact identical specimens, one undergoing stress rupture testing and one undergoing a strength to failure test, at the point when they both reached the hold stress, they would be expected to have exactly the same fiber and matrix stresses as well as the same accumulation of microdamage. In the tensile failure test, as the stress is allowed to continue to increase above the stress rupture hold stress, the process described in the previous paragraph occurs with increasing fiber stress until failure. The process is slightly different in the specimen when the applied composite stress is held constant. As will be shown below, during a hold at constant stress, viscoelasticity in the matrix results in the stress in the matrix decreasing (i.e., the matrix modulus decreases with time resulting in decreasing stress in the matrix with time). Since the overall composite stress must remain constant, this results in increasing stress in the fibers with time. If the hold stress at which the stress rupture test is performed resulted in a fiber stress sufficiently close to the condition at which the specimen would fail, this small additional fiber stress increase due to viscoelasticity (and corresponding fiber stress increases from any additional microdamage that occurs as a result of this fiber stress increase) will result in the overall failure of the specimen. In both cases, though, increasing fiber stress to the point that the strength of the fibers is overloaded in a region is the mechanical cause of the failure.

Micromechanics modeling of a unidirectional composite can be used to illustrate this phenomenon of stress transfer between the matrix and fibers when a composite is held at constant stress. This simplified model also provides a tool to estimate the magnitude of additional stress that can develop in the fibers. When the unidirectional composite is loaded in tension, a condition of strain continuity must exist such that the strain in the fiber (ε_f) is equal to the strain in the matrix (ε_m) as described in Reference 5. If this condition does not exist, then the fiber and matrix are debonded. While the strains are the same, the stresses for the fiber (σ_f) and the matrix (σ_m) are different and given by:

$$\sigma_f = E_f \varepsilon_f \text{ and } \sigma_m = E_m \varepsilon_m$$

where E_f is the modulus of the fiber and E_m is the modulus of the matrix. Since for a typical composite (and especially for a carbon fiber composite), the fiber modulus far exceeds the matrix modulus, the fiber stress is much higher than the matrix stress. That is, the fiber is the dominant load carrying constituent for the composite system, which is desirable since it is also the higher strength component. The relationship between the applied stress in the composite (σ_c) and the stresses in the fiber and matrix components can be shown to be:

$$\sigma_c = f\sigma_f + (1 - f)\sigma_m$$

where f is the fiber volume ratio. Further, the composite stiffness (E_c) is given by:

$$E_c = fE_f + (1 - f)E_m$$

Since carbon fibers are not viscoelastic, the stiffness of the fiber does not depend on time. However, for a viscoelastic polymer matrix, the stiffness of the matrix does depend on time held at load. For viscoelastic (i.e., creep) strain to occur at constant stress, which is what is observed in polymers, the stiffness must decrease (or relax) as the creep strain occurs. Therefore, the stiffness for the composite as a function of time at constant stress can be rewritten as:

$$E_c(t) = fE_f + (1 - f)E_m(t)$$

As a result, the time-dependent strain in the composite is:

$$\epsilon_c(t) = \frac{\sigma_c(0)}{E_c(t)}$$

where $\sigma_c(0)$ is the hold stress at time = 0. Note that $E_c(t)$ will decrease as $E_m(t)$ decreases with time during a hold at constant stress. Therefore, the overall composite strain will increase with time, which is consistent with the observation of creep strain that has been reported in the literature for composite coupon specimens and COPVs in References 6 and 7.

The continuity of strain condition continues to apply as the composite creep strain accumulates, and thus the fiber strain, matrix strain, and composite strain as a function of time all remain equal, or:

$$\epsilon_c(t) = \epsilon_m(t) = \epsilon_f(t)$$

The resulting fiber and matrix stresses are given by:

$$\sigma_f(t) = E_f\epsilon_f(t) = E_f\epsilon_c(t)$$

and

$$\sigma_m(t) = E_m(t)\epsilon_m(t) = E_m(t)\epsilon_c(t)$$

Although the fiber stiffness does not change with time during the hold at constant stress, the fiber stress does change. It increases with time because the fiber strain is increasing with time due to the viscoelastic creep of the matrix. As such, to maintain overall constant composite stress, the matrix stress decreases with time. In essence, the effect of viscoelasticity is shedding (or transferring) the stress developed in the matrix during the quasistatic loading from the matrix into the fiber. This increase in fiber stress can lead to failure over time during hold if the fiber stress developed during the loading to the hold stress is sufficiently close to the failure strength, even though the overall composite stress is held constant.

The graphs in Figure 1 provide insight into how this process leads to different failure times (or no failure) during a stress rupture test depending on how close the stress is to the failure strength upon reaching the hold stress. In a), three different loading scenarios are shown with the red solid line illustrating a case where the stress upon reaching hold is very close to the strength of the specimen. The blue dotted line illustrates a case where the stress at hold is slightly lower, while the green dashed line illustrates that the stress at hold is significantly less than the strength of the specimen. Plots b) and c) show the corresponding composite creep strain and resulting fiber stress for these three cases. As can be seen, the increase in fiber stress for the case in red (solid line), where the stress at hold is close to the strength of the specimen, leads to a stress rupture failure in a short time period. The case of the stress in blue (dotted line), which is slightly lower relative to the strength of the specimen, also results in a stress rupture failure, but takes a longer period of time to occur. Lastly, for the case in green (dashed line), where the hold stress is significantly less than that of the strength, the specimen does not fail even for very long periods of time at the hold stress. These graphs show all specimens as having the same strength but loaded to different stress levels for the purpose of explaining the phenomenon. However, in actual testing, the stress level at hold is held constant for a number of specimens, but the strength of the individual specimens varies. In that scenario, if the strength of the specimen is only slightly above the stress level, it would correspond to the case in red (solid lines) where failure would occur in a short time, etc. Thus, this simple micromechanics model explains and is consistent with the observations of the stochastic nature of stress rupture failures observed in testing. The variability in stress rupture failure times (or the case of no failure) reflects the variability in strengths of the individual specimens relative to the stress at which the stress rupture testing occurs.

This simple model is consistent with the characteristics of stress rupture failure observed in testing in that more stress rupture failures are observed soon after the stress rupture hold begins, and there are decreasing numbers of failures as time on hold increases. This is explained by the rate of increasing creep strain, which corresponds to the rate of increasing fiber stress due to matrix viscoelasticity, which is at a maximum at the start of the stress rupture hold. As observed in Figure 1 c), the rate of increasing stress decreases with time and as a result the observed rate of stress rupture failures decreases with time. The results of this testing observation are that the estimated stress rupture reliability decreases at a faster rate at the beginning of the stress rupture hold and then deteriorates more slowly with increasing time.

If the fiber modulus and the matrix stiffness as a function of time were known along with the fiber volume fraction, then the composite strain could be predicted based on the above equations and compared with measured strains. No such data were available for analysis in this study. However, the model can be used to predict the maximum amount of additional fiber stress that can be generated from this viscoelastic phenomenon of transferring matrix stress to fiber stress. In this case, all of the matrix stress that was developed upon reaching the hold stress is transferred to additional fiber stress such that the matrix stress decreases to 0 during the hold time.

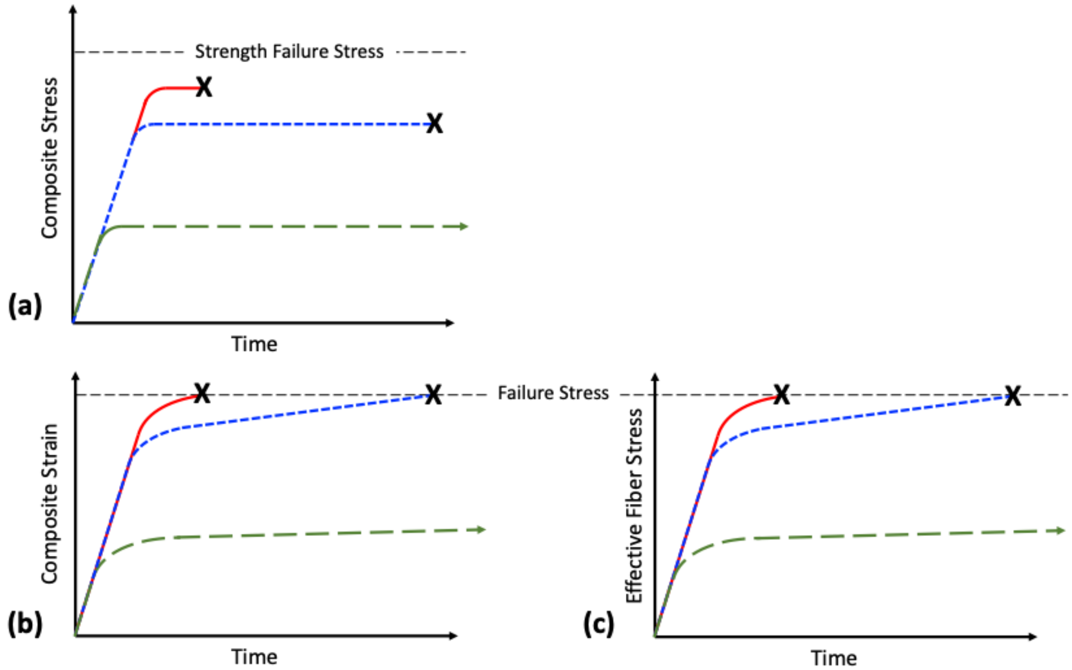

Figure 1. Composite Stress (a) and Strain (b), and Fiber Stress (c) as a Function of Time for Different Hold Stresses Until Stress Rupture Occurs Indicated by 'X'

For illustration, in the case of the unidirectional composite, the following typical properties for carbon fibers and epoxy matrix are assumed:

$$E_f = 291\ GPa, E_m(0) = 5.69\ GPa, and\ f = 0.6$$

Based on these properties, the resulting composite modulus (at the start of the hold stress, i.e., t=0) is $E_c(0) = 176.9\ GPa$. Also, at t = 0, the composite stress upon reaching hold $\sigma_c(0)$, is:

$$\sigma_c(0) = f\sigma_f(0) + (1-f)\sigma_m(0)$$

Rearranging to solve for the fiber stress at time 0 results in:

$$\sigma_f(0) = \sigma_c(0)\frac{1-(1-f)\frac{\sigma_m(0)}{\sigma_c(0)}}{f}$$

which can be rewritten in terms of the moduli as:

$$\sigma_f(0) = \sigma_c(0)\frac{1-(1-f)\frac{E_m(0)}{E_c(0)}}{f}$$

Substituting the previously assumed properties and fiber volume ratio results in:

$$\sigma_f(0) = 1.645\ \sigma_c(0)$$

The matrix stress upon reaching hold (i.e., at t = 0) can be expressed as:

$$\sigma_m(0) = \sigma_c(0)\frac{1-f\frac{\sigma_f(0)}{\sigma_c(0)}}{(1-f)}$$

which, upon substituting for $\sigma_f(0)$ based on the assumed material properties, results in:

$$\sigma_m(0) = 0.0322\ \sigma_c(0)$$

During hold, the modulus of the matrix decreases over time, and in the theoretical limit as time approaches infinity, the matrix modulus goes to 0 (i.e., $E_m(\infty) = 0$). In this case, the matrix no longer supports any load and has transferred all of its stress into the fiber and thus $\sigma_m(\infty) = 0$. At which point, the resulting composite stress becomes equal to the fiber stress as:

$$\sigma_c(\infty) = f\sigma_f(\infty) + (1 - f)\ \sigma_m(\infty) = f\sigma_f(\infty)$$

and thus:

$$\sigma_f(\infty) = \frac{1}{f}\sigma_c(\infty)$$

However, by definition, since the testing is run at constant composite stress, then $\sigma_c(0) = \sigma_c(\infty)$, and so the limiting fiber stress is:

$$\sigma_f(\infty) = \frac{1}{f}\sigma_c(0)$$

Therefore, for our example case of material properties and fiber volume ratio, the normalized increase in fiber stress (i.e., the increase in fiber stress divided by the fiber stress at the start of the hold) is given by:

$$\frac{\sigma_f(\infty) - \sigma_f(0)}{\sigma_f(0)} = \frac{\frac{1}{0.6}\sigma_c(0) - 1.645\ \sigma_c(0)}{1.645\ \sigma_c(0)} = \frac{\frac{1}{0.6} - 1.645}{1.645} = 0.013 = 1.3\%$$

So, for a unidirectional lamina case with the stress applied in the fiber direction, whatever fiber stress is reached when the stress is increased from zero to the hold stress, there will result in an increase of fiber stress by 1.3% if it is held to a long enough time such that all of the matrix stress relaxes into the fiber stress. While this is a small amount, if the fiber stress reached at hold is very near the failure stress for that specimen, then this increase can lead to failure with time. Also, it is noted that this small additional stress results in the unidirectional case. In a COPV, the laminate is comprised of different layers with fibers oriented in different directions. For laminae with the fibers at off axis angles with respect to the loading direction, the amount of viscoelastic creep strain will be increased as is illustrated in Reference 6. This additional viscoelastic creep will result in increased fiber stress for the fibers in the direction of the applied load. Lastly, the above discussion does not account for the additional fiber stress that will accumulate as a result of microscale damage accumulation such as from broken fibers. As illustrated in Reference 6, there is increasing creep strain measured as the hold stress is increased. Proof that this is due to accumulated damage is provided by noting that this increased strain results in permanent deformation that is observed when the specimen is unloaded. As a result of off axis plies and microscale damage, it is expected that the actual increase in fiber stress during a viscoelastic hold will be larger than the 1.3% estimated for a unidirectional specimen held at a stress where significant microscale damage is not created. Experimental results for COPVs held in stress rupture testing are presented below which are consistent with the expectation of larger increases in fiber stress during hold. Before proceeding to those COPV results, the following section discusses how to gain insight into the increase in fiber stress during a stress rupture test.

3.0 Fiber Stress Increase Determination from Composite Strain Measurements

The fiber stress in a composite cannot be directly measured in test. However, the composite strain can be measured directly with strain gauges. It is noted that since the carbon fiber is not viscoelastic, the normalized increase in fiber stress during a stress rupture test hold is equal to the normalized increase in fiber strain. Further, due to strain continuity, the normalized increase in fiber strain equals the normalized increase in composite strain. That is:

$$\frac{\Delta\sigma_f(t,0)}{\sigma_f(0)} = \frac{\epsilon_f(t)-\epsilon_f(0)}{\epsilon_f(0)} = \frac{\epsilon_c(t)-\epsilon_c(0)}{\epsilon_c(0)} = \frac{\Delta\epsilon_c(t,0)}{\epsilon_c(0)}$$

Thus, the normalized increase in composite strain, which can be directly measured in test, is equal to the normalized increase in fiber stress that occurs because of viscoelastic redistribution of stress from the matrix to the fiber.

While the above is developed assuming a unidirectional lamina, the condition of strain continuity can also be applied for a more complicated laminate construction typical of a COPV. In such a case, for strain measured in each fiber direction, the normalized increase in composite strain in the fiber direction is equal to the normalized increase in fiber stress in that direction. For the more complicated laminate, the resulting transfer of stress to fibers in the fiber direction arises as a result of the viscoelastic matrix in the plies in the fiber direction as well as the viscoelastic behavior of off-axis plies, which will lead to even larger contributions to increased fiber stress. As a result, the viscoelastic behavior of a laminate is a function of the laminate stacking sequence as illustrated in Reference 8. Lastly, it is noted that the composite strain measurements will not only capture the viscoelastic strain that occurs, but also the viscoplastic strain that results from microdamage, which further increases the fiber stress during a hold stress. As a result, the composite strain measured provides direct insight into the fiber stress increase that occurs due to viscoelastic and viscoplastic deformation during a stress rupture hold test.

Composite strain measurements were recorded for 17 COPVs that were tested in accelerated life stress rupture experiments [7]. While the tests were only performed for 1385 hours on each COPV, the resulting creep strain was then fit to a statistical model based on the Schapery viscoelastic model form in Reference 8. The model was then extrapolated out for a period of 10 years to provide an estimate of the accumulated normalized increase in composite strain (and fiber stress) that would occur over very long times such that it should approximate an infinite hold. For these vessels, the average percent increase in strain at 10 years was 3.3% with a 95% statistical upper bound of 5.4% [4,9]. As predicted, this is larger than the 1.3% predicted for a unidirectional composite, but on the same order of magnitude. The interpretation of these extrapolated normalized increase in fiber stress estimates is that a particular COPV of this design would need to be loaded in a stress rupture test to a level such that the increase in fiber stress would exceed the fiber strength (i.e., using the 3.3% average, loaded to 96.7% of its strength or higher) for it to fail in stress rupture. Otherwise, the increase in fiber stress during the stress rupture hold would not exceed the strength and would not result in failure.

4.0 Practical Application of Stress Rupture Mechanics

Stress rupture reliability estimates have historically been developed from statistical modeling of accelerated life testing [4]. However, numerous issues have been identified with this approach when applied to carbon-fiber COPVs making the reliability estimates dubious. The improved understanding of the mechanics of stress rupture presented here links static burst strength to the stress rupture failure behavior and suggests an alternate approach to reliability estimation [4]. However, while this alternate approach offers benefits in reduced testing duration, it would likely require subscale test articles to be economically affordable and thereby require similarity rationale to flight vessels. Thus, a strength testing based stress rupture reliability method is also not recommended. Instead, a more direct approach to stress rupture mitigation based on proof testing combined with strain measurements has been proposed [9]. The improved understanding of stress rupture failure mechanics developed herein provides a basis for the acceptance of this proof test screening methodology. In the proposed approach, strain measurements are acquired during proof testing of vessels and during a hold at operational pressure. The strain increase over a short duration at operational pressure can be extrapolated to the desired lifetime, and if it remains below the proof strain with a margin of safety, then stress rupture failure is not expected. Proof screening methodology can be applied on flight COPVs as opposed to surrogate specimens such as strands or subscale vessels, which is in contrast to destructive, failure-based reliability estimation approaches. Proof test screening offers significant advantages over the historical accelerated life testing methodology since it can be performed at the operational pressure, eliminating the need to down-extrapolate a failure-based statistical model derived from excessively high-pressure levels (i.e., well above proof) that are not representative of operational conditions.

5.0 Summary and Conclusions

An improved understanding of the mechanics leading to stress rupture was provided. This improved understanding recognizes the similarities in the mechanics of failure of a COPV under quasistatic burst conditions to one failing by stress rupture. In both cases, the tensile stress in the fiber is the controlling parameter. While the tensile stress in the fiber is accumulated in the same manner (i.e., quasistatic increase in pressure) for both types of failure over the vast percentage of the loading profile, there is a small difference in the final increase in fiber stress between burst testing and stress rupture failure. In the former, the final increase in fiber stress is a result of the increased pressure, while in the latter, the final increase in fiber stress is driven by a redistribution of stress from the matrix into the fibers due to composite viscoelasticity. This simple model is consistent with phenomena observed during accelerated life testing such as the stochastic nature of observed failures and the decreasing likelihood of failure with increasing time after applying the pressure to reach constant stress testing conditions. Simple micromechanics models are used to demonstrate the stress redistribution mechanism from matrix to fibers under conditions of stress rupture testing, and maximum fiber stress increases are estimated for a simple unidirectional laminate. Additionally, a methodology to use strain measurements to infer fiber stress increases is developed and results from strain measurements on COPVs provided. The application of this improved understanding and modeling in a proof screening mitigation approach provides an alternative to stress rupture reliability based on burst strength testing and accelerated life testing above proof pressures. Thereby, it addresses the issues previously identified with accelerated life testing of carbon fiber COPVs. Furthermore, the proposed methodology could be applied to flight vessels at operational pressure levels, without disqualifying their use as flight components, thereby overcoming the need to establish similarity with surrogate specimens.

6.0 References

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