WHITEPAPER

Prepreg composite laminates for eVTOL primary structure: constitution, cure processing and certification evidence

A literature-based review of how prepreg constitution, cure processing, porosity and moisture govern the properties of carbon/epoxy laminates, and of the substantiation framework for VTOL-capable aircraft structure.

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Abstract

Primary structure of electric vertical-lift aircraft is increasingly specified in carbon/epoxy prepreg laminates, where weight is traded against manufacturing cost, damage tolerance and environmental durability. This paper reviews the physical relations and the published evidence that connect prepreg constitution and cure processing to laminate properties and to the substantiation required for certification. It covers ply and laminate stiffness (rule of mixtures and classical laminate theory), the effect of resin content on fibre-dominated properties, cure kinetics and its role in the design of cure cycles, the influence of porosity on interlaminar shear strength, and moisture diffusion with its consequence on glass-transition temperature and flexural properties. It then summarises the building-block approach, environmental criteria and damage categories of FAA Advisory Circular 20-107B and the structural paragraphs of the EASA special condition for small-category VTOL-capable aircraft. Worked examples use generic properties or values taken from the cited papers. The evidence differs in scale and material system, includes model results and coupon measurements, and none of the reviewed sources reports full-scale data on a lift-rotor aircraft structure; the paper states the limits of each source. Open problems concern out-of-autoclave porosity control at structural scale, long-term hot/wet behaviour under realistic service spectra, and the transfer of coupon-level knockdown factors to bonded and thick details.

1. Introduction and scope

A prepreg is a fibre bed pre-impregnated, fully or partly, with a B-staged thermosetting resin. Laminates are made by stacking plies at selected orientations and curing the stack under temperature and pressure. In the autoclave route the cure pressure is of the order of several atmospheres and consolidation is robust; in vacuum-bag-only (VBO) processing the consolidation pressure is limited to 0.1 MPa (1 atm), so entrapped air and moisture must be removed through the prepreg before the resin gels [2]. The choice between routes, the cure cycle and the resin content therefore determine void content, fibre volume fraction and degree of cure, which in turn set the stiffness, strength and durability that the structural analysis assumes.

This paper treats polymer-matrix continuous-fibre laminates made from unidirectional or woven prepreg. It does not treat thermoplastic matrices, sandwich core materials, additive manufacturing or fabrication of specific parts. Statements are attributed to the source that

reports them; values calculated here from stated assumptions are labelled as such. For a general introduction to the product category see the site's composite materials guide.

2. Theory and governing equations

2.1 Ply constitutive relations

A unidirectional ply is an orthotropic material: it has two perpendicular planes of symmetry, and four independent elastic constants suffice in plane stress, the longitudinal and transverse moduli E_1 and E_2 (Pa), the in-plane shear modulus G_{12} (Pa) and the major Poisson ratio ν_{12} (dimensionless) [7]. For thin laminates, strains and stresses with a through-thickness index are eliminated and the two-dimensional relation applies [7]:

$$\begin{Bmatrix} \sigma_1 \\ \sigma_2 \\ \tau_{12} \end{Bmatrix} = \begin{bmatrix} Q_{11} & Q_{12} & 0 \\ Q_{12} & Q_{22} & 0 \\ 0 & 0 & Q_{66} \end{bmatrix} \begin{Bmatrix} \varepsilon_1 \\ \varepsilon_2 \\ \gamma_{12} \end{Bmatrix}$$

with $Q_{11} = E_1/(1-\nu_{12}\nu_{21})$, $Q_{22} = E_2/(1-\nu_{12}\nu_{21})$, $Q_{12} = \nu_{12}E_2/(1-\nu_{12}\nu_{21})$, $Q_{66} = G_{12}$ (all in Pa) and $\nu_{21} = \nu_{12}E_2/E_1$. Here σ_i are normal stresses and τ_{12} the shear stress (Pa), ε_i and γ_{12} the corresponding strains (dimensionless). Unlike yielding in metals, failure of composites involves several interacting modes: matrix cracking, delamination, fibre fracture and fibre/matrix interface failure, which occur at different scales from the ply to the panel [7].

2.2 Fibre-dominated properties and resin content

To first order the longitudinal modulus follows the rule of mixtures,

$$E_1 \approx V_f E_f + (1 - V_f) E_m$$

where V_f is the fibre volume fraction (dimensionless) and E_f, E_m are the fibre and matrix moduli (Pa). Because $E_f \gg E_m$ for carbon/epoxy, E_1 is nearly proportional to V_f.

Worked example from cited data. Grande et al. measured unidirectional laminates of one carbon/epoxy system at constant fibre areal weight with increasing resin content [6]. At 35 % resin by weight the fibre volume fraction was 57.3 % and the measured tensile modulus 21.6 Msi (149 GPa); at 43 % resin it was 48.9 % and 18.5 Msi (128 GPa). Scaling the first value with V_f alone gives $149 \times 0.489/0.573 = 127$ GPa, within 1 % of the measured 128 GPa, which is consistent with a matrix contribution that is small and with the micromechanical predictions reported in the same source [6]. The same data show that tensile failure strain was unaffected (0.0163 and 0.0159) and that ultimate tensile strength fell from 384 ksi (2.65 GPa) to 318 ksi (2.19 GPa) because the constant areal weight produced thicker plies [6]. Compression behaved differently: ultimate compressive strength fell from 207.4 ksi (1.43 GPa) to 195.5 ksi (1.35 GPa), but failure strain rose from 0.0113 to 0.0126 [6]. The authors note that the trend of failure strain with resin content agrees qualitatively with an earlier analysis and had not been compared quantitatively with it [6].

2.3 Laminate stiffness

In classical laminate theory the in-plane, coupling and bending stiffness matrices are sums over the plies of the transformed stiffness $\bar{Q}_{ij}$ (Pa) of each ply,

$$A_{ij} = \sum_k \bar{Q}_{ij}^{(k)} (z_k - z_{k-1}), \quad B_{ij} = \tfrac{1}{2} \sum_k \bar{Q}_{ij}^{(k)} (z_k^2 - z_{k-1}^2), \quad D_{ij} = \tfrac{1}{3} \sum_k \bar{Q}_{ij}^{(k)} (z_k^3 - z_{k-1}^3)$$

where z_k is the through-thickness coordinate (m) of the upper face of ply k measured from the mid-plane. The units are N/m for A_{ij}, N for B_{ij} and N·m for D_{ij}. For a symmetric stack every B_{ij} is zero and in-plane and bending behaviour decouple. The theory does not predict interlaminar stresses at free edges, which is one reason delamination must be treated by separate analysis and test.

Worked example (generic). Take $E_1 = 140$ GPa, $E_2 = 9$ GPa, $G_{12} = 5$ GPa, $\nu_{12} = 0.3$ (assumed typical carbon/epoxy values, not taken from the cited papers). Then $\nu_{21} = 0.0193$, $Q_{11} = 140.8$ GPa, $Q_{22} = 9.05$ GPa, $Q_{12} = 2.72$ GPa, $Q_{66} = 5.0$ GPa. For a symmetric cross-ply $[0/90]_s$ laminate of four plies of 0.125 mm ($h = 0.5$ mm), $A_{11} = A_{22} = (2Q_{11} + 2Q_{22})t = 37.5$ MN/m, so the effective in-plane modulus is $A_{11}/h = 75$ GPa in both directions. The bending stiffness is not isotropic: $D_{11} = 1.30$ N·m and $D_{22} = 0.27$ N·m, because the 0° plies lie in the outer layers where they contribute most to D_{11}. This is the reason stacking sequence, and not only the ply count, enters the buckling and deflection analysis. A related experimental observation in [6]: flexural stiffness of unidirectional specimens with equal numbers of plies increased with the square of the ply thickness, because the second moment of area grows with the cube of thickness while modulus falls approximately in inverse proportion to ply thickness; the authors found this important for impact response and sublaminate stability [6].

2.4 Cure kinetics and the process window

For an epoxy prepreg the degree of cure α (dimensionless, 0 to 1) is commonly described by an n-th order Arrhenius model [3]

$$\frac{d\alpha}{dt} = A \exp\!\left(-\frac{E_\alpha}{RT}\right) (1 - \alpha)^n$$

where A is the pre-exponential factor (s^{-1} or min^{-1} according to the fitting convention of the source), E_α the apparent activation energy (J/mol), $R = 8.314$ J/(mol·K), T the absolute temperature (K) and n the reaction order. Wang et al. fitted dynamic differential scanning calorimetry data of a rapidly cured vacuum-bag resin and report $E_\alpha = 88.4$ kJ/mol, $n = 0.8923$ and a mean $A = 7.949 \times 10^{10}$ [3]. Their temperature-extrapolated characteristic temperatures (gel T_i, cure peak T_p and post-cure T_f) were 110.0, 126.8 and 147.8 °C [3]. The units of A are not stated in the extracted table, so the absolute rate is not used here. The temperature sensitivity is independent of A: with $E_\alpha = 88.4$ kJ/mol, the rate constant at 127 °C (400.15 K) is $\exp\!\left[\frac{E_\alpha}{R}\left(\frac{1}{353.15} - \frac{1}{400.15}\right)\right] = 34$ times that at 80 °C (calculated here). A cure schedule therefore separates a lower-temperature stage, in which the resin flows, the air is removed and the stack consolidates while conversion remains low, from a higher-temperature stage in which most of the conversion occurs.

2.5 Voids and interlaminar shear strength

A frequently quoted empirical relation reports that, for each percent of void content up to about 4 %, interlaminar shear strength decreases by about 7 %, independently of resin, fibre type or fibre surface treatment [2]. For a void content of 2 % this implies a reduction of about 14 % (calculated here from the cited relation). Because the consolidation pressure of VBO processing is limited to 0.1 MPa, trapped gases that remain when the resin gels may produce void contents above 1 % [2]. Aerospace structural composites are reported to require porosity below 1 % for primary and below 2 % for secondary load-bearing structure [3]; these thresholds are quoted in [3] without a regulatory citation and the applicable limit for a given part is set by its qualified material and process specification [1].

2.6 Moisture diffusion and glass-transition temperature

Moisture uptake through the thickness of a plate of thickness h (m) exposed on both faces follows, at early times, the Fickian relation [5]

$$\frac{M_t}{M_\infty} = \frac{4}{h}\sqrt{\frac{D t}{\pi}}$$

where M_t is the moisture content at time t (mass fraction), M_∞ the saturation content, D the diffusion coefficient (m²/s) and t the time (s); the approximation holds for $0 < M_t < 0.6 M_\infty$ [5]. Bone et al. fitted unidirectional carbon/epoxy flexural specimens immersed in water at 60 °C and obtained $D = 3.01 \times 10^{-13}$ m²/s with $M_\infty = 1.18$ % for the Fickian model, while for the neat resin they obtained 15.2×10^{-13} m²/s and 2.65 % [5]. They found that the Fickian model described uptake up to about 56 days and that a two-mechanism (Langmuir-type) model was needed beyond that, with $M_\infty = 3.87$ % [5].

Worked example. For $h = 2$ mm and $D = 3.01 \times 10^{-13}$ m²/s, the time to reach $M_t = 0.6 M_\infty$ is $t = \pi (0.6 h / 4)^2 / D = 9.4 \times 10^{5}$ s, about 11 days (calculated here). Because $t \propto h^2$, a 6 mm laminate takes nine times longer for the same fraction of saturation. In the same study, the unaged glass-transition temperature (dynamic mechanical analysis) was 100 °C, flexural modulus 91 GPa and flexural strength 1,180 MPa; the glass-transition temperature fell by 18 %, flexural modulus by 14 % and flexural strength by 20 % with long exposure, with strength at about 90 % of the original after 7 days and about 80 % after 56 days [5]. The fall in glass-transition temperature was non-linear, sharper between 0 and 0.5 % moisture content [5]. The test was conducted in water at 60 °C and in four-point flexure, which the authors describe as a worst-case, accelerated condition and not a service spectrum [5].

3. Design and sizing methods

3.1 Design of the cure cycle

The cure cycle is chosen from the kinetic and rheological data of the resin [3]. Wang et al. evaluated initial hold temperatures of 70, 80 and 90 °C (40 min) followed by 127 °C for 2 h in a vacuum-bag-only glass-fibre prepreg. The minimum resin viscosity was 70.3, 39.5 and 37.7 Pa·s; the resulting porosity was 2.2, 0.72 and 2.8 %; and the short-beam shear strength was 56.3, 66.1 and 54.8 MPa [3]. At 70 °C the resin was too viscous to penetrate the plies; at 90 °C viscosity increased during the hold and hindered impregnation; at 80 °C the viscosity was stable, and the gases between plies could be discharged before gelation [3]. Cure time at 127 °C was examined by dielectric monitoring: degrees of cure of 90.9, 96.1 and 98.4 % after 1, 2 and 3 h, with only small improvements in short-beam strength beyond 2 h, which gave a 14 % higher strength and 7.3 % higher glass-transition temperature than 1 h [3]. These are results for one resin system and one fabric prepreg; they show the sensitivity of porosity to the hold temperature, not a transferable set point.

3.2 Process routes and porosity control

Autoclave processing consistently yields defect-free parts but has drawbacks in capital and operating cost, part size, throughput and resource use [2]. VBO processing can match autoclave quality under favourable conditions, but adverse conditions such as poor vacuum, incomplete air evacuation or high humidity often produce porosity [2]. Prepregs with a discontinuous resin pattern (semi-pregs) leave dry channels for air evacuation. Schechter et al. produced such semi-pregs on fibre beds of different weaves, fibre types and areal weights between 200 and 670 g/m² and report porosity-free flat and curved laminates, with a slight increase in bulk factor of about 0.1 for contoured laminates [2]. The abstract of [2] states that the findings demonstrate robustness of out-of-autoclave cure; the work is limited to the laminates produced in that study, and the authors present design considerations and limitations for semi-preg fabrication [2].

3.3 Resin content and toughening

Grande et al. studied the trade-off between resin content, interlaminar resin-rich layers and damage tolerance. They report that increasing resin content decreased the fibre-dominated tensile properties, that mode II interlaminar fracture toughness increased with the thickness of the resin-rich interlayer while mode I toughness was negligibly affected, that compressive failure strains increased with resin content in undamaged, notched and impacted specimens, and that analytical tools for matrix cracking and for compression after impact (CAI) were successfully applied to materials with such layers [6]. They also relate CAI to mode II toughness through the sublaminate area created by delamination: smaller sublaminate areas give higher residual compression strengths [6].

3.4 Qualification-driven design

AC 20-107B requires that material requirements in procurement specifications be based on qualification test results for samples produced with the related process specifications,

that processing tolerances, material handling and storage limits and key characteristics be established, and that qualification data cover the properties important to the control of materials and processes [1]. Discrepancies permitted by the specification should be substantiated by analysis supported by test evidence, or by tests at coupon, element or subcomponent level [1]. Design values therefore come from a controlled material and process, not from handbook means.

4. Parameters and typical ranges

Parameter	Reported value or range	Type	Source
Consolidation pressure, vacuum-bag-only processing	limited to 0.1 MPa	Process limit	[2]
Interlaminar shear strength reduction per percent void content (up to 4 %)	about 7 %	Empirical relation quoted in source	[2]
Porosity requirement, primary / secondary load-bearing structure	below 1 % / below 2 %	Value quoted in source	[3]
Void content of a conventional vacuum-bag prepreg laminate	generally above 2 %	Statement in source	[3]
Porosity vs initial hold temperature (70 / 80 / 90 °C)	2.2 / 0.72 / 2.8 %	Measured	[3]
Short-beam shear strength vs initial hold temperature (70 / 80 / 90 °C)	56.3 / 66.1 / 54.8 MPa	Measured	[3]
Minimum resin viscosity at the hold (70 / 80 / 90 °C)	70.3 / 39.5 / 37.7 Pa·s	Measured	[3]
Degree of cure after 1 / 2 / 3 h at 127 °C	90.9 / 96.1 / 98.4 %	Measured (dielectric)	[3]
Apparent activation energy, reaction order	88.4 kJ/mol, 0.8923	Fitted (DSC)	[3]
Fibre volume fraction vs resin content (35 → 43 % by weight)	57.3 → 48.9 %	Measured	[6]
Tensile modulus, same range	149 → 128 GPa	Measured	[6]
Ultimate compressive strength and failure strain, same range	1.43 → 1.35 GPa; 0.0113 → 0.0126	Measured	[6]
Fickian diffusion coefficient, carbon/epoxy composite / neat resin	3.01 / 15.2 ×10^{-13} m^2/s	Fitted (60 °C water)	[5]
Saturation moisture content, Fickian / two-mechanism fit (composite)	1.18 / 3.87 %	Fitted	[5]
Loss of glass-transition temperature, flexural modulus, flexural strength	18 / 14 / 20 %	Measured (long immersion)	[5]
Ultimate load factor	1.5 × limit load	Requirement	[4]

5. Failure modes, testing and qualification

5.1 Building-block substantiation

AC 20-107B describes the strength of composite structure as established incrementally through analysis and tests on specimens of increasing complexity (coupons, elements, details, subcomponents), the so-called building-block approach, used to address variability, environment, structural discontinuities, damage and manufacturing defects [1]. The large number of tests that provide the statistical basis comes from the lowest levels, while structural details are validated in fewer subcomponent and component tests; these validate the ability of the analysis to predict local strains and failure modes [1]. Scatter due to material and process variability is covered through allowables or design values derived under the applicable regulation [1]. In the absence of sufficient building-block evidence, overload factors must be applied in the component test and substantiated by tests or past experience, accounting for the expected material and process variation [1].

5.2 Environment

AC 20-107B requires environmental design criteria that identify the critical exposures, including humidity and temperature, with experimental evidence that design values are attained with a high degree of confidence under the critical exposures [1]. It notes that the worst case is not the same for all details: hot/wet conditions can be critical for some failure modes while cold/dry conditions may be worse for others [1]. The maximum moisture content considered relates to what is possible in the service life and may depend on part thickness, moisture diffusion properties and realistic exposures; accelerated methods should be representative, and acceleration factors that chemically alter the material, such as temperatures that cause post-cure, should be avoided [1]. The data in [5] illustrate the point: the 10 % rise in flexural modulus observed at 60 °C was attributed to post-curing, and the authors normalised their results to the post-cure value [5]. Environment-dependent residual stresses must also be addressed [1].

5.3 Damage and defects

AC 20-107B defines damage categories by detectability and residual capability. Category 1 is damage that may go undetected by scheduled or directed inspection, including barely visible impact damage and allowable manufacturing defects such as small delamination or porosity, for which ultimate load capability must be retained for the life of the structure; Category 2 is damage reliably detectable at specified inspection intervals, for which the structure must retain loads above limit load; Category 3 is damage reliably detected within a few flights by personnel without special inspection skills [1]. The porosity and ply-hold effects of Section 3 therefore enter certification as allowable-defect limits that must be substantiated by test, not only as manufacturing quality measures [1].

5.4 Regulatory text for VTOL-capable aircraft

The EASA special condition SC-VTOL-02 (Issue 2, June 2024) sets ultimate loads equal to limit loads multiplied by a factor of safety of 1.5 unless otherwise provided

(VTOL.2230), requires the structure to support limit loads without detrimental or permanent deformation and to support ultimate loads (VTOL.2235), and requires inspections or procedures to prevent structural failures due to foreseeable causes of strength degradation (VTOL.2240) [4]. For Category Enhanced, the procedures must be capable of detecting structural damage before it could result in structural failure and provisions for in-service monitoring of parts with an important bearing on safety must be established [4]. The design-principles paragraph requires that single failures do not result in a catastrophic effect (VTOL.2250(c)) [4]. These requirements are expressed at aircraft level; their translation into allowables, inspection methods and intervals for a composite laminate follows the material and process qualification of Section 3.4.

6. Open problems and limitations

1. **Scale of the evidence.** The cure-cycle results in [3] are from one resin system and a glass-fibre prepreg; those in [2] are for semi-pregs produced in a laboratory study. Neither reports structural-scale parts of the size of a lift-rotor aircraft, so the transfer of the reported porosity levels is not established by these sources.
2. **Accelerated moisture exposure.** The moisture data in [5] come from immersion in 60 °C water of a pultruded carbon/epoxy material in four-point flexure. Realistic service exposure, thicker laminates and the interaction with thermal cycling were not examined, and the two-mechanism fit shows that a single Fickian coefficient does not describe long exposures.
3. **Property knockdowns.** The 7 % per percent relation between void content and interlaminar shear strength is quoted as an empirical trend [2]; the short-beam results of [3] (66.1 MPa at 0.72 % porosity against 56.3 MPa at 2.2 %, a difference of about 15 % over 1.5 percentage points of void) are of the same order but are not directly comparable, since they involve different hold temperatures and test types.
4. **Resin-content trade-offs.** The resin-content study [6] holds fibre areal weight constant, so ply thickness and fibre volume fraction vary together; the separate effects of the two were not isolated, and the compressive trend was not compared quantitatively with the analysis cited by the authors.
5. **Free-edge and out-of-plane effects.** Classical laminate theory does not predict interlaminar stresses at free edges, and the sources reviewed treat delamination mainly by test and by empirical relation with toughness.
6. **Regulatory quantities.** The porosity thresholds quoted in [3] are not traced there to a regulation; the allowable-defect limit of a given part follows from its own qualification [1].

7. Conclusions

- Fibre-dominated stiffness follows fibre volume fraction almost proportionally: in one data set, a fall in fibre volume fraction from 57.3 % to 48.9 % lowered the tensile modulus from 149 to 128 GPa, within 1 % of the rule-of-mixtures scaling [6].
- Laminate stiffness depends on the stacking sequence as well as the ply count; a symmetric $[0/90]_s$ laminate with typical ply constants has an effective in-plane modulus near 75 GPa but a bending stiffness ratio D_{11}/D_{22} of about 4.9 (calculated).
- With vacuum-bag-only processing, consolidation pressure is limited to 0.1 MPa, and the hold temperature before gelation controls whether trapped gases are removed; one study found 0.72 % porosity at 80 °C and 2.2–2.8 % at 70 and 90 °C [3].
- Empirically, each percent of void content up to about 4 % reduces interlaminar shear strength by about 7 % [2].
- Moisture uptake reduces glass-transition temperature, flexural modulus and flexural strength; reported reductions after long immersion were 18, 14 and 20 % in one carbon/epoxy system, with diffusion that is Fickian only at early times [5].
- Certification evidence is built from coupons to subcomponents under controlled material and process specifications, with environmental criteria and damage categories defined in AC 20-107B and structural requirements for VTOL-capable aircraft in SC-VTOL-02 [1][4].

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