MATERIALS SCIENCE

STRENGTH AND MECHANISM OF FRACTURE OF COMPOSITES
RANDOMLY REINFORCED WITH SHORT CARBON FIBRES

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A method is proposed for predicting the strength of composites randomly reinforced with short fibres. The fundamental requirements for composites and production technology for thermoplasts with assigned properties and randomly reinforced with short fibres are formulated.

Predicting the strength of unidirectional composites reinforced with long high-strength fibres is based on the concept concerning fragmentation of reinforcing fibres when tensile loads act on the composite [1-4]. The pronounced scale dependence of the strength of fibres on their length is a factor that complicates the calculation. The sites and sequence of fibre breaks are determined by the different degree of hazardousness of the defects — stress concentrators. In a unidirectional composite, the reinforcing fibres restore their load-carrying capacity at distance \( l_{\text{ref}} \) from the site of a break (Fig. 1). With an increase in stresses, the ineffective length (\( l_{\text{ref}} \)) increases. Fragmentation of the fibres into pieces will continue until the length of the fragments is reduced to the critical length \( \delta = 2l_{\text{ref, max}} \) and shear stresses \( \tau^* \) on the fibre—matrix interface exceed the shear strength \( \tau_m \). The fibre fragments that attain the critical length will “slide” in the matrix on further deformation. The critical length \( \delta \) is calculated with the Kelly—Tyson equation [5]:

\[
\delta = \frac{\sigma_0 \cdot d_f}{2\tau_m},
\]

where \( \sigma_0 \) is the maximum stress in fibre fragments attaining the critical length; \( d_f \) is the fibre diameter; \( \tau_m \) is the shear strength on the fibre—matrix interface.

Since random reinforcement of thermoplasts with short fibres (\( l_f \leq \delta \)) was traditionally used to give composites special properties (friction, antifriction, thermal conductivity, electrical conductivity, etc.), the strength problems of randomly reinforced composites were not a priority. However, thermoplasts randomly reinforced with carbon fibres (CF) have recently been competing with nonferrous metals in parts and structures in machine building which are subject to tensile, compressive, and flexural loads. In comparison to unidirectional composites, randomly reinforced thermoplasts have the decided advantage of processability, which allows processing them with highly productive methods and the possibility of recycling.

In the present article, we propose an approach to prediction of the strength of composites randomly reinforced with fibres whose length \( l_f \) is one order of magnitude greater than the diameter \( d_f \) but does not exceed critical length \( \delta \).

When such a composite is stretched, the fibres oriented in the direction of or at some angle \( \alpha \) to the acting force are taut. At \( \alpha = 90^\circ \), the fibres do not absorb the tension and are usually regarded as a disperse filler which does not strengthen the matrix [3]. Such a composite can be represented as a two-phase system consisting of strengthening oriented fibres and a matrix with an inert filler. The strength of fibres with a length of \( l_f \leq \delta \) is no lower than the strength of fibres of critical length \( \delta \)

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Fig. 1. Stress distribution along a fibre with breaks.

Fig. 2. Modulus of elasticity of modified randomly reinforced polypropylene vs. volume content of CF.

according to the scale dependence. As a consequence, their ability to absorb tension in the composite is limited by the interfacial shear strength, and fragmentation of such short fibres into even smaller parts does not take place.

The ability of reinforcing fibres to absorb a load is usually assessed from the stresses $\sigma_\alpha = \sigma_0 \cos \alpha$, where $\sigma_0$ is the stress in the fibre at $\alpha = 0^\circ$. In orientation of the fibres in the direction of the tensile force ($\alpha = 0^\circ$), the “composite strengthening effectiveness factor” ($K_{ef}$) is set equal to unity [6]. For a homogeneous statistical distribution of the fibres, coefficient $K_{ef}$ is theoretically equal to 1/6, which is also characteristic of the modulus of elasticity of the composite [6]. However, the real distribution of the fibres by orientation is usually inhomogeneous over the volume of samples and articles [3]. For this reason, the theoretical calculation of the statistical distribution by orientation of the fibres is far from real and is unsuitable for predicting the strength of a composite. The real content of oriented fibres is a function of their diameter and the elastic-strength properties, binder melt viscosity, and manufacturing process parameters for the composite. In a real randomly reinforced composite made by dual-screw extrusion, $K_{ef} = 0.2-0.3$.

The linear dependence of the modulus of elasticity on the volume content of filler fibres is characteristic of thermoplasts randomly reinforced with short glass [7, 8] and carbon [9] fibres (Fig. 2). The modulus of elasticity of the polymer matrix corresponds to a zero filler content in the graph. It is natural to assume that at 100% filling, the line is extrapolated to a value for the modulus of elasticity of CF of $E_f = 235$ GPa. However, the calculated modulus of elasticity of the composite was only 52.5 GPa. This is because the modulus of elasticity for randomly reinforced short fibers in a composite is actually only determined by those fibres which are oriented in the direction of the tensile force. As a consequence, the ratio $E_{to 100\%}/E_f$ reflects the fraction of effective filler fibres that determine the rigidity of a real composite.

Oriented fibres absorb the basic load in the first stage of deformation. While the shear stresses on the fibre—matrix interface are no greater than the shear strength, deformation of the effective fibres and matrix will be the same.

To assess the contribution of the effective fibres to strengthening of a composite, the maximum stresses that can be absorbed by them must be determined. The strength of fibres (at $l_f \leq \delta$), according to the same dependence, is no lower than the strength of fibres of critical length $\delta$. As a consequence, their ability to absorb a tensile load in a composite will be limited by the interfacial shear strength.

It follows from Eq. (1) and the similarity of triangles $ABC$ and $abc$ in (Fig. 1) that $\sigma_\delta = \delta = \sigma^* l_f = 2 \tau_m / d_f$, hence, the maximum stresses in effective short fibres

$$\sigma^* = l_f \cdot 2 \tau_m / d_f .$$

The stresses in a composite in the first stage of deformation are:

$$\sigma_c^* = \sigma^* V_f^* + \sigma_m (1 - V_f^*) = E_f \varepsilon_c (1 - V_f^*),$$

where $\sigma^*$ is the maximum stress in effective short fibres; $\sigma^*_m$ is the stress in a carbon-filled matrix; $E_f$ and $E_m$ are the modulus of elasticity of the fibre and matrix; $\varepsilon_c$ is the relative deformation of the composite; $V_f^*$ is the volume fraction of the effective fibres.

Substituting the values obtained in Eq. (3), we determine the maximum stresses in the composite in the stage of effective strengthening.