FRACTURE KINETICS OF HIGHLY RESISTANT MATERIALS BASED ON ENGINEERING THERMOPLASTICS

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Engineering thermoplastic composites are discussed. It is shown that modification and changes in the nature and type of the filler are adequately reflected in the energy parameters of acoustic emission which is associated with damages of the thermoplastic matrix, organic fillers and their interface.

INTRODUCTION

The leading industries, such as machine building, chemical and electronic ones, are in urgent need of polymer composite materials (PCM). The materials must be sufficiently strong, resist high strains, possess high impact strength, and in addition meet a number of specific requirements. Polymer materials devised with the use of engineering thermoplastics, particularly, aromatic polyesters (polyarylate, polycarbonate) reinforced with fibers are rather promising.

In recent years there was observed a great interest in the incorporation of organic fibers into PCM. Despite the abundant information on the constitution, production processes and properties of PCM at different conditions of operation (1 - 3), progress in the development and application has been restrained owing to scarce and contradictory information on their behaviour at static and dynamic loading, as well as on the mechanisms of damage accumulation in the materials, and their failure.

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Valuable information on the initiation of microcracks and the processes of damage accumulation in the materials can be obtained by measuring the energy parameters of the acoustic emission (AE) (4, 5) resulted from the collective breaking of interatomic bonds, and liberation of energy by deformation. The amount of that energy depends on the mechanism of plastic deformation or fracture. Despite the fact that AE has been widely used in studying deformation and fracture of solids, it is understood inadequately in the engineering composites based on thermoplastics. Experiments on the subject in question have been carried out mostly on carbon- and glass-filled plastics based on thermosetting resins (6 - 8), mainly, with the elastic deformation, i. e., in the range of low-frequency emission, and the results were but qualitative. Data on AE in deformed filled thermoplastics are single and rather fragmentary.

The purpose of the present work was to study the laws of AE-parameters' changing on straining (uniaxial compression) of composites based on aromatic polyesters reinforced with organic fibers.

EXPERIMENTAL

Two types of thermoplastic polymers in wide use in modern engineering materials have been used for investigations, namely, polyarylate and polycarbonate. Two types of aromatic polyamides having the elastic modulus of 18.5 and 135 GPa have been used as fiber-fillers.

Since previous experiments (9, 10) revealed that AE signals are very weak in strained thermoplastics, strict requirements are inposed on the recording instruments. We have devised a testing complex comprised of a unit for changing the load at a preset speed, and a device for measuring AE parameters that includes a AE receiving converter, a low-noise preamplifier, a high-frequency filter, and a major amplifier. The complex also incorporates a number of periphery devices that widen the capabilities thereof - a storage oscillograph for visual observations, a tape-recorder, spectrum and amplitude analyzers, a digit-printing device, and a graphing device.

The AE parameters for polyester-based materials with a chaotic distribution of organic fibers have been studied on cylinder specimens (diameter of 10 mm, height of h = 15 mm) on a testing machine ZD-4 (DDR) at strain rates of 10^{-4} s^{-1} and 10^{-3} s^{-1}.

RESULTS AND DISCUSSION

We consider some general laws characteristic of deformation of the materials in question. When experimental results on deformation and fracture of pure and fiber-reinforced polymers were compared it appeared that there was no AE in the pure polymer until the load was about 80 % of the breaking value. With increasing the load further, the rate of emission counting $\dot{N}$ was insignificant; it increased sharply, however, immediately before the specimens failed, Figure 1. Acoustic emission initiated on deformation of fiber-reinforced thermoplastics at a low strain of ε = 6 % can be explained by that the tested polymers possess high plasticity, and deformations resulted from the applied load are developing slowly thus giving but only weak acoustic signals which do not exceed the noise level. Only when the basic structural elements that provide the material's integrity are damaged there is a large increase in $\dot{N}$.

The organic fibers incorporated in the polymeric matrix differ largely from the binder in their elastic properties. Therefore, under compression, stresses in the interface of the composite increase considerably, and interfere the adhesional contact between the fiber and matrix, thus promoting microcracks that become AE sources. The higher the rigidity of the reinforcing fibers E_f / E_m, the greater the contribution, to the AE, of the sources resulted from the breaking of individual monofilaments found, under strain, in a complex stressed state when microcracks can be initiated. This is characteristic of monofilament sites where blocks of branched chains (microfibril ends) are linked (11). Within those sites, the concentration of the 'passing' molecules can be so low that the local modulus of elasticity becomes much lower than that inside the microfibrils. These point faults, when strained, give rise to microcracks. Despite the low energy, 10^{-15} J, (12) liberated by the initial act, the cumulative effect in the bulk material exceeds significantly the noise level. Since the energy liberated is proportional to E_f^2, an increase in the elastic modulus of the reinforcing fibers leads (with a similar matrix) to an earlier AE.

For materials with different types of loading diagrams σ_c (ε), variations in the speed of emission counting $\dot{N}$ are different, Figure 1. It was found out that for materials possessing distinct 'yield spots', $\dot{N}$ increases sharply at strains equal to AE occurrence. On the 'yield spots' $\dot{N}$ decreases; however, prio to the specimens' failure, it increases again. Variations in the loading rate from 10^{-4} s^{-1} to 10^{-3} s^{-1} lead to transformed dependences of $\dot{N}(\varepsilon)$. However, for all tested materials the dependences have maxima at the yield limit.

The maxima of N differ largely depending on the nature of the reinforcing fibers and loading conditions. For instance, an increased E_f / E_m and the rate of straining $\dot{\varepsilon}$ for polyarylate-based materials lead to an increased rate of AE counting.

We tried to relate the acoustic events observed experimentally to the fracture mechanism for the materials under consideration. It is known that strained fiber-reinforced polymers may have the following AE sources - fiber breaking, crack development in the matrix, fiber-matrix adhesional bond breaking, dry friction in the system in the absence of adhesional bond (4, 13). In terms of the above features, the behaviour of $\dot{N}(\dot{\varepsilon})$ curve can be explained as follows. At the end of the elastic and transient regions microcracks are being initiated in the most strained points of the fibers and in the interface. When the load approaches the yield limit, the microcracks coalesce which results in microcracks of critical dimensions in the monofilaments.

The collective character of this process causes in the specimens AE impulses of increased intensity. An assumption that on this stage the main AE mechanism is rupture of the weakest monofibers, and that only partly the damage of the interface, does not contradict the known facts (14). This is supported by experimental results in Figure 1 a, which shows variations in the AE speed counting, and compressive stresses for the specimens containing original fibers (curves 1 and 1') and modified fibers (curves 2 and 2'); there are differences in the adhesional interaction of components in the interface. Variations in the adhesional interaction within the system were the result of chemical treatments of the fiber surfaces with the objective to improve adhesion to the matrix. Curves 1' and 2' indicate that improved adhesion bonding between fibers and matrix did not change the behaviour of $\dot{N}(\dot{\varepsilon})$ prio to the 'yield spot'. At the same time, the modification of fiber surfaces affects AE behaviour on the 'yield spot', especially before fracture, which event is distinctly shown by a second extremum observed before the fracture of the specimen on the curve. Abrupt changes in the $\dot{N}(\dot{\varepsilon})$ curve directions with increasing deformation evidence to changes in the fracture mode of the material.

Analysis of the deformation features for the materials free of a distinct 'yield spot', Figure 1 b, indicates a high acoustic activity, curve 3', before the material fractured, curve 3.

The results obtained allow to establish a dependence between the AE-rate counting and certain behaviours of the deformation processes for polymer materials reinforced with high-strength organic fibers, and show that $\dot{N}$ characterizes

a mean-rate process of their structure damage.

To illustrate the possibility of establishing qualitative correlations between the material's deformation and AE parameters, we considered the effect of the reinforcing-fiber length on the emission counting rate, Figure 2. It was found out that the fiber length influences the deformation behaviour of polyarylate composites. For instance, with short fibers (l = 1 to 2 mm) the load diagram changes significantly, curve 1(the ultimate stress is increased, and 'yield spot' disappeared), and as the result, the acoustic events changed in character, curve 1'. The AE counting rate increases in the elastic deformation region. Since pure elastic deformations cannot be a source of AE, its existence in this region evidences to microdamages. AE can be explained, probably, by the fact that substantial number of short fiber tips (10 to 100 times more than in case of long fibers) are the centers of stress concentration. Fracture proceeds by the machanism of crack initiation either due to debonding along the fibers, or breakdown of the polymeric binder. Under low compressive stresses, however, damages are local, the specimen retains integrity and continues to withstand increased external loads.

Further deformation proceeds as individual acts, and microdamages are random in character, which is registered as continuous spectrum of high AE until the specimens fail. The results of the experiments carried out support the previous hypothesis (11) stating that faults which cause the short-fiber breakdown differ largely from the faults which lead to long-fiber breakdown. On increasing the fiber length (l = 50 mm), the load diagram is changed in appearance, Figure 2, curve 2, and this is adequately shown in the AE counting rate, curve 2'.

The phenomenon described concerns an important problem - the role of fiber length in the strength of materials. Brookfield et al. (15), for instance, determined critical fiber lengths. They showed the fracture to occur with the fiber lengths unable to transmit further load increments. We take similar critical fiber lengths for the materials in question. From this viewpoint we learn how the fiber lengths affect the parameters of the amplitude distribution. The analysis of the data obtained shows the mean rate of the damage processes to be lower for shorter fibers. The energy liberated during fracture, and the total area of the accumulated microdamages are smaller in this case, hence the material is more stable in the initial loading period, when it is filled with shorter fibers.

The modification of fiber surfaces aimed at improving the interaction with polymer matrix, changes the kinetics of microdamages accumulation in the material on increasing the

deformation rate, which is also shown in the parameters of AE amplitude distribution. It was observed, particularly, that improved adhesional interaction activated the AE sources resulted from interface fracture at relative strains twice as large as in the case of unmodified fibers. The following factors become increased: emission energy W (from 8×10^3 to 10^5); total emission N_Σ (from 0.25×10^3 to 9.8×10^3), and average emission amplitude $\bar{A}$ (from 5.2 to 9.6).This can be explained not only by increased energy on the interface and the rate of microdefects growth under the conditions of total fracture (13), but by the preferential processes of numerous fiber breakdown, or even block breakdown. The dispersion of AE amplitudes evidences, particularly, to a nonuniform development of these processes.

CONCLUSION

AE technique allows to reveal particularities of kinetics of damage accumulation in the engineering thermoplastics filled with high-strength organic fibers, including modified fibers, and to explain differences in the strength and deformation characteristics. The dependences of $\dot{N} = f(\varepsilon)$ derived for deformation rates of $\dot{\varepsilon} = 10^{-4}$ to 10^{-3} s^{-1} allow to determine the safe deformation ranges for the above materials, which information is required when the operation regimes are being set.

SYMBOLS USED

$\dot{N}$ = rate of AE counting (impulse/sec)

σ_c = stress (MPa)

ε = strain (%)

E_f = elastic modulus of fiber (GPa)

E_m = elastic modulus of matrix (GPa)

W = emission energy (relative units)

N_Σ = total emission (relative units)

$\bar{A}$ = mean amplitude of emission (relative units)

l = fiber length (mm)

$\dot{\varepsilon}$ = rate of deformation (s^{-1})

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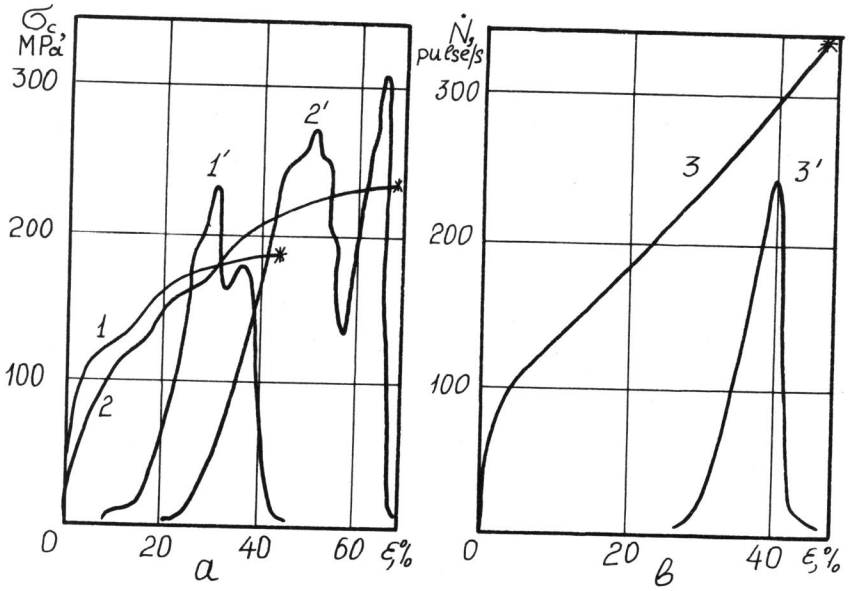

Figure 1 Rate of AE counting $\dot{N}$ and σ_c versus ε

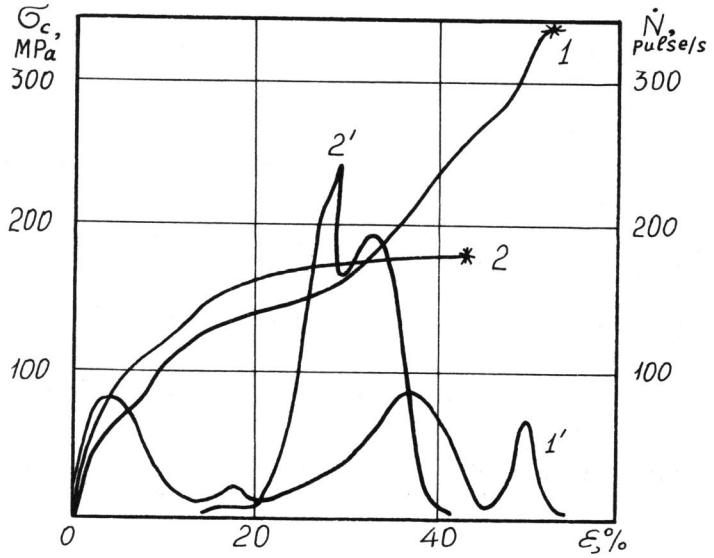

Figure 2 Rate of AE counting $\dot{N}$ and σ_c versus ε and 1