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STRENGTH AND BEHAVIOUR OF COMPOSITE MATERIALS IN FAILURE*

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1. Static strength

In connection with the growing need of prediction and prolongation of the life of structures the strength and behaviour of material in failure is coming to the foreground of interest of the mechanics of materials, particularly in extreme environmental conditions. So far considerable progress has been made in the calculation of strength and the description of the deformation behaviour up to failure of relatively simple, homogeneous or quasi-homogeneous systems or some simple heterogeneous systems, e.g. polymer matrix with parallel fibres oriented in one direction.

Theoretical prediction of fracture strength of random-oriented multiphase systems or systems with a complex geometrical configuration of components, however, is far more difficult and has not been fully elucidated so far. Various theories, the number of which, obviously thanks to the very complexity and obscurity of the problem is very large, do not form the subject of this paper. An exhaustive summary of knowledge in the field of failure and strength of materials (at least until the beginning of the seventies) is given in the seven-part work by Liebowitz [1], one two-volume part of which is concerned with composites. The problems of failure of plastics and composites with a plastics phase are dealt with in detail by Nemec *et al.* [2]. An excellent brief review of approaches to strength assessment is given also by Javornicky [3], and some common problems of failure in composites are mentioned by Pluhar and Zilvar [4].

More important for understanding the properties of the material is to know the *deformation mechanism, mechanism of the failure process and total failure* (fracture) rather than the absolute strength value which cannot be determined reliably anyway even in the case of homogeneous materials until a *general energetic theory* based on statistical characteristics of the components of the material has been derived.

Principal difficulties in this field are connected:

— with the complexity of stress distribution in a non-homogeneous body (stress concentration, influence of contraction under load),

— with the effects of the "generic" stress and the dependence of the stress state on the whole history of the material and the technology of its preparation,

— with the uncertainty connected with the criteria of failure, effects of temperature, moisture, ambient environment, fatigue, defects and initial microdefects, etc.

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The basic concept of the *theory of damage* (Janson, Hult [7], Katsanov [8], Rabotnov [9]) is oriented on the characterization of the occurrence of defects in the *whole* volume of the continuum with the assumption that it is only their cumulation in a certain extent that results in failure; the resulting strength is the function of the number of defects of the system.

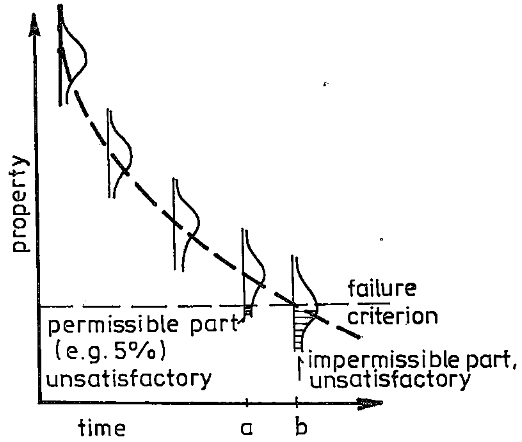

Fig. 1. Attainment of failure criterion on the basis of statistical property distribution and predetermined quantile; a — failure time with the consideration of statistical property distribution, b — failure time with the consideration of the mean property

Fracture mechanics, on the other hand, is concerned with the mechanics of the crack alone (Liebovitz [1], Knott [10]); it is based on Griffith's theory of a brittle crack [11] according to which the crack will not attain stability at the given stress if the reduction of elastic energy originating from the stress relief due to crack widening is just in equilibrium with the increase of energy required for the formation of a new surface (Inglis [12], Orovan [13]). An important consequence of Griffith's theory is the fact that fracture strength is a statistical quantity and depends on the probability that a defect of sufficient magnitude has been already present in the material. The great dispersion of strength observed, e.g. in ceramic or polymer materials, consequently, is the initial characteristic of the material and is not due, primarily, to the variation of experimental results.

The character of failure of composites depends to a considerable extent on the character of the matrix, defined by the specific total fracture energy, given according to Richardson [14] in Table 1.

A qualitative contribution to fracture mechanics is represented by the theories of the *critical crack opening displacement* and particularly by the attempts at *introducing process dynamics* into the failure theory, i.e. the effect of elastic stress and strain waves (incl. the initial stresses), crack propagation barriers and changes of internal energy of the system.

Considerable influence on strength is exercized by the *pores*, almost always present in the composite, which not only reduce the effective cross-section transferring the load and act as stress concentrators resulting in the reduction of tensile strength on the one hand, but also leading to an increase of toughness and fatigue strength by the dispersion of the defects and the braking of direct crack propagation on the other hand.

In the case of materials consisting of several phases, such as cement concrete, the mechanism of failure and the failure itself are complicated not only by the complexity of the structures, the presence of hydration pores, shrinkage pores and microcracks, the presence of thermodynamically different phases (solid, liquid and gaseous), but also by the time-variable structure of the basic matrix in which hydration proceeds practically without any time limitation. As it is illustrated in Figs. 2a, 2b and 2c, cracks in concrete may originate or be initiated either on the boundary between filler and cement stone, or in the cement stone or the mortar matrix, or in the filler particles, according to the relative strength in these three places.

When concrete is loaded, microdefects (microcracks) originate which cause irreversible changes of the internal structure of concrete; energy is dissipated in the form of heat, mechanical vibrations and formation of new surfaces. Visible failure on the surface originates only when the stresses attain 80–90% of strength; microcracks at lower stresses can be observed by various indirect methods (sound emission, ultrasonic pulse velocity) or by accurate measurements of deformations in various directions.

To explain and describe the mechanism of the failure of concrete a number of structural models has been devised, such as the model consisting of random-oriented elements failing in shear in one direction (BRANDTZAEG [18]) according to Fig. 3, or the model consisting of spheres in a cubical arrangement connected by thin bars (vertical, horizontal and diagonal elements) according to Fig. 4 (REINIUS [19]) or the models consisting of four identical particles in a mortar matrix in a plane (Fig. 5, BAKER [20]), or of six particles in space (Fig. 6, ANSON [21]) resulting on the origin of transverse tensile stresses under compressive load, as well as the Navratil's model (Fig. 7) which, thanks to the consideration of the inner surface, can characterize best the behaviour of concrete in failure also with regard to ambient environment (BAREŠ [22]).

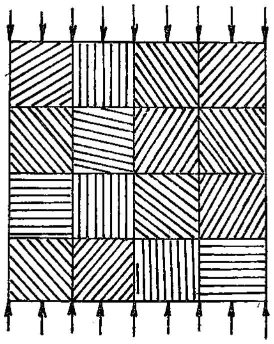

Fig. 3. Brandtzaeg's structural model of concrete [18]

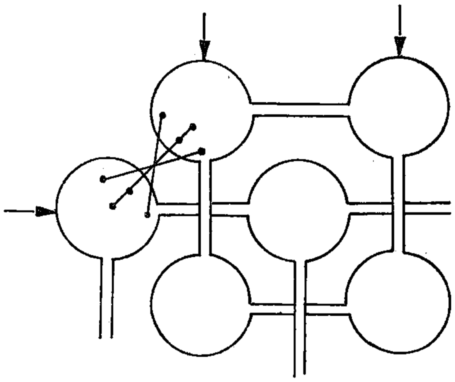

Fig. 4. Reinius' structural model of concrete [19]

All theories of crack development and concrete failure are based, once again, on Griffith's theory and suffer from the same shortcomings: assumption of elasticity and homogeneous continuity, constant modulus of elasticity and constant specific surface energy.

considerable importance is played by the spectrum of the stress relaxation times of the polymer matrix (HUGO [23, 24]).

The first "deterioration" process of the matrix (dissipating the supplied energy) begins at a very low value of tensile deformation (about 0.1%), but may result also in strengthening. With further stress increase the elastic energy in microvolume continues to increase which results in the end in a sudden origin of crazes statistically distributed in the stressed volume. Crazes are precursors of fracture and their concentration and stability (governed by the geometry of the complex structure and the spectrum of relaxation times of the matrix) determine the useful strength of the composite (HUGO [25, 26]).

In the process of failure crazing precedes crack widening. The failure, therefore, proceeds through crazes and not through compact material. The origin of crazes enables the dissipation of energy (while the strength of the body simultaneously decreases) and depends on molecular weight, temperature, frequency and environment. The region of the craze grows continuously with increasing load (or the number of cycles), but at a decreasing speed, as the length of the individual the craze increases. When certain critical conditions have been attained, the crack suddenly widens across the whole craze and stops at its tip (this discontinued cracking is accompanied by audible sounds — Fig. 8). Crazes grow during the first 10% of their life to about 80% of their final length; for some 80% of their life their increments are negligible; in the last 10% of their life a speedy increase follows again. In the middle part of the craze life there is a metastable equilibrium between the orientation of the fibrils (strain hardening) and the growth of pores (strain softening). In the first part unbalanced conditions originate when the softening due to the formation of craze pores prevails over hardening by orientation. This mechanism of discontinued growth of crazes and cracks results in the formation of visible zones on the surface of the material (perpendicular to the direction of crack development) differing in colour and structure and called the "discontinued growth zones".

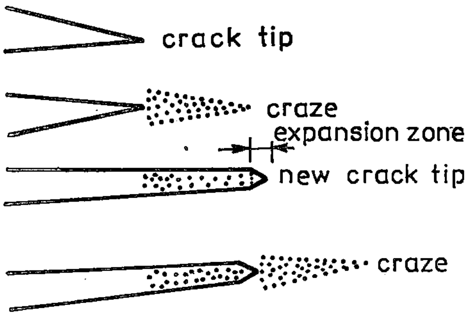

Fig. 8. Process of discontinued development of crack

Filler prolongs the time of stress relaxation of the matrix, increases the velocity of growth of elastic energy in microvolume and in the end results in the preference of plastic deformation in matrix microvolume (in a worse case in the preference of microcracks) to the development of retarded elastic deformation. A typical consequence of this effect

In the case of the ABS[1], however, this mechanism was not ascertained and plastic flow of the system was observed instead (in which also brittle molecules participated), resulting in the orientation of the matrix. So far no successful explanation of this phenomenon has been found.

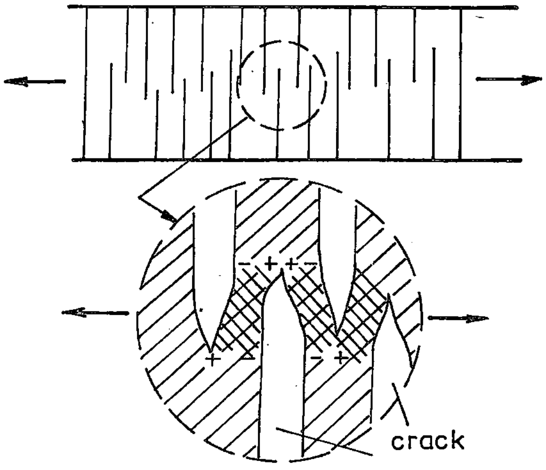

Fig. 9. Schematic representation of cracks originating during tensile deformation in a high impact strength polystyrene (HIPS) (+ tension, · compression) [29]

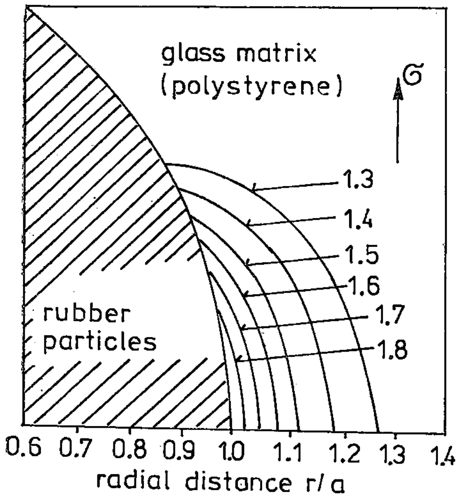

Fig. 10. Stress concentration in the HIPS under uniaxial tension, calculated by means of Goodier's equation in the proximity of the equator of an isolated rubber particle and in polystyrene

Another theory explaining the improved behaviour of rubber-modified polymers (NIELSEN [28]) assumes that in the course of mechanical strain the major part of dissipation heat concentrates around the particles with stress concentration and thus causes a local temperature increase above the glass-transition temperature, during which the matrix is capable of major deformations without failure. New investigations of polyethylene, however, have shown that the heating (which does take place around the particles) is the consequence of flow and not its cause.

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[1] Acrylonitryl–Butadiene–Styrene.

optimum compatibility with the glass-polymer to ensure the necessary adhesion between the phases.

The toughness of filled thermoplastics is based on two mechanisms (HUGO [26], HUGO, SOVA [35]): on the conforming elasticity of the matrix and on the consumption of energy for the plastic deformation of microvolume and for the origin of crazes or microcracks uncapable of further development. The first takes place at elevated temperatures and its effect decreases with the decreasing volume quota of the matrix. The second proceeds at lower temperatures (particularly fairly below the Tg) and its effect increases with the rising quota of the reinforcing filler or with the growing aspect ratio l/d. The equivalence of time and temperature is applicable here to a considerable extent.

The opinion that higher toughness is ensured by the mechanism of crazes in the proximity of dispersed pliable particles has been extended recently also to filled thermoplastics. The effect of crazes required to increase toughness is ensured by that the filler particles of micron dimensions are coated with a suitable organic pliable layer. The problems consist in the control of the degree of dispersion. The optimum dimensions of the particles of the dispersed phase and the thickness of the pliable coat which would ensure maximum increase of toughness while minimizing the drop of strength and other necessary properties (e.g. modulus of elasticity) remains an open question.

3. FATIGUE STRENGTH

Some tough plastics (such as the rubber-modified thermoplastic polymers) are, in spite of their excellent behaviour under static load, sensitive to repeated loads, e.g. pressure pulses in plastic pipes may produce premature brittle failure under a surprizingly low load. Similarly with the static load there is a certain combination of craze and shear response also under repeated load. Nevertheless, tests have shown that the craze mechanism is often dominant for the development of fatigue cracks. However, if the effect of hysteretic heating is decisive for fatigue failure or if the matrix is partly tough (e.g. PVC), the shear response may play an important role. In spite of that it seems that the crazing plays a role of decisive importance also under cyclic load *for bodies without notches*, as it is shown in Fig. 12 (BUCKNALL [36]). It is possible to observe changes progressively increasing in subsequent cycles with stresses sufficiently high to induce flow and whitening: the polymer shows greater softening and a larger hysteresis area. The secondary, lower yield limit appears, the value of which decreases with further cycles. The crazing is connected with the upper yield limit, the flow with the lower (KAMBOUR [34]). The relatively considerable recovery which takes place after load relief is due to the "healing" of the crazes which, consequently, incorporate not only plastic deformations, but also significant reversible viscoelastic deformations.

During cyclic loading temperature increases, the more so the greater the amplitude of applied stresses. Figure 13 shows that in tension the cyclic loading over the same period of time is far more unfavourable that static load. At higher stresses tough failure takes place in the form of the development of internal cracks, at lower stresses brittle failure takes place originating on the surface.

All above mentioned and other experimental results point out how important is to take into account the deformation softening and hysteretic heating of bodies which are without notches or contain only very small defects. The consequence is that an addition of rubber to polymers *reduces the fatigue life* of modified systems, so that their higher toughness can be utilized under repeated or alternating loading only for a shorter time.

In the case of notched bodies the behaviour is somewhat different, although their toughness under static and dynamic loads is also increased by an addition of rubber phase. Hysteretic heating takes place only at the crack tip, which can result in localized flow and an increase of the crack tip radius, which is advantageous. The notched bodies are sensitive to loading frequency in the same way as the polymers not modified by rubber addition.

If we sum up briefly the knowledge about the fatigue of these systems, we can state that the fatigue response of tough (modified) polymers is influenced mostly by the higher level of microelastic damping in comparison with the nonmodified matrix. When the volume of the material exposed to cyclic loading is big such as in the case of notchless bodies, particularly at high frequencies and loads, the failure is most probably due to the considerable softening produced by hysteretic heating (Fig. 14). When the failing surface is small, such as in the case of a body with notches, local heating may be advantageous for some materials when it blunts the crack. For other materials even in a notched body, however, the zone of failure may be so large that the heating is considerable and the softening of the matrix promotes the crack development.

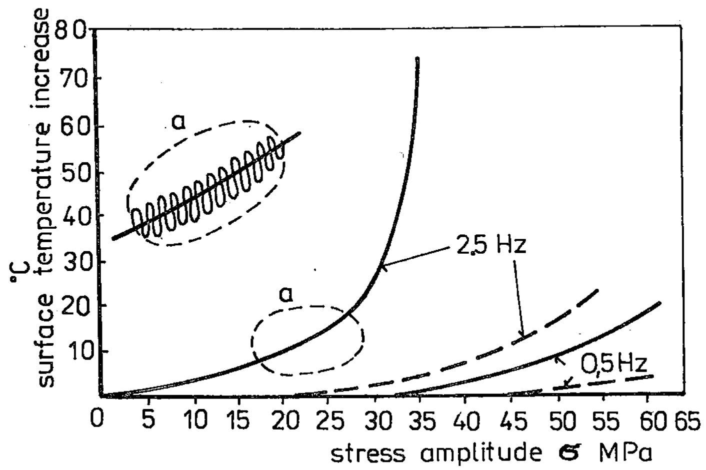


Fig. 14. Increase of temperature of the PMMA under repeated loading in bending [38]. Actually the temperature does not rise continuously, but oscillates about the mean value due to the thermoelastic effect; ——— without cooling, - - - - - - - - - - with cooling

Very little has been done in the field of fatigue of filled composites with a brittle (nonmodified) matrix, and the majority of published results was concerned with dental or bone cements. In the majority of cases neither their tensile strength nor their toughness, attains the properties of the polymer alone, or exceed them only slightly (BEAUMONT, YOUNG [39]). The toughness and the resistance of such materials to fatigue depend greatly that ionomeric bonds at the interface should be advantageous is entirely in agreement with the contemporary ideas about the acid-base interaction between composite components (FOWKES, MOSTAFA [44], MARMO *et al.* [45]). In any case it is necessary to give preference to all studies of interface properties which are of fundamental and practical importance for a successful creation of tough and strong composites of the 1st type.

If the high-modulus dispersed phase makes a transition from the spherical shape to the oblong, fibrillar shape, a marked increase of toughness and strength of the composite takes place, due to the filler, without changing the classification of the composite as the composite of the 1st type. The biggest effect in the direction of load is achieved by the inclusion of oriented continuous fibres, in other directions by the inclusion of random-oriented short fibres or the combination of layers of various orientations into the laminate.

The static and fatigue behaviour of fibre-reinforced composites have been afforded much attention, although the research concerned was mostly of experimental and empiric character. The reasons for the shortage of fundamental integral studies of fibre-reinforced composites are probably due to the following complications:

— the criteria of failure are very different; even if failure is progressive, physical integrity may be preserved for many loading cycles (CHAMIS [46]);

— the failure of fibres, matrix and interface is accompanied by numerous independent processes (see Table 2); the processes taking place in the polymer are influenced by the change of state due to hysteretic heating and time-dependent behaviour;

Table 2

Factors influencing fracture work, [47]

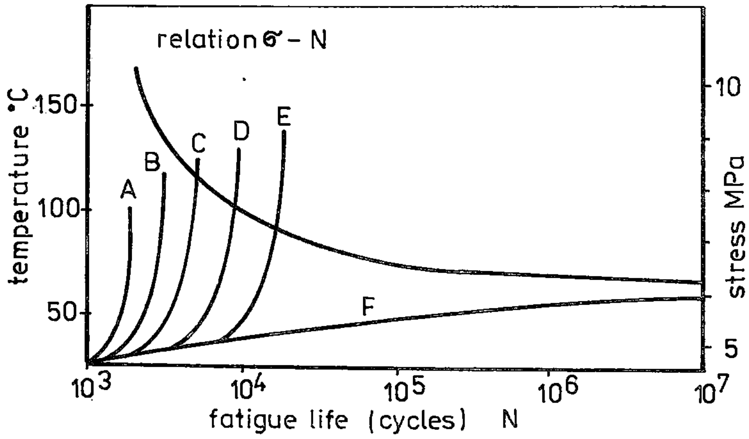
Type of work	Origin	Kind of energy dissipation
Internal work of fibres	brittle fracture of fibres	stored elastic energy
	brittle flexure during pull-out	plastic flow and narrowing
	tough fracture of fibres	plastic flow and narrowing
Work of interface	difference in tensile strain across the interface	friction slippage of plastic shear in the matrix
	pull-out of fibres	ditto
Work of interface and matrix	cracking (cleavage) of the matrix parallel with fibres	surface energy of the matrix and bonding energy of fibre — matrix contact
Internal work of matrix	matrix failure	surface energy of the matrix and plastic flow

— the character of failure, therefore, is very complex (REIFSNIDER [48]) and depends on the type of load, the relative orientation of stresses and reinforcement axis, overlapping of stress fields, constraint and stress concentration of the plane of the individual layers as well as perpendicularly to it, presence and character of defects;

— the concept of cracks must be redefined and in some cases the conventional approach of fracture mechanics must be subjected to revision, [48];

the load applied at an angle 45° with regard to fibre direction the interface failure and the brittle fracture of the matrix always take place.

At higher frequencies the hysteretic heating becomes an important or even decisive factor of fatigue life (Fig. 17). The origin of elevated temperature during cyclic loading may have two effects: reduction of the Young's modulus and the accompanying reduction of the *effective* length of the fibres and, consequently, an increase of the *critical aspect ratio l/d* (CESSUA *et al.* [53], DALLY, BROUTMAN [54]). In some cases the fibres may assist in the reduction of hysteretic heating (e.g. if their thermal conductivity is high).



Sample	Stress MPa	No. of cycles, N	Temperature, °C
A	10.3	2×10^3	100
B	9.0	4×10^3	115
C	8.3	6.1×10^3	125
D	7.6	9.5×10^3	130
E	6.9	1.9×10^4	141
F	6.3	10	60

Fig. 17. Relation between fatigue life (up to fracture) and stress of a highly crystalline polymer (polytetrafluorethylene) at the frequency of 30 Hz and temperature growth during cyclic loading [52]

In all tested cases the failure began with the failure of the interface; however, further crack development differed very considerably (according to matrix type). For example, cracks propagated very speedily in polystyrene, but not in polyamide or polyethylene, (DALLY, CARRILLO [55]). The reason for the primary failure at the interface is that there is always some internal stress in the composite, e.g. due to its cooling from the hardening temperature, which induces tensile stresses at the contact surface. The application of bonding agents has always improved the fatigue life (even if the optimum value of interface strength is not known) [53].

The tests of a polyester laminate reinforced with a mat of cut glass strands (OWEN, DUKES [56]) have revealed the sequence of events during failure: loss of adhesion, cracking to expand radially around the fibres similarly with the origin of crazes. Further the pulling-out of the fibres from the matrix acquires importance.

Reinforcement with short fibres results almost always in a marked improvement of fatigue strength; the reinforcement with carbon fibres, which have at least three times higher rigidity and ten times higher thermal conductivity than glass fibres, is considerably more effective. One of the principal reasons obviously consists in the reduction of the hysteretic heating at the given frequency and load (Fig. 19). Carbon fibres are also suitable for prosthetic materials, as they are compatible with living matter. The fracture mechanism of various systems reinforced with short fibres and on various load levels differs considerably; therefore, any generalization of the material fatigue response in composites must be regarded with great caution and great attention must be afforded not only to the differences of components, but also to the differences of defects induced into the composite in the course of production (even if the materials are apparently similar). For example, the injection-moulded products (with preferential orientation) show non-linear and tough behaviour; the toughness increases under repeated load by hysteretic heating. The pressure-moulded products have a linear working diagram and fail by brittle fracture, by the development of long fatigue cracks. Specimens with short life fail by the development of the fatigue crack radially around fibres and in a relatively brittle manner. Specimens with long life show a more tough failure and a considerable adhesion between the fibres and the matrix.

Some studies (e.g. DI BENDETTO, SALEE [59], BRETI *et al.* [60]) of the influence of various fibres on the development of fatigue cracks made with polyamide reinforced with 20% by volume of random-oriented carbon, polyaramide or mixed (1:1) fibres have shown that the resistance to the development of fatigue cracks of the polyaramide system is considerably higher than that of the carbon system, is approximately the same in the case of the hybrid system (with the same concentration of both fibres and, consequently, with approximately the same modulus as that of the polyaramide system), which makes the hybrid system superior to the carbon system, and that the resistance to the development of fatigue cracks of the hybrid system is higher, when it is saturated with water (8.3%) than when it has a balanced moisture content (2.8%), and that the sensitivity of the velocity of fatigue crack development to mean stress increases with the growing Young's modulus. The mechanism of failure is similar to that described above: first considerable flow of the matrix and of the adhesion layer was observed, particularly in the case of carbon fibres. In the polyaramide system the fatigue failure manifested itself first by the loss of adhesion of the fibres perpendicular to the stress direction. The loss of adhesion then proceeded in the fibres placed at a continuously decreasing angle to the load axis until cracks originated in the fibres themselves. In the case of carbon fibres, moreover, the resin expanded in load direction because of the high value of interphase adhesion.

It has been already mentioned that the best improvement of composite properties can be attained by the use of continuous fibres; apart from the application of continuous fibres alone, however, their combination with short fibres may be of particular advantage in some cases. This combination results primarily in an increase of the modulus and strength of the matrix, particularly if whiskers are used. This is advantageous in the glass fibre composites, while in the composites reinforced with boron fibres (with a high modulus) the increase of the modulus of the matrix has only little effect (increase of transverse the specific differences between a carbon composite and a metal, particularly the sensitivity of the composite to low-cycle fatigue in compression (SCHUTZ, GERHARZ [62]).

Both in the case of glass and carbon fibre-reinforced laminates local failure at low stresses takes place at the very beginning of fatigue life in the first half of the loading cycle (BROUTMAN, SAHN [63], FUWA *et al.* [64], MCGARRY [65], NEVADUNSKY *et al.* [66]). This failure may, but need not reflect in a significant change of composite behaviour in accordance with its extent and the type of system, and its importance can be assessed according to the extent to which its mechanical properties change. As the complete understanding of the fatigue process necessitates the description (characteristic) of the failure in the course of the whole period of fatigue life and the finding of the relation between failure and critical properties, a number of methods has been elaborated for the assessment of composite failure under cyclic loading [66]. From their number various ultrasonic methods as well as the acoustic emission methods which are sensitive to major pores, cracks and regions of adhesion loss, originating in the case of fibre failure or during other sudden events taking place in the material, seem to be most effective. As it is shown e.g. in Fig. 21, in a well produced composite an emission takes place during the first cycle; subsequently, the number of emissions is considerably lower, until the catastrophic failure begins (which proves that the degradation of the matrix, fibres and interface plays principal role in the course of fatigue life, [64]). However, whatever method is used, considerable caution is required in the interpretation of the significance of the failure: the question is not merely whether failure has or has not taken place, but how singnificant the failure is with regard to the long-term behaviour of the composite.

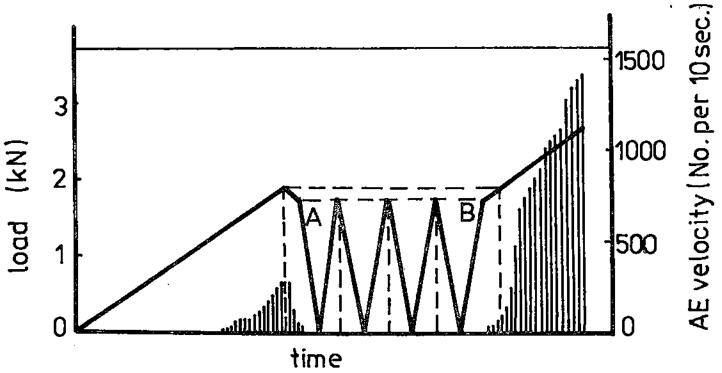


Fig. 21. Indication of local microcracking by acoustic emission under repeated loading of a well produced fibre-reinforced composite [64]

Detailed study of axial fatigue failure (fracture microstructure) of unidirectional glass composites (E-glass-polyester) has shown that the failure takes places in three stages characterized also by the change of the modulus and the hysteresis energy (KIM, EBERT [67]). In the first stage the modulus drops considerably (by as much as 60%) until the second (longest) stable stage has been attained; the drop is due to the development of surface defects on the fibres (following the failure in the first half of the first loading cycle). The third stage corresponds with the final phase of fatigue life, in which the modulus drops tion of damage in the course of cyclic loading reduces the strength until the residual strength and applied stresses are equal (Fig. 22). Fatigue strength, however, is heavily influenced also by environmental effects: temperature may influence the viscoelastic states, fracture toughness, chemical stability, etc.; generally speaking, the higher the ambient temperature, the greater the damage. Fatigue life is also indirectly proportionate with the initial absorption of water in the matrix resin. Similarly with filled polymers in bulk (see Fig. 14) also fibre-reinforced composites are influenced by the frequency of loading. Low frequency cycling gives sharper, less dispersed damage zones that higher frequencies; on the other hand, high frequencies are unfavourable, if the hysteretic heating (resulting from high internal friction in the resin and the low thermal conductivity of the composite) attains a higher value and overcomes the favourable effect, manifesting itself by the blunting of the defect (Fig. 23). Because of heat generation during load cycling the asymptotic approach of the σ-N curve to a certain horizontal value does not usually take place and there is no fatigue limit (Fig. 24).

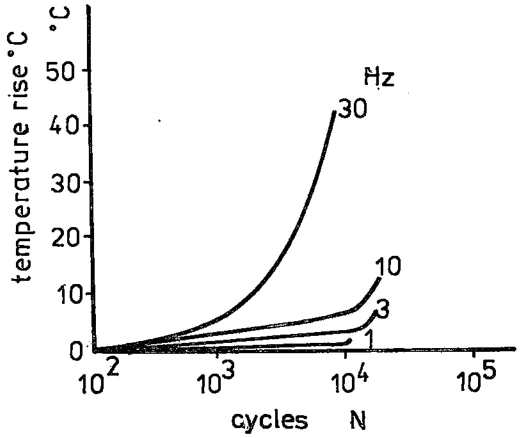


Fig. 23. Temperature increase of an epoxy-carbon fibres laminate with perpendicularly oriented layers and a central opening under cyclic loading equal 66% of the mean static strength. Speedy temperature increase is connected with failure [72]

Stresses on the interface are considerable; therefore cracks originate frequently on the interface which propagate across the remainder of the composite until complete failure takes place (BROUTMAN [75]) under repeated loading after crack origin on the interface vacancies less than 0.05 μm thick originate without reducing significantly instantaneous strength (BROUTMAN [76]).

On the other hand, the possibility of creating a tough system out of two brittle materials is the consequence of the very presence of the interface between these two materials of different strengths and moduli. Similarly as the presence of weak cleavage planes is important for the toughness of a homogeneous material (COOK, GORDON [77]) it is also important that a certain plane (or region) of weaker properties (or potential cleavage), perpendicular to the plane of crack development, be present in the composite. The failure on the interface causes the development of a secondary crack which prevents the development of the primary

4. CONCLUSIONS

The paper pointed out the extraordinarily complex character of the problems connected with the prediction of the static and fatigue strengths and toughness of composites.

Every research in this field, if it should be successful, must accept and respect this complexity, must include it into initial considerations and interprete its results with extreme caution.

Theoretical as well as experimental work should proceed particularly towards

— the conception of a non-stationary energetic theory of strength based on statistical characteristics of components, structure and statistical description of deformation and failure mechanisms, e.g. by an adequate syntheses of damage theory and fracture mechanics,

— the elucidation of the criteria and mechanisms determining toughness (and fatigue strength) and the conception of a general theory of the control of practical toughness of structural systems using the viscoelastic, elastoplastic and craze effects,

— the finding of criteria and mechanisms of fatigue strength in connection with the mechanisms determining static strength and toughness,

— the elucidation of the mechanisms of deformation and failure of the interface and the finding of the methods of control and forecasting of its properties,

— the assessment of fibre-reinforced composites and their systems (including their joints) in real application conditions.

REFERENCES

1. H. LIEBOWITZ (ed.), *Fracture*, Vols. I–VII, Academic Press, 1968–1972
2. J. NEMEC, S. V. SERENSEN *et al.*, *Pevnost plastickych hmot*, Statni nakladatelstvi technicke literatury (SNTL), Praha 1970
3. J. JAVORNICKY, *Kompozitni materialy*, in: *Proc. Mathematical Analysis of Mechanical Behaviour of Composite Materials*, Seminar of Czechoslovak Society of Mechanics, Czech. Acad. Sci. (CSAS), Prague 1981, p. 26
4. J. PLUHAR, V. ZILVAR, *Nektere spoleene problemy porusovani v kompozitnich materialech*, Kompozitni materialy, CSSM-CSAS, Praha 1981, p. 43
5. R. A. BAREŠ, *Kompozitni materialy*, SNTL, Praha 1988
6. W. WEINBULL, Ing. Vetenskaps Akad. Handl., **151**, 1939
7. J. JANSON, J. HULT, *Fracture mechanics and damage mechanics: A combined approach*, J. de Mech. Appliquée, **1**, 1977, 69–84
8. Л. М. КАЦАНОВ, *Основы механики разрушения*, Изд. ,,Наука'' Москва 1974
9. Ю. М. РАБОТНОВ, *Влияние концентрации напряжений на длительную прочность*, Мех. тверд. тела, 2, 1967, 36—41
10. J. F. KNOTT, *Fundamentals of Fracture Mechanics*, Butterworths, London 1973
11. A. A. GRIFFITH, *The phenomena of rupture and flow in solids*, Phil. Trans. Roy. Soc. of London, A **221**, 1921, 163–197
12. C. A. INGLIS, *Stresses in a plate due to the presence of cracks and sharp corners*, Proc. Inst. Naval Architects, **55**, 1913, 219–241
13. E. OROVAN, Phys. Soc. Rep. Prog. Phys., **12**, 1948, 186–197
14. M. O. W. RICHARDSON (ed.), *Polymer Engineering Composites*, Appl. Sci. Publishers LT Ltd., London 1977

51. R. E. Lavengood, L. E. Guldbransen, Polymer Eng. Sci., **9**, 1969, 365–369
52. M. N. Riddel, G. P. Kov, J. L. O'Toole, Polymer Eng. Sci., **6**, 1966, 363
53. L. C. Cessua, J. A. Levens, J. B. Thompson, Polymer Eng. Sci., **9**, 1969, 339–345
54. J. W. Dally, L. J. Broutman, J. Comp. Mat., **1**, 1967, 424–430
55. J. W. Dally, D. H. Carrillo, Polym. Eng. Sci., **9**, 1969, 434–437
56. M. J. Owen, R. Dukes, J. Strain Anal., **2**, 1967, 272
57. M. J. Owen, T. R. Smith, R. Dukes, Plast. Polym., **37**, 1969, 227–232
58. J. Theberge, B. Arklesm, R. Robinson, Ind. Eng. Chem. Prod. Res. Dev., **15**, 1976, 100
59. A. T. Di Bendetto, G. Salee, Tech. Rep. AMMRC CTR, 77-12, 1977
60. P. Breti, R. W. Hertiberg, J. A. Manson, ACS Symposium Series, 1979
61. B. Harris, Composites, **8**, 1977, 214
62. D. Schutz, J. J. Gerharz, Composites, **8**, 1977, 245
63. L. J. Broutman, S. Sahn, Ann. Tech. Conf. SPI, 24th Paper 11 D, 1969
64. M. Fuwa, B. Harris, A. R. Bunsell, J. Phys. D. Appl.-Pys., **8**, 1975, 1460
65. F. J. McGarry, in: *Fundamental Aspects of Fibre Reinforced Plastics Composites* (ed. R.T. and H. S. Schwarto), Wiley/Interscience, New York 1968, 63
66. J. J. Nevadunsky, J. J. Lucas, M. J. Salkind, J. Composite Mat., **9**, 1975, 394
67. H. C. Kim, L. J. Ebert, J. Composite Mat., **10**, 1978, 139
68. C. K. H. Dharan, in: *Failure Nodes in Composites* II (ed. J. N. Fleck, R. L. Mehan), The Metallurgical Soc. of AIME, New York 1974, 144
69. J. C. Halpin, K. L. Jerina, T. A. Johnson, ASTM STP I 21, 1973, 5
70. J. W. Hamer, R. T. Woodhams, Polymer Eng. Sci., **21**, 1981, 603
71. S. M. Pudd, J., Frankiewitz, Phys. Chem. Glasses, **2**, 1961, 111, 115
72. C. T. Sun, V. S. Chan, ASTM STP, **674**, 1979, 418
73. J. K. Lavism, F. E. Bartell, Anal. Chem. „O", 1948, 1182
74. K. H. Boller, *Resume of fatigue characteristics of reinforced plastics laminates subjected to axial loading*, Techn. Documentary Report No. ASD-TDR-63-768, 1975
75. L. J. Broutman, *Glass resin joint strengths and their effect on failure mechanisms in reinforced plastics*, Polymer Eng. Sci., **6**, 1966, 263
76. J. L. Broutman, Proc. 19th Conf. SPI Reinforced Plastics Div., 1964, sec. 9-C
77. J. Cook, J. E. Gordon, *A mechanism for the control of crack propagation in all brittle systems*, in: Proc. Roy. Soc., **282 A**, 1964, 1391
78. J. Taborsky *et al.*, *Effect of interfaces*, 1st Conf. on Mechanics, Prague, **5**, 1987, 78–81

ПРОЧНОСТЬ И ПОВЕДЕНИЕ КОМПОЗИЦИОННЫХ МАТЕРИАЛОВ ПРИ РАЗРУШЕНИИ

Резюме

Прочность и поведение композиционных материалов при разрушении, особенно в критических условиях, являются наиболее важными характеристиками, определяющими качество и долговечность конструкций, изготовленных из них. В работе приводится обзор количественных аспектов прочности и деформируемости вплоть до разрушения. Эти аспекты дают полную картину исследуемой проблемы.

В работе делается упор на выяснение механизма деформаций, механизма процесса разрушения, а также скорее на полном разрушении, чем на абсолютном значении прочности. При этом подчеркивается значение структуры материала, как независимого физико-химического явления (структура с дефектами). Приведены основные величины и механизмы, действующие в материалах различной структуры, на примере структур с твердыми и хрупкими включениями или с деформируемой матрицей, а также упоминаются возможности повышения прочности.

Особое внимание уделяется усталостной прочности, а также процессам, происходящим в композитах (особенно с волокнистой структурой), во время повторяющихся и циклических нагрузок.