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Submicrocrack formation under stress

Prof. S. N. ZHURKOV, Dr. V. S. KUKSENKO, Dr. A. I. SLUTSKER
Ioffe Physical Technical Institute, U.S.S.R. Academy of Sciences,
Leningrad, U.S.S.R.

Summary

Micro cracks generated in crystalline or amorphous polymers by tensile loading have been detected by small-angle X-ray diffraction techniques. These cracks measure some hundredths of a micron. They have the shape of flat discs and are orientated with their planes perpendicular to the loading axis. The time dependence of crack formation and development in bodies under constant tensile load, was studied. The concentration of cracks under load was found to be very high (10^{15} - $10^{16}/cm^3$) which makes the distances between the cracks commensurable with their own dimensions. All this makes us believe that cracks affect each other, the stress field around one crack acting upon the contours of its neighbours. This fact may result in an acceleration of the crack growth and contribute to their confluence. All these phenomena are discussed and suggestions are made that polymer fracture is a process associated with the 'collective' growth of cracks and not with that of single cracks as previously considered.

Introduction

It is known that polymers often become opalescent under stress. Haward was the first to notice this phenomenon [1]. Then Zhurkov, Marikhin and Slutsker [2, 3] observed diffusive scattering by light scattering and small-angle X-ray diffraction techniques and attributed it to the appearance of submicrocavities in a loaded polymeric material. Measurements of the distribution of visible light scattering in transparent polymers and small-angle X-ray diffraction permitted calculation of the dimensions, shape and concentration of submicrocavities. It was proved that the cavities generated under load had the shape of discs orientated perpendicular to the axis of extension. The disc dimensions ranged from tens to hundreds of angstroms. The concentration of the submicrocavities caused by loading was found to be quite considerable; as high as 10^{15} – $10^{16}/cm^3$ in order of magnitude. According to Zhurkov, Marikhin and Slutsker [2], the cavities are, for the most part, distributed uniformly throughout the specimen being deformed. Close to the surface, however, they are somewhat more numerous. The shape of the cavities and their orientation with respect to the mechanical stress enabled Zhurkov *et al.* [2, 3] as well as Kuksenko *et al.* [4] to consider them as initial cracks for subsequent fracture. From this standpoint a study of the submicroscopic defects generated in mechanically loaded bodies, by light scattering and small-angle X-ray diffraction techniques, is important for investigation of the mechanical properties of polymers.

The present paper which is a further development of this trend, aims to reveal the relationship between the deformability, the fracture of polymers, and the formation of submicrocracks under load.

The authors have examined the mechanical properties (deformation and fracture) and the submicroscopic crack formation (number and size) in both crystalline and amorphous polymers.

The detection of submicrocracks

To detect submicrocracks under load the small-angle X-ray diffraction technique was used. Its theoretical and practical bases, having been stated by Guinier [5], will not be discussed here. The experimental conditions under which our measurements have been made will only be briefly considered.

Small-angle scattering in polymers was measured with a slit unit which recorded the X-ray radiation by scintillation counters; this equipment is described by Slutsker, Egorov, [6].

Figure 1 displays a diagram of the test equipment. A flat, dumb-bell shaped polymer film specimen was placed in a vacuum chamber in the path of a well collimated X-beam. Any required load could be imposed upon the specimen to be examined without disturbing the vacuum. (This is necessary to exclude parasitic scattering by air). The specimen remained in a loaded state during the measurements of small-angle scattering. To calculate the size and the number of scattering defects generated by the load, the angular distribution of scattering intensity was measured. In the simplest case, for cavities of identical size irregularly dispersed throughout the body, the intensity of scattering for the transmission of a thin beam will be approximately described by Guinier's formula [5].

$$I(\phi) \simeq A.N.V^2.e^{-(4\pi^2/3\lambda^2).L^2.\phi^2} \qquad (1)$$

I – diffracted intensity at angle ϕ
L – cavity dimension in the scattering measurement direction
V – scattering cavity volume
N – number of scattering cavities in cm^3
λ – wavelength of X-ray radiation
A – coefficient taking into account the scattering ability of cavities and the geometrical test conditions.

The measurements of scattering have been carried out in the angular range from several minutes of angle to 1-2 degrees. When using Cu $K\alpha$ radiation with the wavelength $\lambda = 1{\cdot}54$ Å, the dimensions of cavities which can be recorded lie in the range of 10-1,000 Å. Figure 2 (a) shows the scattering curves for a number of polymers in a strained state. The slit beam was orientated with its long side along the axis of extension, being directed perpendicular to the plane of the film. It is seen that the intensity of the diffuse scattering increases sharply with the increase of the stress. The diffusive scattering begins with loading of the polymer and diminishes or entirely disappears as soon as the load is removed.

Formula (1) enables us to estimate from the angular dependence of scattering, the dimensions and concentration of the cavities. For this purpose the experimental data are plotted in the coordinates $\log I - \phi^2$. Cavities of identical size fit well to a straight line whose slope is determined by the particle size L. while the intensity at $\phi = 0$, which is found by extrapolation of the straight line to $\phi = 0$, will be determined by the cavity concentration N.

The cavities generated in extended polymers are not of strictly identical size. In fact, the scattering curves transformed in the coordinates $\log I - \phi^2$ are not straight lines; they are slightly curved. Figure 2 (b) shows the scattering curves plotted in the coordinates $\log I - \phi^2$. The fact that the curves are not straight lines complicates the calculation of the cavity number and size but, fortunately, it does not evoke any fundamental difficulties. According to Guinier [5] one may estimate by averaging to a sufficiently reliable degree, both the number, the shape and the dimensions of cavities generated by mechanical loading.

The reliability of the data on the cavity size and number calculated from the scattering curves is confirmed, according to Zhurkov, Marikhin, Slutsker [2], Kuksenko, Slutsker, and Jastrebinski [4], by the density decrease in deformed specimens.

Thus the scattering technique allows measurement of the cavity size and concentration, and offers the possibility of following the kinetics of submicrocrack formation under various mechanical conditions. Moreover, it makes possible a study of the relationship existing, between the number and dimensions of cracks generated under loading, and the deformability and fracture of polymers.

Kinetics of submicrocrack formation

Measurements of the time dependence of cavity concentration and dimensions were carried out in the following way. A specimen placed in a small-angle X-ray diffractometer was loaded, the tensile stress being maintained constant during the whole test by means of a special device suggested by Tomashevski and Slutsker [7]. Measurements of small-angle X-ray diffusive scattering were carried out at various times. The dimensions of cracks directed perpendicular to the applied load being of interest from the mechanical standpoint, the measurements were done in the transverse direction which permitted the assessment of the greater dimension of the discs. Figure 3 illustrates the results of measurements for a uniaxially orientated capron film at a stress of 20 kg/mm^2. The specimen was extended along the axis of orientation. Curves 1-3 correspond to different points of time for a specimen under load. Using formula (1), one could find from the slope of log $I - \phi^2$ curves at low values of ϕ, the transverse dimensions of disc-shaped cracks. For a capron film they were found to be ~ 250 Å. An analysis of the curves has shown that time changes occurring in crack dimensions are very slight since the slopes of curves 1-3 change quite insignificantly. Hence the rise in diffuse scattering intensity with time, is mainly due to the crack multiplication. To estimate the number of cracks generated by loading, the volume of a crack should be known (formula 1), which means that longitudinal dimensions of the cracks are required too. To that end, the scattering curves measured in the longitudinal direction were treated in a similar way. Then, supposing the cracks to be of elliptic form, the average crack capacity V was calculated and finally, the concentration of the cracks generated under loading was determined. Thus the kinetics of crack formation in polymers under load was measured. Figure 4 (curve 1) shows changes in the crack concentration N vs. time for a specimen at constant stress. These data indicate that, for a capron film, while the crack concentration changes with time increases by several orders, changes in crack dimensions are relatively insignificant.

To find out whether this kind of crack formation and growth kinetics is peculiar only to crystalline polymers, measurements were carried out on amorphous polymers too. In the Fig. 4 (curve 2) the results obtained on amorphous polyvinylchloride are presented. The pattern proved to be similar to that of the crystalline polymer.

The cross-dimensions of cracks in a polychlorovinyl film were found to be 600 angstroms. It is seen that, at a constant stress, the rate of generation of cracks varies widely with time. The rate of their formation is initially fairly high but in the course of time it vanishes. This consistent pattern in the crack accumulation kinetics is characteristic both of amorphous and crystalline polymers (Fig. 4, 1 and 2).

It should be noted that the form of curves showing the accumulation of cracks under loading resembles that of the creep curves in solids. This likeness, apparently, indicates the existence of some connection between crack formation and creep in polymer materials. However this problem requires a further study.

Dependence of crack formation on stress

It should be of interest to follow the kinetics of crack formation at different tensile stresses. The experiment was carried out on identical specimens of an orientated capron film, under the same conditions. The specimens were placed in a vacuum chamber of an X-ray diffractometer and subjected to various stresses. The kinetics of crack accumulation

(N) were measured in the same way as shown in Fig. 4 (curve 1), and the velocity of the concentration increase (dN/dt) at the initial stage was calculated. Curve 1 in Fig. 5, represents the crack accumulation velocity at the first stage related to the stress σ applied to the specimen; curve 2 is the same dependence plotted in semi-log coordinates. From these data it is evident that the rate of submicrocrack formation increases with the magnitude of the applied load by an exponential law. This dependence proves to be general for all polymers, both crystalline and amorphous. It should be noted that the observed exponential dependence of the crack formation rate upon the applied load is essential for understanding the crack formation mechanism as well as the mechanism of polymer fracture.

Measurements of submicrocrack dimensions

The dimensions of cracks originating in polymers do not vary appreciably with the load magnitude and time. The scattering curves in the coordinates log $I - \phi^2$ are not rectilinear which, as we have noted, indicates the cracks not to be of identical size. The shape of the scattering curves permits analysis of the dimension distribution of the cracks and measurement of the time changes in this distribution occurring in a loaded polymer. According to Guinier [5], the estimate of the dimension distribution of cracks is based on the fact that large cracks can scatter mainly into small angles, while scattering from small cracks involves larger angles. Thus, the diffusive scattering intensity in the region of large angles may be approximately considered as conditioned only by small cracks. This permits: (1) an assessment of the dimensions of small cracks from the slope of the tangent to the 'tail' part of the scattering curve, (2) a determination of the number of small cracks and (3) separation from the total scattering of a component from small cracks which, in its turn, allows estimation of the dimensions and the number of large cracks. All this offers the possibility of investigating the time and size dependence of crack distribution.

The results of analysis suggest that it is mainly small cracks that are being formed in a polymer under load. For example in capron, their concentration is found to be 80 to 90% of the total number. Small cracks are 2 to 3 times smaller than the larger ones. The former as well as the latter are disc-shaped and orientated with their planes perpendicular to the stress axis. It was revealed that although the average dimensions of cracks generated under load undergo only slight time changes, the proportion of large cracks gradually increases. In the preruptured state, the proportion of large cracks in propylene is doubled, as compared to that observed on loading. Moreover, it was noticed that the 'head' part of the scattering curve (dependent chiefly on large cracks), becomes steeper in the course of time. It indicates that in a specimen under load more and more large cracks appear that did not exist before; these might have been formed either by crack growth or crack coalescence.

Connection between crack formation and strength

As shown above, concentration of cracks generated by mechanical loading undergoes similar changes, dependent on time and tensile stress, both in amorphous and crystalline polymers. These cracks are of very small dimensions, on average not exceeding some hundreds of angstroms. The concentration of submicrocracks in a loaded polymer amounts to 10^{15}-$10^{16}/\text{cm}^3$. Submicroscopical porosity resulted from mechanical stress and revealed by small-angle X-ray diffraction techniques would be expected to affect the polymer properties. In the first place it would be of interest, to our mind, to elucidate the role of cracks as submicrospace concentrators of stress and interpret their participation in the mechanism of polymer rupture. For this purpose the concentration of developing cracks and the lifetime of polymer under load with uniaxial extension were measured simultaneously. The experiments were carried out on an orientated capron film from which were cut flat dumb-bell shaped specimens. These were put into the vacuum chamber of an X-ray diffractometer and subsequently loaded; measurements of the creep curve (until rupture) and the angular dependence of diffuse scattering intensity were then carried out. Tensile stress (σ) dependence of lifetime (τ) is shown in Fig. 6. The next Figure (Fig. 7) shows, as an example, curves of crack accumulation occurring with time under those conditions.

The three curves presented in Fig. 7 correspond to the specimens marked with arrows in Fig. 6, whose lifetime is found to be $\tau_1 = 116$ sec, $\tau_2 = 6$ hours, $\tau_3 = 105$ hours. The curves of crack accumulation are similar to the creep curves, previously mentioned. From the data obtained one may see that the kinetics of crack formation under creep conditions, both with large and small tensile stress are similar. Submicrocracks are being formed fairly intensively at the beginning and then the rate of crack formation gradually reduces. We have followed the kinetics of crack accumulation until rupture, the last point of the curves (Fig. 7) corresponding to the crack concentration observed in the specimen just before rupture. Of great interest is the fact that the value (Nr) of the crack concentration before rupture will be nearly the same whether the rupture is caused by a large or a small load. At the same time the magnitude of stress, as we see from Fig. 7, considerably affects the crack formation rate which increases with the growth of load by several orders. The concentration, Nr, for orientated capron was found to be $1{\cdot}2 \times 10^{15}/\text{cm}^3$. To make certain that the value of Nr was independent of the load, measurements were carried out with variations of rupture stresses ranging from 38 kg/mm^2 to 48 kg/mm^2. Such a change in rupture load results in lifetime varying by 4 to 5 times. But, as seen from Fig. 8, Nr is maintained constant, with sufficient accuracy.

It is essential to do concentration measurements with different methods of strength testing. Figure 9 shows curves of crack accumulation in capron, with constant rate of deformation (curve 1) and with gradual growing of stress (curve 2). The shape of crack accumulation curves has undergone sharp changes as compared with the constant load test (cf. Fig. 7), but the final concentration Nr has remained the same at $1{\cdot}2 \times 10^{15}/\text{cm}^3$. This allows us to consider the submicrocrack concentration Nr to be characteristic of the prerupture state and, apparently, to affect in a considerable way its strength properties. Similar behaviour of the crack concentration may be observed in other polymers. In Table I are presented values Nr for four polymers. There are also presented averaged crack dimensions L in the transverse direction. It will be seen from the Table that both crack dimensions and final crack concentration Nr characteristic of the prerupture state do not considerably vary from polymer to polymer.

Formation of cracks close to a cut

If we make a cut at the edge of a flat specimen (test diagram is shown in Fig. 10) and observe the X-ray scattering from such a specimen when loaded, as a rule, the scattering will be negligible and will lie on the border of the test sensitivity. In this case submicrocracks are not being formed throughout the specimen and the diffuse scattering is lacking. This result depends on the low magnitude of load which such a specimen can bear. As was shown, the rate of crack formation increases exponentially with the growth of stress (Fig. 5), and in case of specimen with a cut it does not reach the required value. Under such conditions cracks develop very slowly and they are too scarce to be discovered by the scattering technique. However, the stresses close to the tip of the cut being considerably higher, a greater number of submicrocracks should be expected there. To make sure of this we directed a narrow beam of X-rays perpendicular to the surface of the specimen into the region adjoining the cut.

Scattering was measured by passing a beam at different distances close to the cut tip. As it was expected, the intensity of diffuse scattering which occurs in this case reduces when moving off from the tip. Crack dispersal under these conditions is shown in Fig. 10. Concentration of cracks rapidly diminishes and at a relatively short distance from the tip becomes immeasurably small. Drawing nearer to the tip of the cut the crack concentration considerably increases and tends to the limit value. It seems of interest that concentration maxima of cracks generated just at the tip of a growing macrocrack was found to be equal to the typical concentration Nr, which one may observe in prerupturing state in a specimen without cut in uniaxial extension. (The experiment with cut having been carried out on the same capron films as before). As a macrocrack propagates, the whole picture shifts with it, so that the concentration in advance of the crack tip is always concentration Nr.

Discussion

Small-angle X-ray diffraction technique used to study submicrocracks have allowed us to follow the kinetics of crack formation and the effect of this process on polymer strength. The above examples indicate that the process of submicrocrack accumulation in a loaded body develops smoothly. The rate of crack formation increases according to an exponential law with increasing load and eventually approaches zero under constant load. The crack formation process results in producing a quite definite crack concentration Nr in a loaded polymer. This final concentration of submicrocracks in the prerupture state proved to be fairly stable and not to depend on the magnitude of rupture stress and the method of loading in a tension test. This experimental fact, to our mind, permits us to suggest a theory for the mechanism of polymer rupturing in which submicrocracks play a decisive part. First of all it should be emphasized that with a crack concentration (Nr), for the polymers in question (Table 1) the average distances (R) between submicrocracks orientated in parallel will be commensurable with their dimensions (L). With such distances the fields of stress created by submicrocracks can overlap. Hence the crack will find itself under the action not only of average stresses created by external load, but also of an additional field from a neighbouring crack. In this case two parallel cracks will affect one another, creating favourable conditions for their growth and coalescence. A large crack resulting from coalescence will find itself in still more favourable conditions than the initial crack and will again join up with its neighbours to form a new crack of a still greater size. This process of crack coalescence, which had accidently started in a spot where cracks were lying in the most opportune way quite close to each other, will accelerate and, as a result, a main crack will arise and, subsequently, the specimen will rupture.

Such is the outline of the mechanisms of polymer fracture arising from this point of view.

Now we will try to discuss in detail, in the light of all facts available, the development of fractures in polymer material under the action of tensile stress.

According to the kinetic conception of Zhurkov [8], breakdown of a solid results from a gradual accumulation of chemical bond ruptures. Under the action of thermal motion and mechanical load, chemical bonds break down and the concentration of broken bonds in the body increases contributing to its rupture. For the rupture of a body to take place time is required to accumulate a sufficient number of broken bonds. This time

determines the lifetime of a material under load. As known from experimental data the time of rupture or lifetime of a material under load (τ) is related to the magnitude of the rupturing stress (σ) and absolute temperature (T) by relation established by Zhurkov [8]

$$\tau = \tau_0 \, e^{(u_0 - \gamma\sigma)/kT} \qquad (2)$$

which fits quite well to materials of different nature and structure (metals, non-metallic crystals, glasses, polymers and others).

In accordance with this kinetic equation, the rupture tensile strength (σ) is related to the test time (τ) and temperature (T), and is also determined by the parameters of the material: τ_0, U_0 and γ. Parameter τ_0 – period of atom thermo-fluctuations in solid, being for solids 10^{-13} sec, U_0 – activation energy of the rupture of interatomic bonds, γ – structure sensitive coefficient dependent on the microstresses acting on the bond being ruptured.

With constant temperature (T) equation (2) will be simplified and can be written as

$$\tau = \tau_0 \, e^{-\alpha\sigma}$$

where A and are constant coefficients.

If we assume the lifetime of a body under load to be really determined by the kinetics of the accumulation of chemical bond ruptures, then according to equation (3) one should expect the rate of bond rupture to increase exponentially with the growth of mechanical stress (σ), Zhurkov and Tomashevski [9], using a paramagnetic resonance technique for recording ruptured chemical bonds in polymers, have demonstrated that the relationship between the lifetime (τ) and the rate of chemical bond rupture (V), is valid. Further one may assume that, due to heterogeneity of real bodies, the bond rupture does not proceed uniformly throughout the polymer. Owing to exponential dependence, the probability of chemical bond rupture in locally overstressed microregions will be higher than average and such spots will become centres of primary bond destruction. This will result in accumulation of bond ruptures and formation of submicrocavities.

In that connection we can see why the rate of chemical bond rupture and the number of initial cracks prove to be interrelated. The rate of breaking in both cases is found to depend exponentially on tensile stress. Zhurkov and Tomashevski [9] have revealed exponential dependence of the rate of the chemical bond rupture, while in the present work is demonstrated exponential dependence between the rate of crack formation and the magnitude of stress (Fig. 5). Thus the generated submicrocracks will increase in number until their concentration reaches the value Nr which was discussed above. Then the process of the initial crack accumulation will be accompanied by the process of uniting submicrocracks into larger cracks which, finally, will bring about the specimen rupture.

The supposition of a mechanism of polymer rupture through coalescence of submicrocracks is quite in agreement with the estimate of the concentration (Nr) in the prerupturing state. As was shown on four large polymers, the density of submicrocracks is found to be so high (Table 1) that one must admit their interaction and consequently, their coalescence.

To substantiate the assumption of crack coalescence in polymers of high crack concentration, one may use theoretical calculations of the crack interaction. According to Panasuk [10], for the case of two thin flat cracks situated collinearly in an elastic medium, interaction occurs when $a/L<3$, where a is the distance between the nearest ends of cracks, L – crack length. With $a/L>3$, the cracks are considered to be virtually isolated and, as soon as a definite stress is reached in the body, they begin to grow independently of each other. With $a/L<3$, under the action of tensile stress coalescence of the cracks occurs first by cross-piece rupture and subsequently the newly produced large crack propagates.

The values of R/L presented in Table 1 lie close to the value a/L applied in the calculations. Therefore the values R/L obtained from experiments, which for the examined polymers are found to be 2-3, indicate that a sufficiently effective interaction of cracks resulting in their joining up is quite possible. Of course, the cracks in polymers under load and the idealized cracks in calculations are of different shape. However, at least two facts may be referred to as enhancing the crack interaction.

(1) Table 1 presents the average values R/L. No doubt that among the cracks spread in polymer in a chaotic way there will be a lot of cracks running close to each other. Naturally they will exert a greater influence on each other.

(2) For the growth of real cracks the stress at their tips does not need to be increased up to the theoretical value (which is always required according to Griffith's criterion). According to Zhurkov and Tomashevski [11] a crack will grow at any stress, even if it is very low, its growth rate being exponentially dependent on tensile stress. Hence any additional stress from a neighbour crack field results in a sharp acceleration of the given crack. Therefore even rather remote cracks can exert a marked effect on the growth rate of that crack.

Coalescence of cracks, once started in some place, must develop, to our mind, at a brisk pace. It would be of interest to interpret the growth of artificial macrocracks by means of this mechanism. According to Zhurkov's and Tomashevski's data [11] the rate of crack growth in polymer can vary with load by many orders, from microns per minute to

velocities approaching the velocity of sound. Investigations on the rate of crack propagation were carried out with uniaxial expansion on thin plates with a through crack at the edge. No limit predicted by Griffith's concept was revealed. As has already been noted, however small the load might be, the crack goes on growing. In all cases the crack growth rate depends exponentially on the magnitude of tensile stress.

To compare the data from the paper referred to, we have investigated, under similar conditions, the crack formation at the tip of an artificially created macrocrack. It was proved that during the macrocrack propagation the concentration of submicrocracks at its tip will be the same concentration Nr that is observed in specimen without macrocracks in the prerupturing state. As the tip of the large crack is being removed, the concentration of submicrocracks moves with it.

The increase of submicrocrack concentration at the end of the artificial crack is undoubtedly connected with the increase of the stress produced by the crack itself. This is substantiated by the fact that when moving off from the crack end, the concentration of submicrocracks (Fig. 10) decreases and so does the stress field. Hence the stress concentration around a macrocrack causes the appearance and development of a great many submicrocracks in this ergion. As soon as their concentration reaches the value Nr and the fields of individual submicrocracks begin to overlap, the process of coalescence will develop and larger cracks will be produced. The latter being in immediate proximity to the tip of the macrocrack may join it and thus increase its length.

In that scheme the growth of macrocracks does not proceed in the direct way but by 'imbibing' the smaller cracks which make their appearance in the field of macrocracks independently.

Conclusion

The data and the discussion presented here should be considered as being the first direct experimental results as well as the first attempts at following the process of fracture from the earliest stages. Undoubtedly further investigations are required to elucidate the question of generality of all revealed regularities and perfect reliability of the suggested picture of rupture. However even the first experiments obtained are so encouraging that a further study in this direction seems to be indispensable.

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Table 1

Polymer	N_r cm^{-3}	*L* μ	$R = \sqrt[3]{1/N_r}$ μ	*R/L*
Polycaproamide	1·2 × 10^{15}	0·025	0·09	3·5
Polypropylene	0·7 × 10^{15}	0·04	0·11	2·8
Polyvinylchloride	4·5 × 10^{12}	0·06	0·11	1·8
Polyvinylbutyril	3·5 × 10^{12}	0·05	0·15	3·0

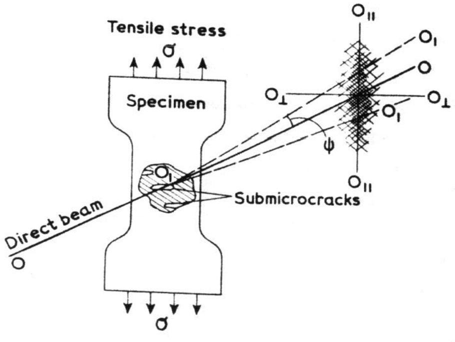


Fig. 1. Scheme of X-ray scattering submicrocracks.

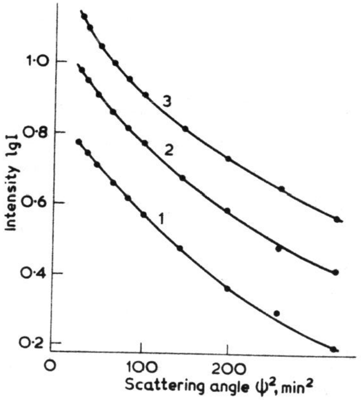
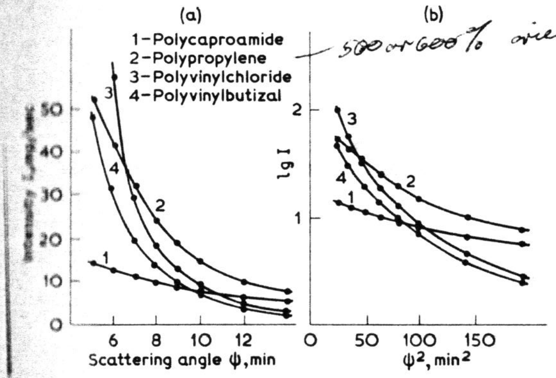


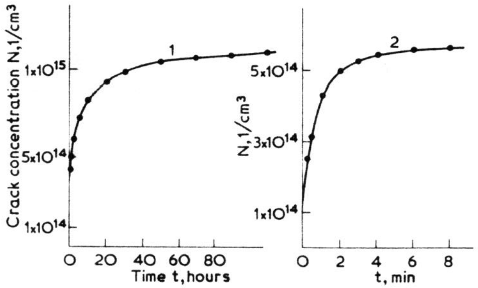
Fig. 3. Scattering curves in the transverse direction for capron loaded σ = 20 kg/mm² at different points of time: 1-30 minutes; 2-1 hour; 3-120 hours.



— 500 or 600 % orientⁿ. tested at 100° c.

Fig. 2. Scattering curves in the transverse direction for extended polymers.

Fig. 4. Curves of submicrocrack accumulation under constant load. (1) capron σ = 20 kg/mm²; (2) polyvinylchloride σ = 6 kg/mm².



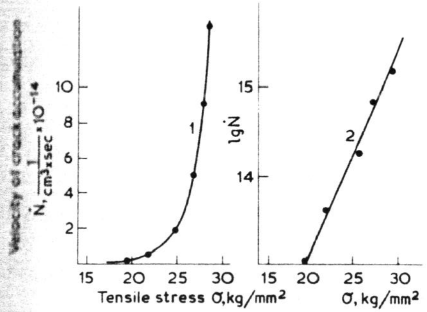
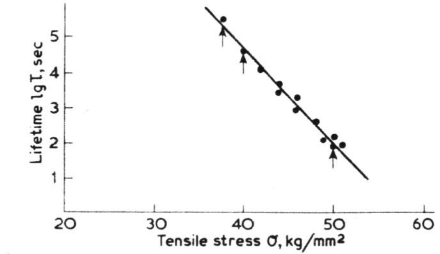


Fig. 5. Dependence of the rate of crack accumulation at the initial stage on tensile stress for capron.

Fig. 6. Time dependence of the strength for capron.



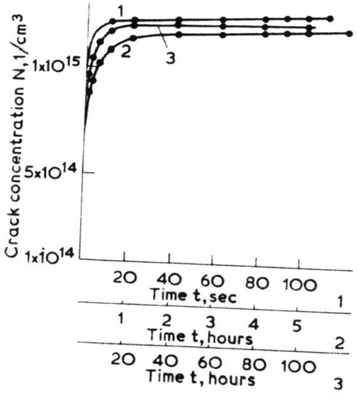


Fig. 7. Curves of submicrocrack accumulation for capron loaded by (1) $\sigma = 50$ kg/mm^2; (2) 40 kg/mm^2; (3) 38 kg/mm^2.

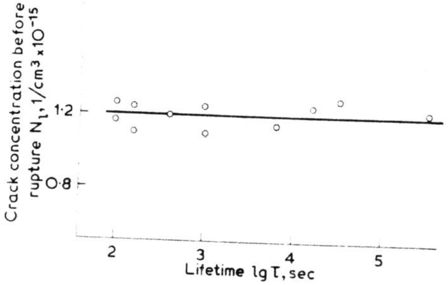


Fig. 8. Relationship between submicrocrack concentration in advance of rupture and lifetime of specimen under load for capron.

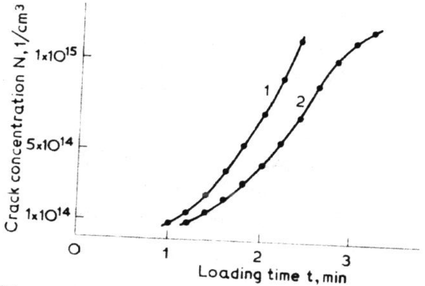


Fig. 9. Curves of submicrocrack accumulation for capron: (1) with constant rate of deformation; (2) with constant rate of loading.

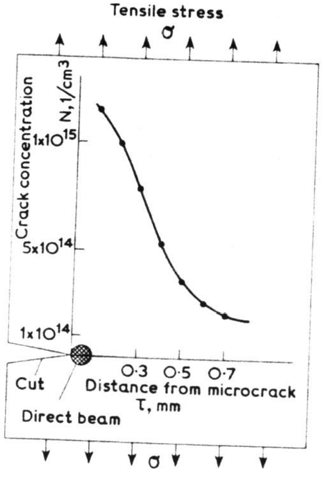


Fig. 10. Scheme of submicrocrack measurements in the region of cut tip. (1) Change in crack concentration along the macrocrack growth direction for capron.