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MECHANICS

1967

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Abstract

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UDC 678:539.4.01

MECHANICS

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ON THE LOOSENING OF THE STRUCTURE AND STRENGTH OF POLYMERIC MATERIALS

(Presented by Academician V. V. Novozhilov on 6 X 1966)

As is known, irreversible deformation of solids is characterized by an internal shear mechanism. This is manifested in the fact that, even in substantially irreversible deformation processes, the change in the volume of a specimen is usually small and close to that determined by the material' s law of elasticity. In polymers the situation is similar with high-elastic deformation, the main contribution to which also comes from the strain deviator.

Let $\theta = (V - V_0)/V_0$ be the relative change in the volume of the specimen. In the general case $\theta = \theta' + \theta''$, where θ' and θ'' are the reversible and irreversible components, respectively. In accordance with what has been noted, usually $|\theta| \ll 1$ and, moreover, $|\theta''| \ll |\theta'|$. One of the possible reasons (admissible already on purely logical grounds) for a different situation is the development in the body, under certain conditions, of a network of microscopic pores and cracks. In this case the component θ'' may be comparable with θ' and represents an important macroscopic characteristic of the state of the material.

V. V. Novozhilov recently drew attention to the possibility of an irreversible change in the volume of a specimen (element of a medium) due to a certain kind of microscopic structural defects developing in the course of deformation (^{1}). In (^{2}) the results of an experimental verification of the considerations set forth in (^{1}) are described. These works consider engineering metals for which, as shown in (^{1}), microscopic loosening of the structure leads to a value of θ'' noticeably different from zero under alternating cyclic loading of specimens. There is reason to believe that, in many high-polymer materials, an analogous effect should already be observed in experiments with single loading. Below, the results of a number of such experiments on polymer specimens, carried out by the authors, are presented and discussed.

Solid cylindrical specimens (d = 6 mm, l_0 = 36 mm) of fluoroplastic-4, rigid vinyl plastic, and organic glass were subjected to uniaxial tensile loading by direct suspension of a weight from the end of the specimen. In this case some of the fluoroplastic specimens were tested at a load value unchanged for each

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Fig. 1

Figure 1: Fig. 1

Fig. 2

Figure 2: Fig. 2

specimen, i.e., in a creep regime under a constant conditional stress; the other part was tested with the load increasing in small and equal steps throughout the entire process. Specimens of vinyl plastic and organic glass were tested only under loading of the second type. In this case the loading regime can be characterized by the value of a certain average rate—the angle of inclination α of the equivalent straight line in the plane referred to the coordinates conditional stress—time. In all cases the test was carried through to failure of the specimen.

In each experiment the value $\theta = (V - V_0)/V_0$ was determined for a number of current states and, in addition, the residual value after failure of the specimen. It is essential that, in specimens of fluoroplastic and organic glass, the working part remained cylindrical right up to the onset of macroscopic failure. In specimens of vinyl plastic, at a certain stage of the process a neck appeared; however, after a short interval of time (in comparison with the total test time) following the onset of necking, visible failure of the specimen also occurred. Taking this into account, the current volume of the working part

was determined from the results of direct measurement (with carefully calibrated calipers and a micrometer with scale divisions of 0.05 and 0.01 mm, respectively) of its length and diameter. The volume after fracture was determined hydrostatically by weighing on an analytical balance.

Fig. 1. Results of tests of fluoroplastic specimens: **a**—at constant stress σ (referred to the undeformed cross section); **b**—at a fixed nominal loading rate

In Figs. 1 and 2 the obtained values of θ at the moment of fracture (dots) and after fracture (crosses) of the specimens are presented. As can be seen from Fig. 1, for fluoroplastic specimens the value of θ at the moment of fracture reached 25%, while the change in volume by this moment was almost entirely residual and practically did not depend either on the value of the stress σ (in creep experiments at constant σ) or, under loading of the second type, on the rate α.

Fig. 2. Results of tests of vinyl plastic specimens

For specimens of organic glass, the value of θ at the moment of fracture was about 2%, while the residual component was practically equal to zero. A somewhat different picture was obtained for vinyl plastic specimens: here θ at the moment of fracture again almost coincides with its residual component; however, in contrast to fluoroplastic specimens, it has a value of 2–3% and depends appreciably on the loading rate (Fig. 2). It should be noted that each experi-

Fig. 3

Figure 3: Fig. 3

mental point in Figs. 1 and 2 represents an average obtained from tests of no fewer than 8 specimens (in all, about 200 specimens were tested).

We also note that the specimens were loaded under conditions ordinary for laboratory mechanical tests of materials, i.e., in the absence of radiation (noticeable from the standpoint of its influence on macroscopic mechanical properties), exchange of matter with the external environment, and with an almost constant and uniform temperature field ($\approx 20^\circ$) in the latter. The obtained values of the residual volume change of the specimens during deformation under these conditions can be explained only by assuming that, as a result of deformation, a network of microscopic pores or cracks develops in the specimen. In fact, along with such microscopic loosening of the structure, the cause of a noticeable residual volume change of specimens under quasistatic loading in the indicated conditions could only be phase transformations, for example crystallization or recrystallization of the substance. However, as is known, owing to the orienting action of the field of macroscopic stress during static tension of polymer specimens, internal changes of this type proceed in such a way that the degree of crystallinity ultimately increases. The consequence of this is densification of the substance, i.e., a decrease in the volume of the working part of the specimens, whereas in the experiments, we emphasize, by the moment of macroscopic fracture the residual volume was always greater than the initial one.

The fact that the cause of this is microscopic rupture is illustrated by the results of electron-microscopic analysis of the stru-

ture of the specimens. Fig. 3 presents a typical electron micrograph obtained when photographing silver–carbon replicas from freshly cleaved surfaces of fluoroplastic specimens after testing. A system of crack-strips is visible, approximately perpendicular to the axis of action of the external tensile forces. As the deformation increases, these cracks, in the most defective places, merge into main (macroscopic) cracks, leading to failure of the specimen.

Naturally, therefore, the experimental results described above make it possible to formulate a comparatively simple macroscopic fracture criterion. Indeed, as already noted, for fluoroplastic specimens at the moment of failure $\theta \approx \theta''$, and failure occurs at one and the same (to within statistical scatter), independent of the stress (Fig. 1a) and of the loading rate (Fig. 1b), value of θ''. If this critical value of θ'' is denoted by C, then, consequently, for an arbitrary state of the specimen (with the accuracy mentioned) $\theta'' \leq C$, while at macroscopic failure

$$\theta'' = C. \qquad (1)$$

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Fig. 3. Electron micrograph of a loosened specimen of fluoroplastic-4. The straight line $P - P$ is the direction of the axis of action of the external tensile forces. 24 000×

Since θ'' is a scalar (invariant with respect to transformations of the coordinate system), equation (1) is, in essence, a fracture condition in the spirit of those considered in theories of strength in classical materials resistance.

If, however, one does not restrict oneself to materials of the fluoroplastic type, then in this condition C is certainly not a constant (for a given material and temperature), but some function of the trajectory of the process in stress space. Thus, in contrast to fluoroplastic, for vinyl plastic the critical value of the residual volume change depends on the loading rate (Fig. 2). Moreover, there is no guarantee that, in experiments with more complex loading regimes, a dependence of C on other details of the loading trajectory (besides the rate) may not be revealed.

On the other hand, a considerable residual change in the volume of the working part (1-2% V_0 and more) was also observed in experiments with quasistatic loading of specimens made of other polymers ([3],[4]). Therefore, undoubtedly, with the reservation noted concerning C (and with the reservation concerning statistical scatter), condition (1) covers a broad class of polymeric materials. Determining the limits of this class and specifying the dependence of C on the loading trajectory for the materials included in it are difficult problems, requiring the accumulation of experimental data and sufficiently complete information about the microscopic mechanism of fracture.

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Received
28 IX 1966

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