

PHYSICO-MECHANICAL PROPERTIES OF POLYMER MATERIALS AND THE FRICTION PROBLEM

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Aspects of the friction behavior of polymer materials associated with their molecular structure and responsible for the relatively small value of the modulus of elasticity as compared with the ultimate strength are examined. It is shown that, in view of the important influence of hydrostatic pressure on the mechanical properties of polymer materials, the mechanical characteristics obtained from uniaxial testing cannot be used in calculating the contact area and the forces of friction. Formulas are presented for calculating the mechanical characteristics under these conditions. The results of indentation experiments designed to simulate contact processes are discussed. It is shown that the resistance to deformation of the asperities on the surface of polymer materials is of the order of the yield stress, and not two or three times greater, as with metals. The results of contact creep studies are described and evaluated. The results of investigations of the mechanical and antifriction properties of filled polymer materials show that the forces of friction are inversely proportional to the modulus of elasticity, while the thermophysical characteristics are a function not only of the thermophysical characteristics of the filler and the base, but also depend to a great extent on the shape of the filler particles; thus, when a fibrous filler with a low coefficient of linear expansion is used, the thermal stability of the friction material can be considerably improved. The results of a study of the adhesion interaction of polymer materials under conditions of omnidirectional nonuniform compression and simultaneous deformation are presented. It is shown that the adhesion interaction is strong even at room temperature. Aspects of the mechanical properties of lubricants that determine their effectiveness in polymer friction are considered.

1. Solid friction is a complex process taking place under conditions that vary widely with time. The factors involved include temperature, strain rate, surface finish, chemical composition, the physico-mechanical properties of the surface layers, and the forces of adhesion on the small areas of actual contact. In the contact zones the material of the surface layers is subjected to omnidirectional nonuniform compression and shear. Both the normal and the shear stresses are functions of the contact area and the adhesion forces, which, in their turn, are functions of the elaborately interrelated parameters listed above. Under actual friction conditions the rubbing parts are separated by a certain intermediate layer — a lubricant or some adsorbed film, whose properties are also functions of velocity, temperature, pressure, clearance, etc.

Obviously, there is little real prospect of developing a general quantitative theory of friction in the near future. Therefore, a considerable proportion of friction research has been concerned with the experimental study of special, though important, cases of boundary or semidry friction. Many machine parts and mechanisms operate in this regime at low relative velocities. The boundary friction regime is also characteristic of the starting and stopping phases of machinery with antifriction systems. It is a feature of this regime that the surfaces of the rubbing parts come into direct contact, a process accompanied by some form of wear. The tangential resistance is determined by the contact area, the forces of adhesion between the parts in contact, and the resistance to deformation of the surface layers. In recent years, various attempts

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have been made to calculate the contact area between metal parts on the basis of certain assumptions concerning the distribution of irregularities over the contact surface [1, 2] and to compute the resistance to plastic deformation in the presence of tangential displacement of the individual contact [3-5]. In this case, it is assumed that the mechanical characteristics of the materials do not depend on the state of stress and, in particular, on the hydrostatic component of the stress, which in relation to the deformation of surface irregularities reaches values of the order of two thirds of the normal stress. These assumptions are based on the work of Bridgman and Vereshchagin and co-workers, who found that for metals at values of the hydrostatic stress close to the working stresses under contact loads the change in the mechanical characteristics is relatively small and, consequently, that Hertz's theory can be used for elastic deformation, and in the case of plastic deformation the model of a rigid-plastic body.

The question arises, to what extent are these assumptions applicable to polymer materials, whose deformation is essentially different from that of metals and other materials with a high Young's modulus. In fact, for metals the ratio of the ultimate strength to the modulus of elasticity is of the order of 0.001-0.01 and, consequently, under loading the change in the specific volume or energy of the system is small as compared with the energy of the binding forces, which also implies a small change in the deformation characteristics. For polymers the corresponding ratio is of the order 0.1-0.01, i.e., one or two orders greater. Thus, the change of volume accompanying deformation is of the order of several percent and the change in energy is of the order of the energy itself, which must lead to a significant change in the deformation characteristics. In fact, an investigation of the effect of hydrostatic pressure on the bulk moduli of thermoplastics [6-8] has shown that even at a pressure of the order of hundreds of kgf/cm^2 , the bulk modulus increases by tens of percent.

The results of a systematic investigation of the effect of hydrostatic pressure on certain physico-mechanical properties of polymer materials are presented in [9-15]. In [9, 10] it is shown that as the hydrostatic pressure increases, there is a significant increase in the modulus of elasticity of polymer materials of all classes: thermosets, amorphous and crystalline thermoplastics, and thermosets with a randomly distributed glass filler (Fig. 1).

For thermosets the $E(\sigma_h)$ curve has two parts: an initial section, on which the modulus of elasticity increases rapidly, and a second section, on which it increases more slowly. As subsequent investigations have shown [11], the presence of these two sections can be attributed to a partially irreversible compaction of the material under hydrostatic pressure. In the same way, for brittle materials it is possible to observe an increase in the ultimate strength and for plastic thermosets an increase in the high-elastic limit and an upward displacement of entire $\sigma(\epsilon)$ curve. The increase in strength is of approximately the same order as the increase in the deformation characteristics (Fig. 2). In [12], it is shown that the effect of hydrostatic pressure on brittle polymer materials stressed in tension depends importantly on the contact between the deformed body and the pressure-transmitting medium. If these specimens are protected from the oil, the strength increases much more rapidly, and at pressures of the order of $1.5-2 \cdot 10^4 \text{ kgf/cm}^2$, fracture becomes plastic rather than brittle. In [11, 13], it was shown that when polymer materials are subjected to high-elastic deformation in the presence of hydrostatic pressure the fall in density is much less than when the same specimens are subjected to uniaxial deformation, and in certain cases the density even increases (relative to the density in the undeformed state). We note that whereas the density of deformed metals changes by hundredths of a percent, the irreversible changes in polymer density are measured in percent (Table 1). It has also been found that hydrostatic pressure affects swelling and the number of cracks observed after deformation [13].

When materials such as PVC, polyethylene, etc., are subjected to uniaxial tension, the specimens turn white because of the appearance of numerous cracks.

When deformed in the presence of considerable hydrostatic pressure PVC retains its color, and such materials as high- and low-pressure polyethylene and teflon become more transparent than in the starting state. Further investigation has shown that in this case the number of transverse cracks is sharply reduced, which also means that the strength is considerably greater. Thus, in the friction process the surface asperities are deformed under conditions such that the mechanical characteristics of the materials are much higher than in the uniaxial state of strain. The orientation of the surface layer should also lead to the development of a layer with mechanical properties higher than those associated with orientation in uniaxial deformation and, moreover, with greater density. Obviously, in calculating the area of contact of such a layer with a friction counterpart it is not possible to employ the ordinary equations of the theory of elasticity or plasticity.

TABLE 1

$\sigma_h, 2$ kgf/cm ²	$\varepsilon, \%$	$D_0,$ g/cm ³	$D,$ g/cm ³	$\frac{D-D_0}{D_0}, \%$
Low-pressure polyethylene				
1	118	0,9440	0,8900	-5,72
1	135	0,9440	0,8630	-8,60
500	118	0,9440	0,9390	-0,53
500	259	0,9440	0,9372	-0,72
Teflon				
1	200	2,1823	1,9627	-10,05
1000	180	2,1748	2,1671	-0,35
PVC				
1	60	1,3916	1,3622	-2,11
1000	60	1,3934	1,3928	-0,04
Ebonite				
1	16	1,1852	1,1861	+0,05
2000 without protection	16	1,1852	1,1896	+0,37
2000 with protected surface	16	1,1852	1,1914	+0,52

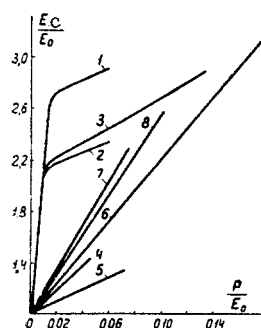


Fig. 1

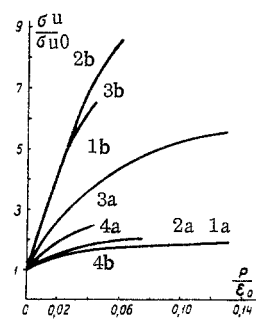


Fig. 2

Fig. 1. Relative Young's modulus in compression as a function of the ratio of hydrostatic pressure to initial Young's modulus: 1) urea-formaldehyde; 2) K-17-2; 3) FKP-1; 4) AG-4V; 5) K-78-51; 6) kapron; 7) PMM; 8) PVC.

Fig. 2. Relative strength as a function of the ratio of hydrostatic pressure to initial Young's modulus: 1) FKP-1; 2) K-17-2; 3) AG-4 V; 4) PMM. a) Compression; b) tension.

The nature of the effect of hydrostatic pressure on the mechanical properties of polymer materials is considered in [14, 15]. Polymer materials are solids with low density and a relatively low value of the bulk modulus due to the fact that it is basically determined by van der Waals forces [16]. With respect to density and elastic properties polymer materials resemble molecular crystals, for which it has been experimentally shown that, with a good approximation, the isothermal bulk modulus may be assumed to be a function of specific volume only. This corresponds to the assumption that the bulk modulus is a function of the potential energy of the lattice and is relatively independent of the phonon component [17, 18].

In [14], it is assumed that the bulk modulus of polymer materials is likewise a function of specific volume only. From this it follows that the equation of state for polymer materials can be written in the form

$$P = P_0(v) + P(T), \quad (1)$$

where $P_0(v)$ is a function of specific volume only, and $P(T)$ is a function of temperature only.

It follows from Expression (1) that for given polymer materials different isotherms can be combined by translation, as has indeed been demonstrated for a number of polymer materials at low homologous temperatures [14]. The temperature-pressure equivalence conditions are written as follows:

$$\Delta T = \frac{\sigma_h}{\alpha_v K}, \quad (2)$$

where α_v is the coefficient of volume expansion, and K is the secant bulk modulus.

Assuming that the temperature dependence of the bulk modulus is linear and using Relation (2), we obtain the following expression for the dependence of the bulk modulus on temperature and hydrostatic pressure:

$$\frac{K}{K_0} = 0.5 \left[1 + \left(1 + \frac{4\alpha_h}{\alpha_v} \frac{\sigma_h}{K_0} \right)^{1/2} \right], \quad (3)$$

where

$$\alpha_h = \left| \frac{1}{K_0} \left(\frac{\partial K}{\partial T} \right)_p \right|,$$

or for small values of the ratio σ_h/K_0 at constant temperature

$$\frac{K}{K_0} = 1 + \frac{\alpha_h}{\alpha_v} \frac{\sigma_h}{K_0}, \quad (4)$$

where K_0 is the initial values of the secant modulus. Equation (3) is satisfactorily confirmed by the experimental data.

Since, as an investigation conducted in our laboratory has shown, Poisson's ratio varies little with hydrostatic pressure, we can substitute the ratio of the elastic moduli E/E_0 on the left of Eq. (3) or (4). In the event of elastic deformation of the surface asperities the stress values at different points of the contact zone are themselves different. Consequently, the hydrostatic components and the values of the moduli are different. Thus, the elastic contact problem is transformed into the problem with modulus of elasticity depending on the coordinate of the point. In this case, its values at the center of the contact area may be almost twice the value in the absence of hydrostatic pressure. We note that hydrostatic pressure is absent only in pure shear and, consequently, assuming a static determination of the modulus, only under those conditions will the value of the modulus not depend on the applied stress. In uniaxial tension the modulus will decrease with increase in stress, while in compression it will increase in conformity with the increase in hydrostatic tension or compression. For PVC, PMM, and nylon 6, the change in the moduli at the high-elastic limit may reach 20-30%.

The author and I. Yu. Maiors have experimentally confirmed this behavior of the moduli by measuring the dynamic modulus of elasticity under combined static tension and compression.

Hydrostatic pressure has a relatively weak effect on the deformation properties of metals because for metals the ratio $\alpha_k/\alpha_v K_0$ is three orders smaller than for polymers. However, at high homologous temperatures there is a very considerable increase in this ratio and therefore the effect of hydrostatic pressure may be substantial. Since for both brittle and dislocation fracture the strength of materials is proportional to the modulus of elasticity, in [15] it is assumed that the strength of polymer materials is also a function of specific volume only. Assuming that the temperature dependence of the strength or high-elastic limit is linear and using Relation (2), the following general expression is obtained for the strength in the presence of hydrostatic pressure:

$$\sigma_{eq} = \sigma_0 + \beta \left(\Delta T \pm \frac{\sigma_h}{\alpha_v K} \right), \quad (5)$$

where σ_0 is the ultimate strength at $T = T_0$ and $\sigma_h = 0$;

$$\beta = \left| \left(\frac{\partial \sigma_u}{\partial T} \right)_p \right|.$$

Here, the coefficient β depends on the strain rate or loading rate, and σ_{eq} is determined by the type of fracture for a given state of stress. For omnidirectional nonuniform compression and shear, the state of stress

typical of the friction process, it may be assumed that failure is always by shear, and Eq. (5) can be rewritten in the form

$$\sigma_i = \sigma_0 + \beta \left(\Delta T - \frac{\sigma_h}{\alpha_v K} \right). \quad (6)$$

We note that it follows from Eq. (5) that the ratio of the yield stresses in uniaxial tension and compression should be equal (for plastic materials) to

$$\frac{\sigma_{s \text{ com}}}{\sigma_{s \text{ ten}}} = \frac{1 + \frac{\beta}{3\alpha_v K}}{1 - \frac{\beta}{3\alpha_v K}}. \quad (7)$$

Relation (7) is confirmed by numerous experimental data.

It follows from Eq. (6) that in brittle fracture the shear strength should be greater than the tensile strength

$$\tau_u = \sigma_{\text{ten}} \left(1 + \frac{\beta}{3\alpha_v K} \right).$$

This relation is also confirmed by the experimental results.

It also follows from the data presented above that under friction conditions the strength of the adhesion bond or the resistance to deformation differs importantly from the resistance in pure or simple shear. Under conditions of dry or boundary friction the forces of friction are determined by the shear resistance [19].

It is well known that the relative velocity affects the coefficient of friction of a metal-polymer pair more strongly than that of a metal pair. This is not only because of the considerable temperature rise associated with the low thermal conductivity of polymer materials but also because the deformation of polymer materials in friction is elastoviscoplastic in character. Since even at low relative velocities the relative strain rate is large [19], a small change in velocity leads to an important change in the relation between the strain and relaxation rates. In the presence of considerable hydrostatic pressure the relaxation rate decreases and apparently corresponds to the equivalent temperature ($T \pm (\sigma_h/\alpha_v K)$). Thus, in friction the rate of the relaxation processes will also differ significantly from the rate in uniaxial deformation. Obviously, if in this case the time-temperature superposition principle [20] is employed, the shift factor a_T should be a function of the equivalent, and not the absolute temperature.

2. A direct method of experimental investigation, simulating contact effects under normal loads only, is the indentation method using a sphere, cone, or pyramid, since the resistance to indentation should be approximately equal to the deformation resistance of the surfaces. In [21] the hardness of various types of polymer materials was investigated by means of indentation tests using balls of various diameters and cones with various vertex angles. It was shown that in this case for all polymer materials the value of n in the formula $P = Ad^n$ lies on the interval $2.5 \leq n \leq 2.85$, and, as distinct from metals, there is no relation between this quantity and the modulus of hardening in uniaxial deformation. For materials such as PMM, as the deformation increases, the stress, after reaching the value σ_s , at first falls slightly, then for some time remains horizontal, and finally again increases. At the same time, according to Tabor's theory [22] at $n = 2.72$ there should be a monotonic increase in stress with strain. This, however, also follows from elementary considerations, since $n = 3$ corresponds to elastic strains.

The noncorrespondence of the data of uniaxial-compression and indentation tests is associated with the effect of hydrostatic pressure on the resistance to indentation.

It follows from simple geometric considerations [22] that for values of n on the above-mentioned interval, the Brinell hardness reaches a maximum at $d/D = 0.7-0.8$. Consequently, only at such ratios d/D does the hardness cease to depend on the applied load and become a characteristic of the material and not of the experimental conditions. In [22], it was shown that the hardness thus measured is closely correlated with the compressive strength. At the same time, it follows from these investigations that noticeable residual indentations appear at a value of the mean load σ_m close to the tensile strength of brittle materials. In this

TABLE 2

Material, σ_c , kgf/mm ²	d/D	$\frac{\sigma_m}{\sigma_c}$	Material, σ_c , kgf/mm ²	d/D	$\frac{\sigma_m}{\sigma_c}$
Plexiglas, 12.8	0.17	0.7	PVC, 8.8	0.18	0.9
	0.23	0.9		0.34	1.4
	0.44	1.5		0.65	1.5
	0.86	2.6		0.88	1.8
Teflon 1.4	0.25	1.5	K-21-22, 14.5	0.21	1.0
	0.59	2.2		0.35	1.5
	0.76	2.5		0.73	2.8

case the contact strength and hence the wear resistance depend importantly on the brittle strength. The limiting factor is evidently the stress at the edge of the indentation, where the material is in a state of pure shear. Thus, the contact strength of brittle polymer materials is low and the maximum specific load much less than that in simple compression.

In calculating the contact area of metallic materials it is usually assumed [1, 4] that the mean stress on the contact area is equal to $\sigma_m = c\sigma_s$, where for nonhardening materials the values of c is close to three. This follows from the data of numerous experiments and from the solution of the plane problem using the model of a rigid-plastic body.

For many polymer materials the situation is different. Here, even at $d/D = 0.4$, i.e., in the presence of very considerable deformation ($d/D = 0.4$ corresponds to an axial deformation at the center of the indentation of about 8%), σ_m is approximately equal to the compressive strength of yield stress and even at $d/D = 0.8$, the mean stress is only 2-3 times greater than the yield stress (Table 2). We note that under actual polymer friction conditions the deformation of the asperities is less than 8%. Thus, the mean stress on the contact areas will be less than the strength or high-elastic limit in uniaxial compression and, as distinct from metals, will depend importantly on the degree of deformation. This somewhat unexpected result is associated with the fact that the Young's modulus of polymer materials is small, and even at considerable strains the contribution of the elastic deformation is large, so that the use of the rigid-plastic model is quite inappropriate.

Hill's solution [23] for an elastoplastic sphere expanded by internal pressure shows that the relative limiting stress σ_{lim}/σ_s is proportional to E/σ_s . For polymers the ratio E/σ_s is one or two orders smaller than for metals and the relative limiting stress and the ratio of the hardness to the yield stress are correspondingly smaller. At the same time, as a result of the high degree of actual hardening there should be very little change in the contact area in shear [4].

In [21], in addition to the short-time hardness, creep and recovery under contact loading conditions were also investigated. In both cases the depth of the indentation proved to depend linearly on the logarithm of time: $h = h_0 + a \ln t/t_0$. The coefficient a , which determines the creep rate, is an increasing function of the initial load. Under dry friction conditions, creep of the contact areas determines the difference between the static and kinetic coefficients of friction. For all the polymer materials investigated, the contact creep rate was found to be quite small and in the friction process the creep effect is significant only at very small relative velocities.

3. The effect of the mechanical properties of polymer materials on friction processes may conveniently be investigated using various kinds of fillers. In this case, the mechanical properties of the material can be strongly modified without seriously affecting its adhesion characteristics, since the adhesion interaction with the opposing friction surface is principally determined by the properties of the starting material.

The effect of various fillers differing in chemical nature (oxides, silicates, metals) and particle size on the properties of high-pressure and low-pressure polyethylene is described in [24, 25]. The thermo-physical, mechanical, and friction properties of the compositions were investigated. It is shown that the coefficient of friction is the lower, the greater the Young's modulus, which, strictly speaking, was to be expected, since in view of the elastoplastic nature of the deformation the contact area is correspondingly smaller at a given load. At the same time, the wear is not directly related to the mechanical properties and, to a considerable extent, is determined by the particle size (the optimal particle diameter is quite large, namely, 0.3-0.5 mm) and by the nature of the adhesion interaction of the particles in the filler. The coeffi-

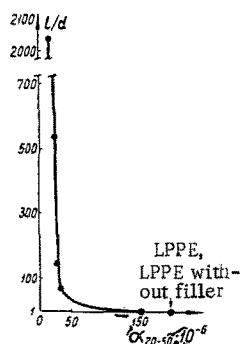


Fig. 3

Fig. 3. Coefficient of linear expansion as a function of the relative length of the fibers. Composition: LPPE + fiberglass (30 wt. %).

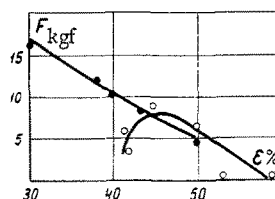


Fig. 4

Fig. 4. Strength of monolithic (●) and welded (○) specimens of K-17-2 material as a function of the degree of deformation.

cient of linear expansion of the composition depends very importantly on the shape of the filler particles. If the latter are spherical, the overall coefficient of linear expansion can be calculated from the rules of additivity and in the best case the change may be of the order of tens of percent. When fibrous fillers, for example, glass fibers, are employed, the coefficient of linear expansion may be reduced by a factor of 5-10, if the relative length of the fibers is sufficient (Fig. 3). The same effect can be obtained with fibers of other materials with a low coefficient of linear expansion and a sufficiently large value of the Young's modulus. We note that in this way, the coefficient of linear expansion of polymer materials can be reduced below that of metallic antifriction materials, i.e., adequate dimensional thermal stability can be obtained. The effectiveness of the fibers is due to the fact that, as distinct from powdered fillers, they not only expand very little themselves, but also impede the expansion of the basic material. We note that whenever relatively hard fillers are used, the wear of both low-pressure and high-pressure polyethylene is reduced by several orders, i.e., becomes much less than the wear of any homogeneous polymer material or antifriction alloy.

4. As already noted, in dry and boundary friction the force of friction is determined by the contact area and by the adhesion interaction. In polymer-polymer friction adhesion bonds due to both van der Waals forces and covalent bonds may develop. In [62, 67], it was shown that the simultaneous plastic deformation of both thermosets and amorphous thermoplastics leads to quite intense adhesion interaction in the contact zones even at room temperature, if the degree of deformation is sufficient, and the strength of the bond thus formed is close to that of a monolithic piece of polymer with the same degree of deformation (Fig. 4). In [28] it was shown that a strong adhesion bond can also be produced by the simultaneous deformation of specimens of a crystalline polymer material (HPPE) at room temperature. Thus, despite various statements to the contrary [29, 30], adhesion bonds do develop in the contact zones at room temperature. Under ordinary conditions, these bonds are difficult to observe, since, owing to the large reserve of elastic energy, they break when the load is removed [28].

A study of the adhesion of polyethylene at various strains and temperatures is reported in [31]. It is shown that the strength of the joint has a maximum value at a certain degree of combined deformation and in a number of cases falls with rise in temperature, which would scarcely be the case if the adhesion phenomenon were diffusional in nature. So far, the question of metal-polymer adhesion has been studied by forming polymer films on the metal and subsequently destroying the bond; this method does not satisfactorily reproduce the conditions that exist in friction, since the conditions of creation of the adhesion bond are quite different.

Attempts to calculate the strength of the steel-polyethylene bond theoretically [32, 33] have yielded rather considerable values. In [43] it was shown that at high specific load seizing is possible even in polymer friction. This also points to very considerable adhesion forces.

One of the important problems of the theory of polymer adhesion is the kinetics of formation of adhesion bonds. There have been studies [33, 35] in which the diffusional origin of the polymer adhesion bond was demonstrated, since, in the opinion of the authors, it follows from the dependence of the adhesion strength on time.

Most investigators concerned with polymer friction assume (and this is confirmed by numerous observations) that the force of friction is basically determined by the effort expended in destroying the adhesion bonds that develop in the contact zones. If adhesion arose as a result of diffusion processes developing at a quite low rate, then at a certain sliding velocity there would be a sharp fall in the forces of friction. In fact, for polymer materials the coefficient of friction usually increases slightly as the velocity rises, and the forces of friction fall at velocities at which the surface layer of the rubbing part melts as a result of the intense heat release. We note that in the process of friction at velocities of the order of tenths of a meter per second, the contact time does not exceed thousandths of a second.

Thus, as a working hypothesis, it is preferable to assume that the forces of adhesion are determined only by the distance between the interacting bodies, especially when the nondirectional van der Waals forces play a decisive role, and that the possible dependence of the forces of friction on contact time is determined by the change in contact area owing to the viscoelastoplastic nature of the deformation of polymer materials.

5. Metal-polymer and polymer-polymer friction pairs can operate without a lubricant only at low specific loads and low velocities. However, in choosing lubricants for polymer materials their special characteristics should be taken into account. Investigations have shown that surface-active lubricants, very effective in boundary friction, lead to increased wear of the polymer element of the friction pair. When inert lubricants are used, it becomes necessary to determine the complex of rheological characteristics that will ensure the maximum load-bearing capacity under boundary friction conditions. In [36], it was shown that if the lubricant has a finite yield stress (at a given strain rate), then under normal load only the load-bearing capacity of the lubricant increases without bound with increase in the ratio of the characteristic contact dimension to the thickness of the film. Thus, if only normal loads are acting, such a lubricant cannot be extruded from the contact zone. Under stationary contact conditions the subsequent decrease in film thickness is attributable to processes analogous to creep, which evidently determines the difference between the static and kinetic coefficients of friction [37]. In the presence of a normal load and tangential forces there is a further decrease in the thickness of the film as a result of the lowering of the effective yield point. In this case, however, the important characteristics of the lubricant are its viscosity and the dependence of the viscosity on temperature and pressure. We note that the assumption that the specific volume is the primary influence on the rheological properties apparently also holds true for lubricants [38]. Accordingly, if, for a given lubricant, the temperature dependence of viscosity is strong, the dependence of the viscosity on pressure will be correspondingly important. At low velocities, when temperature effects are not very great, the increase in viscosity with pressure leads to an increase in the load-bearing capacity of the lubricant. At small distances between the rubbing surfaces, as a result of the effect of the surface force field, there may be a change in the mechanical properties of the lubricant with distance from the surfaces of the rubbing parts. In this case, as shown in [39], lubricants will have a finite load-bearing capacity with low shear resistance.

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