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DEFORMATION AND FAILURE OF ALUMINUM ALLOYS FROM THE STANDPOINT OF THE KINETIC CONCEPT OF STRENGTH

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A thermal-activation analysis was performed of experimental data on the strain and failure of 1201 T1, D16 T*, and* AK4-1 T1 *aluminum alloys. The experiments were conducted under constant loads in creep conditions and under increasing loads. The duration of the tests was varied from fractions of a second to ten thousand hours, and the temperature ranged from* 77 *to* 473 K*. The rate activation parameters in the equations of steady-state creep and plastic strain were determined. Information was obtained on the relationship between plastic strain and failure. The plastic strain rate is shown to be affected by relaxation phenomena. The plastic characteristics of the alloys and their dependences on the temperature and time to failure are given.*

Key words: *failure, aluminum alloys, strength, life, creep.*

Introduction. The interaction and relationship between strain and failure is a complex and poorly studied problem of the theory of strength of solids. The internal processes occurring during deformation and failure of materials are rather complicated and diverse. Insufficient knowledge of the relation between these processes and the macroscopic characteristics of solids subjected to loading has led to a great differentiation of studies in this area. As a result, the problem of estimating the serviceability of materials has been divided into independent parts according to the nature of loading and service conditions (short-term and long-term strength, plasticity and creep, fatigue, thermocyclic strength, etc.). At the same time, the accumulation of a great body of experimental data on the behavior of many materials (polymers, metals, and alloys) under different loading and temperature conditions has made it possible to take a new view of the failure and deformation of solids under loading using a unified approach.

The basis of this approach was the general physical concept on the participation of the thermal motion of atoms in failure and deformation. In the approach, an important part is assigned to the nonregularity of thermal motion, i.e., occurrence of thermal fluctuations. This approach, now called the kinetic concept of failure, was first formulated in the early 1950 by S. N. Zhurkov and is consistently set forth in a number of papers, for example in [1–4]. In the kinetic approach, the plastic deformation of crystalline materials is treated as thermally activated motion of dislocations and other defects under the activating effect of stresses [2, 3, 5]. This process may not be accompanied by occurrence of defects whose dimensions are such that the interatomic-interaction forces in the volume of the defect disappear. In this case, interatomic bonds are disturbed only temporarily, and after termination of the deformation, they restore again, partly or completely. Failure is treated as an irreversible thermally activated process of accumulation of defects (submicrocracks and microcracks, pores) in time, whose further growth results in the occurrence of signs of macrofailure — macrocracks.

In loaded materials, the following two processes occur simultaneously: motion of groups of atoms relative to each other and the development of points of fracture. The kinetics and mechanisms of these processes are different. For different loading conditions, it is sometimes possible to establish the basic mechanism of this or that process and, sometimes, to reveal their interaction [2].

The present paper reports results from studies of the strain–failure relation for structural aluminum alloys based on the kinetic approach. The present investigation is a continuation of previous studies of the strength and

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life of 1201 T1, D16 T, and AK4-1 T1 aluminum alloys used in aircraft engineering [6]. All data obtained in the present study and the experimental data on the strain and failure of these alloys used here refer to uniaxial tension of samples by constant and monotonically increasing loads under different time–temperature conditions.

Relationship between Life and Plastic Strain Rate. The kinetic approach [1–4] considers two different simultaneous physical processes in a loaded body — deformation and failure. Each process has its own rate. Macroscopic deformation and the origin of macrocracks are the results of these two processes.

Experiments with many materials [1–3] revealed that the dependence of the time to failure τ (life) on the applied stress σ and the absolute temperature T is given by the formula

$$\tau = \tau_0 \exp\left((U_0 - \gamma\sigma)/(RT)\right). \tag{1}$$

Here τ_0, U_0, and γ are coefficients and R is the universal gas constant. The reverse of the life, i.e., the average rate of failure $\dot{\omega}_m$, is defined by the equation

$$\dot{\omega}_m = 1/\tau = \nu_0 \exp\left(-(U_0 - \gamma\sigma)/(RT)\right), \tag{2}$$

where $\nu_0 = 1/\tau_0$ is the frequency factor.

The dependence of the steady-state creep rate $\dot{\varepsilon}_p$ on the applied stress σ is given by

$$\dot{\varepsilon}_p = \dot{\varepsilon}_{0p} \exp\left(-(Q_0 - \alpha\sigma)/(RT)\right). \tag{3}$$

The current value of the plastic strain rate of a material $\dot{\varepsilon}$ is defined as

$$\dot{\varepsilon} = \dot{\varepsilon}_0 \exp\left(-(Q_0 - \alpha\sigma)/(RT)\right). \tag{4}$$

In Eqs. (3) and (4), $\dot{\varepsilon}_{0p}$, $\dot{\varepsilon}_0$, Q_0, and α are coefficients.

For time-varying stresses $\sigma(t)$, one commonly use the principle of linear summation of damage and, converting to the rate of failure $\dot{\omega}$ for constant U_0 and γ, the time to failure τ is determined from the relation

$$\int_0^\tau \dot{\omega}(\sigma, T, t)\, dt = \int_0^\tau \nu_0 \exp\Big(-\frac{U_0 - \gamma\sigma(t)}{RT}\Big)\, dt = 1. \tag{5}$$

The coefficients τ_0, ν_0, U_0, $\dot{\varepsilon}_{0p}$, $\dot{\varepsilon}_0$, Q_0, γ, and α are assigned the role of physical constants and parameters. As regards the physical meaning of these coefficients, the following conclusions were made [1, 2, 4]:

1. The coefficient τ_0 is in the range of 10^{-11}–10^{-14} sec, which almost coincides with the period of thermal vibrations of atoms in solids; accordingly, $\nu_0 = 1/\tau_0$ sec^{-1} characterizes the vibration frequency. The coefficients $\dot{\varepsilon}_{0p}$ and $\dot{\varepsilon}_0$ are close to the frequency factor ν_0 but their physical meaning requires a more detailed interpretation.

2. The coefficients U_0 and Q_0 have the dimension of energy (KJ/mole); U_0 is related to the break energy of atomic–molecular bonds are called the initial energy of failure activation, and Q_0 is called the initial energy of deformation activation. Both coefficients depend weakly on material structure.

3. The coefficients γ and α are structurally sensitive parameters. The coefficient γ is called the failure activation volume and is defined as the product of atomic volume by the overstrain coefficient at sites of local failure. The coefficient α is related to the deformation activation volume but its physical meaning has not been completely clarified.

The identical structure of Eqs. (2)–(4), defining the rate of the failure and deformation processes, indicates that there may be a relationship between the life and plastic strain rate. To elucidate this relationship, it is necessary to determine and compare the rate activation parameters (U_0, γ and Q_0, α) of both processes.

Let us consider the relationship between life and steady-state creep rate using as an example the processing of experimental data obtained in tests of 1201 T1 samples. Information on the samples used, equipment, and experimental procedure is given in [6]. The data on the life of samples of this alloy obtained in [6] were supplemented by data on creep under constant loads and temperatures of 398 K and 433 K. During the tests of the samples, creep strains were measured in definite time intervals. Five samples were tested under each temperature–loading conditions. Separately, for each sample, a creep curve was plotted, from which (following the standard procedure), the steady-state creep rate was determined, and the average rate was then obtained for five samples under the given test conditions. In addition, after failure of all samples tested in the present study and in [6], we determined the residual elongation $\delta = (l - l_0)/l_0$ and the true strain in the neck $\varepsilon_* = \ln(F_0/F)$, where l_0 and l, F_0 and F are the length and cross sectional area of a sample before and after the test, respectively. The test conditions and average values of experimentally determined parameters for 1201 T1, D16 T, and AK4-1 T1 alloys are given in Tables 1–3.

TABLE 1

Results of Testing 1201 T1 Aluminum Alloy under Constant Loads

T, K	σ_0, MPa	τ, sec	$\dot{\varepsilon}_p$, sec^{-1}	δ	ε_*	ε_p
398	300	$5.76 \cdot 10^5$	$8.80 \cdot 10^{-9}$	0.061	0.272	$5.07 \cdot 10^{-3}$
	280	$4.83 \cdot 10^6$	$2.75 \cdot 10^{-9}$	0.048	0.153	$1.33 \cdot 10^{-2}$
	260	$2.36 \cdot 10^7$	$1.29 \cdot 10^{-9}$	0.060	0.129	$3.04 \cdot 10^{-2}$
	240	$1.8 \cdot 10^{6\,*}$	$5.97 \cdot 10^{-10}$	—	—	—
423	240	$1.44 \cdot 10^7$	—	0.04	0.085	—
433	280	$2.72 \cdot 10^4$	—	0.072	0.478	—
	260	$1.22 \cdot 10^5$	—	0.067	0.448	—
	250	$4.02 \cdot 10^5$	—	0.070	0.375	—
	240	$1.46 \cdot 10^6$	$1.66 \cdot 10^{-8}$	0.064	0.260	$2.42 \cdot 10^{-2}$
	220	$1.44 \cdot 10^7$	$2.39 \cdot 10^{-9}$	0.046	0.108	$3.44 \cdot 10^{-2}$
	200	$1.8 \cdot 10^{6\,*}$	$1.02 \cdot 10^{-9}$	—	—	—
	180	$1.8 \cdot 10^{6\,*}$	$3.09 \cdot 10^{-10}$	—	—	—
	160	$1.8 \cdot 10^{6\,*}$	$3.42 \cdot 10^{-10}$	—	—	—
448	200	$1.52 \cdot 10^7$	—	0.042	0.095	—
	180	$4.06 \cdot 10^7$	—	0.035	0.070	—
473	160	$1.83 \cdot 10^7$	—	0.036	0.078	—
	140	$4.26 \cdot 10^7$	—	0.036	0.042	—

Note. Superscript asterisk denotes that the tests were interrupted.

TABLE 2

Results of Testing D16 T and AK4-1 T1 Aluminum Alloys under Constant Loads

T, K	D16 T				AK4-1 T1			
	σ_0, MPa	τ, sec	δ	ε_*	σ_0, MPa	τ, sec	δ	ε_*
423	240	$8.78 \cdot 10^6$	0.048	0.693	280	$1.51 \cdot 10^6$	0.039	0.505
448	240	$4.23 \cdot 10^5$	0.065	0.650	240	$8.57 \cdot 10^5$	0.058	0.501
	200	$2.14 \cdot 10^6$	0.082	0.673	—	—	—	—
	160	$1.35 \cdot 10^7$	0.068	0.718	—	—	—	—
	140	$1.66 \cdot 10^7$	0.061	0.866	—	—	—	—
473	—	—	—	—	160	$1.25 \cdot 10^6$	0.079	0.810

For tests under constant loads, Tables 1 and 2 give the initial values of the stresses σ_0. For tests under increasing loads, Table 3 gives, instead of the time to failure τ, the time of loading t_* to the maximum stress σ_* at the strain localization that led to macrofailure of the sample.

The method of processing experimental data for determining the activation parameters Q_0 and α in Eq. (3) is similar to the method for determining U_0 and γ described in [6]. Setting $\sigma \approx \sigma_0 = \text{const}$ in the creep conditions under constant loads, we write relation (3) as

$$Q(\sigma) = Q_0 - \alpha\sigma = RT \ln(\dot{\varepsilon}_{0p}/\dot{\varepsilon}_p). \tag{6}$$

The coefficient $\dot{\varepsilon}_{0p}$ is determined by processing creep data for 1201 T1 alloy, taking into account that the duration of the steady-state creep stage should constitute most of the time to failure. This condition corresponds to the following regimes: 1) $T = 398$ K and $\sigma_0 = 260$ MPa; 2) $T = 433$ K and $\sigma_0 = 240$ and 220 MPa (see Table 1). Setting $\dot{\varepsilon}_{0p} = \varepsilon_p \nu_0$, where $\varepsilon_p = \dot{\varepsilon}_p \tau$ is the strain accumulated at the steady-state creep rate during sample failure, we find the average value $\varepsilon_p \approx 3 \cdot 10^{-2}$. Then, at $\tau_0 = 10^{-13}$ sec [6] and, accordingly, $\nu_0 = 10^{13}$ sec^{-1}, we obtain $\dot{\varepsilon}_{0p} = 3 \cdot 10^{11}$ sec^{-1}.

Equation (6) allows one to plot steady-state creep rates measured at different stresses and temperatures. Using experimental values of $\dot{\varepsilon}_p$ (Table 1) for corresponding σ_0 and T and known $\dot{\varepsilon}_{0p}$, we calculate $RT \ln(\dot{\varepsilon}_{0p}/\dot{\varepsilon}_p)$

TABLE 3

Results of Testing Aluminum Alloys under Increasing Loads

T, K	D16 T				AK4-1 T1				1201 T1			
	σ_*, MPa	t_*, sec	δ	ε_*	σ_*, MPa	t_*, sec	δ	ε_*	σ_*, MPa	t_*, sec	δ	ε_*
77	812	45.0	0.133	0.145	741	30.0	0.089	0.405	709	37.0	0.093	0.240
123	747	34.0	0.085	0.181	699	52.0	0.089	0.439	—	—	—	—
223	694	52.0	0.140	0.198	—	—	—	—	—	—	—	—
293	763	$1.53 \cdot 10^{-3}$	0.153	0.274	666	21.0	0.103	0.452	567	$2.32 \cdot 10^{2}$	0.091	0.229
	706	$1.13 \cdot 10^{-1}$	0.141	0.223	682	$3.01 \cdot 10^{4}$	0.101	0.455	563	$9.30 \cdot 10^{6}$	0.093	0.277
	658	62.0	0.133	0.189	702	$7.25 \cdot 10^{6}$	0.118	0.447	545	$4.62 \cdot 10^{7}$	—	0.272
	644	$4.53 \cdot 10^{7}$	0.127	0.158	698	$3.71 \cdot 10^{7}$	0.099	0.458	—	—	—	—
373	722	$1.40 \cdot 10^{-3}$	0.140	0.275	—	—	—	—	598	$2.04 \cdot 10^{2}$	0.101	0.311
	651	$9.60 \cdot 10^{-2}$	0.120	0.211	657	$2.76 \cdot 10^{4}$	0.115	0.485	—	—	—	—
	634	39.0	0.118	0.198	621	$1.05 \cdot 10^{6}$	0.110	0.486	—	—	—	—
398	—	—	—	—	—	—	—	—	576	$2.12 \cdot 10^{2}$	0.100	0.370
423	662	$1.29 \cdot 10^{-3}$	0.129	0.223	—	—	—	—	—	—	—	—
	652	$9.0 \cdot 10^{-2}$	0.113	0.233	—	—	—	—	—	—	—	—
	610	41.0	0.122	0.236	569	$8.49 \cdot 10^{5}$	0.087	0.592	—	—	—	—
448	—	—	—	—	—	—	—	—	521	$1.94 \cdot 10^{2}$	0.103	0.431
473	686	$9.80 \cdot 10^{-4}$	0.098	0.328	—	—	—	—	454	$1.78 \cdot 10^{2}$	0.111	0.576
	612	$8.40 \cdot 10^{-2}$	0.105	0.298	—	—	—	—	—	—	—	—
	592	38.0	0.114	0.296	483	$7.03 \cdot 10^{5}$	0.090	0.962	—	—	—	—

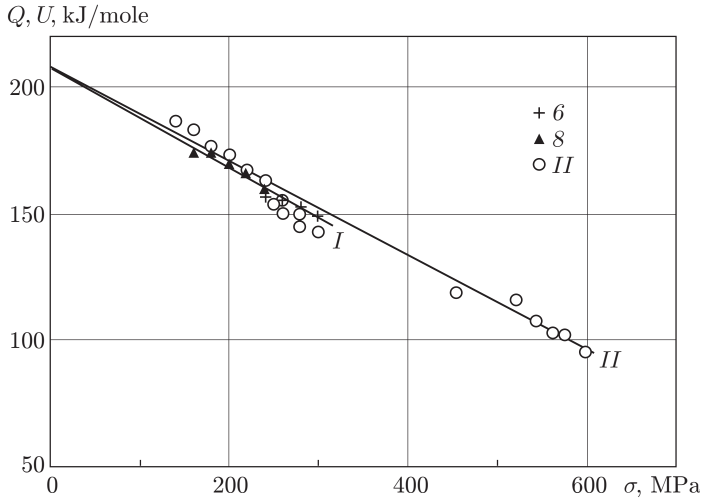

Fig. 1

from (6). From these data, we plot a curve of Q versus σ. If the points corresponding to different values of $\dot{\varepsilon}_p$, T, and σ are grouped along the curve $Q(\sigma) = Q_0 - \alpha\sigma$, this indicates that Eq. (3) is valid and the parameter α is constant. The curve I in Fig. 1 shows the dependence obtained by least-squares processing of the experimental data. The point of intersection of this curve with the ordinate and the slope determine the values of Q_0 and α, respectively. In this figure and in Figs. 2, 4, and 5, digits indicate points that correspond to the following temperatures: 77 (1), 123 (2), 223 (3), 293 (4), 373 (5), 398 (6), 423 (7), 433 (8), 448 (9), and 473 K (10).

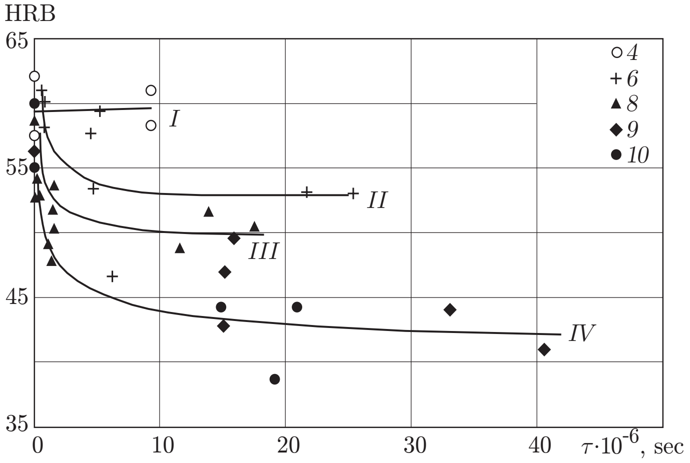

Fig. 2

To illustrate the relationship between steady-state creep and failure rates, Fig. 1 gives data from [6] on the force dependence of the failure activation energy $U(\sigma) = U_0 - \gamma\sigma$ on loading for those test conditions under which $\gamma = \gamma_{\min}$. The curve II corresponding to this dependence encompasses the range of temperatures of 293–473 K. The similarity of the dependences $Q(\sigma)$ and $U(\sigma)$ indicates that the rate activation parameters of both processes are equal: $U_0 \approx Q_0$ and $\gamma_{\min} \approx \alpha_{\min}$. The value $Q_0 = 208$ kJ/mole coincides with U_0 within three significant digits [6]. The coefficient $\alpha_{\min}$ is equal to 0.198 at $\gamma_{\min} = 0.188$ kJ/(mole·MPa) [6]. The difference between $\alpha_{\min}$ and $\gamma_{\min}$ does not exceed the usual errors in determining the activation volume γ (±10% [1]). The indicated experimental dependences suggest an intimate relationship between the two processes.

If the conditions $U_0 = Q_0$ and $\gamma = \alpha$ are satisfied, Eqs. (1) and (3) imply that the product $\dot{\varepsilon}_p\tau = \dot{\varepsilon}_{0p}\tau_0 = \text{const}$, and, hence, the steady-state creep strain ε_p accumulated during the time of sample failure in the case where the duration of the steady-state stage constitutes most of this time, should be approximately constant. The similarity of U_0 and Q_0 and γ and α indicates that there is a relationship between deformation and failure processes at the microlevel [2, 3]. However, at the macrolevel, the constancy of $\dot{\varepsilon}_p\tau$, illustrating this relationship, can be disrupted. There are several reasons for this. Plastic deformation, on the one hand, prepares failure; on the other hand, it promotes relaxation processes [2, 3]. In addition, the action of stresses and plastic strain accelerate the decomposition of the supersaturated solid metal solution in dispersionally hardening aluminum alloys [7].

For 1201 alloy in the T1 state, the temperature interval in which the tests were performed is the region of strain aging. Therefore, the time of exposure of the alloy to the test temperature can have an influence on its strength and strain properties. This influence is enhanced under the activating effect of stresses. The aging of the alloy is accompanied by a change in its strength, which suggests a change in the properties of the alloy.

We studied the effect of the exposure time of 1201 T1 alloy samples on their hardness on simultaneous exposure to temperature and loading. The Rockwell hardness (the B scale) was measured after rupture of the sample at the point of failure. Variation in the hardness of 1201 T1 alloy versus test time and temperature is shown in Fig. 2. Curves I, II, and III correspond to temperatures of 293 K, 398 K, and 433 K, respectively, and curve IV corresponds to temperatures of 448–473 K. As can be seen from the figure, the loading of the alloy at room temperature did not lead to noticeable changes in its hardness. However, at elevated temperatures, the hardness decreased with time, and the higher the temperature, the larger the decrease. However, with increase in the exposure time, the hardness of the alloy practically ceased to change. Therefore, it is assumed that in the region in which the steady-state creep rate was determined, the properties of the alloy changed only slightly.

For D16 T and AK4-1 T1 alloys, the activation parameters Q_0 and $\alpha_{\min}$ were determined from the average plastic strain rate $\dot{\varepsilon}_m$ obtained by processing the experimental data on the creep of these alloys under constant load given in Table 2. The average plastic strain rate was calculated by dividing the residual elongation at fracture δ by the time to failure τ. The data processing technique for determining the activation parameters Q_0 and $\alpha_{\min}$ in Eq. (4) is similar to the method of determining these parameters in Eq. (3). Provided that $\dot{\varepsilon} = \dot{\varepsilon}_m$ and $\sigma \approx \sigma_0 = \text{const}$, dependence (4) is written as

TABLE 4

Comparison of Activation Parameters of Deformation and Failure of Aluminum Alloys

Alloy grade	Q_0, $\frac{\text{kJ}}{\text{mole}}$	$\alpha_{\min}$, $\frac{\text{kJ}}{\text{mole}\cdot\text{MPa}}$	$\alpha_{\max}$, $\frac{\text{kJ}}{\text{mole}\cdot\text{MPa}}$	U_0, $\frac{\text{kJ}}{\text{mole}}$	$\gamma_{\min}$, $\frac{\text{kJ}}{\text{mole}\cdot\text{MPa}}$	$\gamma_{\max}$, $\frac{\text{kJ}}{\text{mole}\cdot\text{MPa}}$
1201 T1	208	0.198	0.268	208	0.188	0.264
D16 T	194	0.132	0.206	193	0.135	0.203
AK4-1 T1	194	0.126	0.237	193	0.126	0.234

$$Q(\sigma) = Q_0 - \alpha\sigma = RT \ln(\dot{\varepsilon}_0/\dot{\varepsilon}_m), \tag{7}$$

where, according to [1], the coefficient $\dot{\varepsilon}_0$ is set equal to 10^{12} sec^{-1}, which also corresponds to the experimental data obtained.

For each alloy, curves of $RT\ln(\dot{\varepsilon}_0/\dot{\varepsilon}_m)$ versus σ, having the shape of straight lines of the form (7), were plotted using the measured values of $\dot{\varepsilon}_m = \delta/\tau$ (see Table 2), and the parameters Q_0 and $\alpha = \alpha_{\min}$ were then determined. The values of Q_0 and $\alpha_{\min}$ obtained for D16 T and AK4-1 T1 alloys and the values of U_0 and $\gamma_{\min}$ from [6] are given in Table 4. A comparison of these data shows that the deformation and failure activation parameters for D16 T and AK4-1 T1 alloys are similar in magnitude, as well as the parameters of 1201 T1 alloy in the case where Q_0 and $\alpha_{\min}$ were determined from steady-state creep rates.

As shown by studies [6] of the failure of aluminum alloys, the structurally sensitive parameter γ can change and differ in magnitude from $\gamma_{\min}$, depending on test conditions. The activation volume γ varied in the range from $\gamma_{\max}$ to $\gamma_{\min}$, which was explained by the influence of local stress relaxation. Under certain test conditions (elevated temperature, long-term loading or low level of actual stresses), relaxation processes occur primarily in the initial period of failure of the material; therefore, most of the failure process proceeds for the minimum value of the local stresses in the material and, hence, for the minimum value of the activation volume γ. The maximum value $\gamma_{\max}$ of the activation volume was reached in those cases where relaxation of local stresses was nearly absent (cryogenic temperature, shock loading). The strain characteristics of the alloys (see Table 3) were also measured in these experiments.

Relaxation processes can also lead to variation in the parameter values of dependences (3) and (4) [2]. Using the experimental data given in Table 3, the value of $\alpha_{\max}$ can be estimated under the same test conditions under which $\gamma_{\max}$ was determined. Substituting the variable $\sigma(t)$ for the variable σ in the plastic strain rate equation (4), we integrate it, assuming that the residual elongation δ is the plastic strain corresponding to the moment of macrofailure:

$$\int_0^\tau \dot{\varepsilon}(\sigma, T, t)\, dt = \int_0^\tau \dot{\varepsilon}_0 \exp\Big(-\frac{Q_0 - \alpha\sigma(t)}{RT}\Big)\, dt = \delta. \tag{8}$$

Let the stresses increase at constant rate w until rupture of the sample under stress σ_* at time t_*. Substituting $\tau = t_*$ and $\sigma(t) = wt$ and $(w = \sigma_*/t_*)$ in (8), we obtain

$$B\sigma_* = -\ln(At_*(1 - \exp(-B\sigma_*))/(B\sigma_*)), \tag{9}$$

where $A = \dot{\varepsilon}_0 \exp(-Q_0/(RT))/\delta$ and $B = \alpha/(RT)$. Here Q_0 and $\dot{\varepsilon}_0$ have the same values as in the tests with constant loads. The parameter $\alpha = \alpha_{\max}$ is calculated from the values of σ_*, t_*, T, and δ (Table 3) by (9) using the method of successive approximations. The averaged values of $\alpha_{\max}$ for the three tested alloys are given in Table 4. As can be seen from a comparison of $\alpha_{\max}$ and $\alpha_{\min}$, the relaxation processes has a significant effect on the plastic strain rate. For comparison, Table 4 includes the values of the activation volume $\gamma_{\max}$ from [6] for the same alloys.

The close values of the deformation activation parameters Q_0, $\alpha_{\max}$, and $\alpha_{\min}$ and the corresponding failure parameters U_0, $\gamma_{\max}$, and $\gamma_{\min}$ indicate that the two processes are interrelated and exhibit identical sensitivity to the evolution of the material structure. All three alloy grades have identical dependences $U(\sigma)$ and $Q(\sigma)$ in spite of the fact that these alloys differ in elemental composition and thermal treatment. At the same time, as is evident from Table 1–3, the residual strain ε_* undergoes the most significant changes with variation in the test conditions of each alloy.

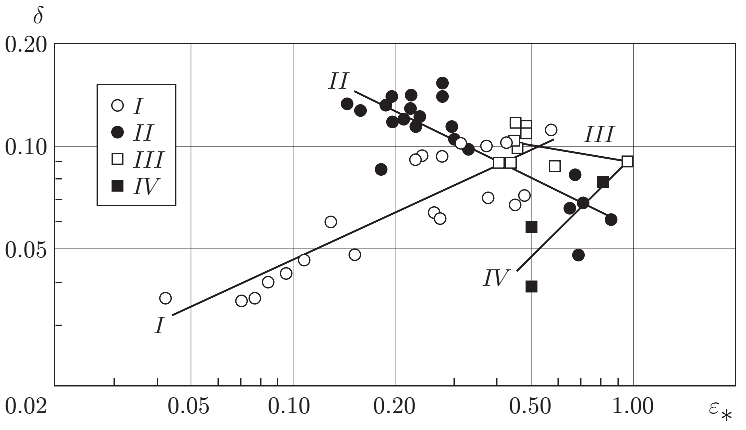

Fig. 3

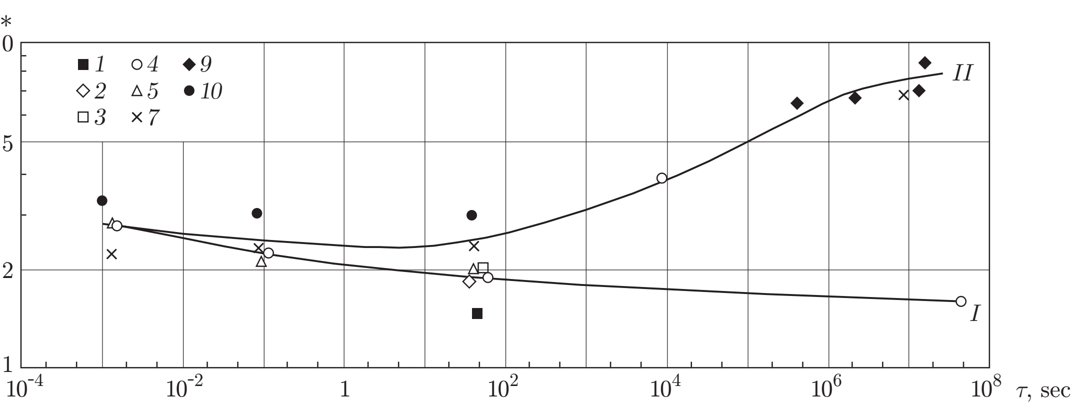

Fig. 4

Plastic Strain of 1201 T1, D16 T, and AK4-1 T1 Alloys. The relation between ε_p, δ, and ε_* depends on the grade and thermal treatment of an alloy (which determine features of the aging process in the alloy) and on how this process is related to failure and deformation. It should be noted that failure is a process of local scale but it occurs simultaneously in many local volumes. The life determined in experiments refers, as a rule, to one of these volumes. Residual plastic strain is the total result of the local processes occurring in places of failure localization and in other volumes. Plastic strain is nonuniformly distributed over the material volume. Residual elongation is a plastic flow characteristic averaged over the sample volume. It is largely contributed by the strain of those material volumes in which the failure process is at different stages of development.

The local strain in the sample neck ε_* is related to the point of failure. It would seem that the strain localization in the neck should make a considerable contribution to the residual elongation of the sample. However, the process of alloy aging at elevated temperature under pressure gives different effects, depending on the alloy composition and structure. Figure 3 shows statistical relations between the residual elongation δ and the local strain ε_* in the sample neck plotted from the data of Table 1–3 using the least-squares method. For 1201 T1 alloy, the correlation between them is positive (curve I), and for D16 T alloy, it is negative (curve II). For AK4-1 T1 alloy, the correlation between δ and ε_* depends on the combinations of temperature and stresses at which failure occurred. This alloy undergoes phase aging up to the second hardness peak [8]. If failure proceeds at higher stresses or higher temperature, there is a weak negative correlation between δ and ε_* (curve III, increasing loads). At lower stresses in the same range of temperatures, both parameters have smaller values. As a result, a positive correlation between them is obtained, in our case, in a rather small interval of life (curve V, constant loads).

Features of the alloy composition and structure have an effect on the aging process and the nature of the dependence of the residual strain on the temperature and loading time. Figure 4 shows curves of the residual

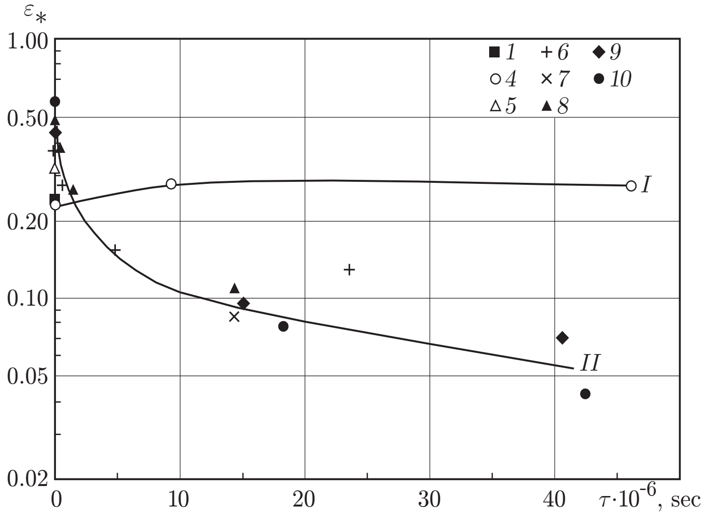

Fig. 5

strain ε_* in the neck of D16 T alloy samples versus temperature and time to failure. Curve I corresponds to a temperature of 293 K and curve II to the temperature range of 423–473 K. We note that in the temperature range of 77–293 K, the residual strain ε_* changed insignificantly, whereas the duration of the tests was varied from one-thousandth of a second to ten thousand hours. Different trends of ε_* are observed in the same time interval at higher temperatures. In this case, D16 T alloy has a plasticity minimum, different in magnitude and time for different temperatures. For AK4-1 T1 alloy, the plasticity value gained during artificial aging at a temperature of (468 ± 5) K remains unchanged [8] if the test temperature was lower than this value. At higher temperatures, the residual strain ε_* increases (see Tables 2 and 3)

D16 T and AK4-1 T1 belong to the Al–Cu–Mg system, whereas 1201 T1 alloy belong to the Al–Cu–Mn system. In addition, the copper content in the latter is higher than the amount completely dissolved under heating for quenching [8]. All these factors influence the course of the aging process of this alloy. Figure 5 gives curves of ε_* versus temperature and time to failure for 1201 T1 alloy. Curve I corresponds to a temperature of 293 K, and curve II corresponds to temperatures of 398–473 K. At room temperature, ε_* remains unchanged during both short-term and long-term failure. In short-term standard fracture tests of samples under heating, ε_* increases with a rise in temperature. However, an increase in the time of tests at elevated temperatures leads to a considerable decrease in the residual strain ε_*. Such a decrease in alloy plasticity under long-term loading at elevated temperature enhances the failure hazard of structural members with stress concentrators, which should be taken into account in designing.

The above experimental data on the strain and failure of three grades of structural aluminum alloys and thermal activation analysis data were used to determine the kinetic rate parameters of steady-state creep and plastic strain. A comparison of these parameters to the same rate parameters of the failure over a wide temperature–time range showed that there is an intimate relationship between these processes and that the structurally sensitive coefficients α and γ has identical responses to the relaxation phenomena in the alloys under the same test specifications.

In the cases studied here, the temperature-force conditions of the tests were such that deformation and failure processes, interacting with each other, had close kinetic parameters. In those cases where different rates of strain, failure, and structural changes result in a complex dependence of kinetic parameters on test conditions, it is expedient to invoke rheological models [9] to evaluate time variations of the activation parameters.

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