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DAMAGE ACCUMULATION IN ALUMINUM ALLOYS UNDER PLASTIC DEFORMATION AND CREEP

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The accumulation of local and bulk damage in D16 AT *and* 1201 T1 *aluminum alloys used in aircraft engineering is studied. The local damage level is calculated from data of thermoactivation analysis of the residual life of* D16 AT *alloy specimens after preliminary plastic deformation. The bulk damage level is determined from the elastic-modulus defect by measuring the natural frequency of* 1201 T1 *alloy specimens. Life tests of the specimens were performed at constant tensile loads and elevated temperatures. The dependence of the local damage on preliminary plastic strain at room temperature is obtained. The residual life of the specimens is calculated with allowance for the damage to the material in the initial stage of failure and compared with experimental results. Data are given on the kinetics of bulk-damage accumulation in various test regimes.*

Key words: *aluminum alloys, fracture, local and bulk damage, plastic strain, life, creep, modulus defect.*

Introduction. D16 T, AK4-1 T1, and 1201 T1 aluminum alloys produced by special thermal treatment are inhomogeneous and unstable in nature. Moreover, like any commercial alloy, these alloys contain defects at the microstructural level. In addition, submicro- and microcracks, pores, and other defects of various scale levels are formed and developed in them under loading. The damage accumulation process is closely related to plastic deformation. In order that the breaking of atomic bonds in a local volume responsible for fracture be irreversible, atomic rearrangement, i.e., microplastic strain, should occur in this volume. At the macrolevel, the damage accumulation and structural changes (including those related to plastic deformation) result in a decrease in the elastic modulus, an increase in damping, and changes in the strain and failure rates.

Several fracture criteria are known. In these criteria, the damage parameters are usually strain, the energy expended in failure, or the relative loading time. For complex structurally unstable alloys, however, these criteria agree with experiments only in particular cases.

The concentration criterion [1, 2] is considered using the kinetic concept of strength within the framework of the general approach to problem solving. According to this concept, fracture is the process of irreversible accumulation of damage formed as a result of thermofluctuation breaking of atomic bonds in a mechanically stressed material. For any loading type, macrofailure (macrocrack formation) occurs when a defect concentration threshold is reached in a local volume, resulting in avalanche-like defect merging.

The rate of increase of the damage concentration and size is variable and depends on the material, loading conditions, and the type of stress–strain state (SSS). The SSS type also determines the location of zones of high damage concentration in the material structure. The kinetics of damage accumulation in structurally unstable materials, in particular, precipitation-hardening aluminum alloys, is little understood.

The present work is a continuation of previous studies [3–5] of the fracture of the aluminum alloys used in aircraft engineering. Examples of estimation the local and bulk damage in specimens of these alloys subjected to uniaxial tension are considered.

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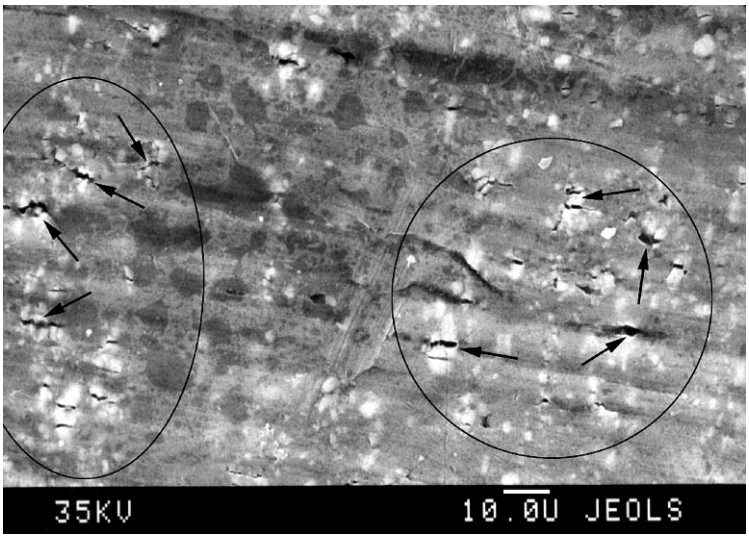

Fig. 1. Zones of surface microcracks in an AK-1 T1 specimen subjected to monotonic loading to a residual strain of $\varepsilon_p = 0.067$ at room temperature (the arrows show the largest microcracks).

Estimating the Local Damage to the Material from the Residual Life of Specimens. The use of the concentration criterion to determine the strength and life of aluminum alloys shows that it is necessary to distinguish between local and bulk types of damage [6]. Local damage was understood to be microcracks and pores (discontinuities) in a local material volume whose threshold concentration results in the formation of a macrocrack. This process is accompanied by damage accumulation in many other local volumes, whose number depends on the critical conditions. In each of these volumes, fracture is at different stages of development and the damage concentration is different in magnitude. The total damage in all local volumes is the bulk damage. The bulk damage is related, for example, to plastic strain, which is also the overall result of material flow over all local volumes [6].

The heterogeneous structure of the alloys is responsible for a nonuniform distribution of internal stresses in the loaded material. This, in turn, leads to a nonuniform distribution of the microcrack concentration in local volumes. Figure 1 shows an electron photomicrograph of the surface of an AK4-1 T1 specimen deformed plastically to a residual strain of $\varepsilon_p = 0.067$ at room temperature. One can see zones of high microcrack concentration between which there is almost no damage. When the microcrack concentration (or, generally, the relative fraction of voids) reaches the threshold value in one of these zones, the microcracks begin to merge to form a macrocrack nucleus. Theoretical estimates of this process can be found, for example, in [7]. As soon as the average distance between microcracks becomes equal to approximately three crack lengths [2], they begin to merge in an avalanche-like fashion. This estimate agrees with experimentally measured relative loss of material density by the moment of its complete failure under constant stresses [8]. According to the data of computer modeling of damage accumulation in specimen cross sections [9], the threshold concentration of discontinuities in the plane of fracture that leads to avalanche-like merging of discontinuities is 8–10%. This criterion, referred to as the concentration fracture criterion, has been supported at various scale levels [2].

Let us consider a procedure for estimating the local damage in D16 AT alloy from the residual life of specimens which were deformed plastically by tensile loading at room temperature. Flat specimens were made from 3 mm thick sheets of hardened and naturally aged D16 AT alloy. The preliminary plastic strain was 2, 4, 6, or 8%. The specimens were subjected to life tests under constant loads at temperatures of 398–473 K and stresses of 180–420 MPa. Primary test data for each alloy specimen in the initial state and after plastic deformation are given in [3]. The mean logarithmic life values of the specimens are presented in Table 1 for various test stresses and temperatures.

The equation linking the life τ, the stress σ, and the absolute temperature T [1] is written as

$$\tau = \tau_0 \exp\Big(\frac{U_0 - \gamma\sigma}{RT}\Big), \tag{1}$$

TABLE 1

Experimental and Theoretical Lifes of D16 AT Specimens in the Initial State and after Preliminary Plastic Deformation

Test conditions		Life τ_*, sec				
T, K	σ_0, MPa	$\varepsilon_p = 0$	$\varepsilon_p = 2$	$\varepsilon_p = 4$	$\varepsilon_p = 6$	$\varepsilon_p = 8$
398	420	$5.76 \cdot 10^4$ $8.22 \cdot 10^4$	$4.86 \cdot 10^4$ $1.28 \cdot 10^5$	$1.21 \cdot 10^5$ $1.66 \cdot 10^5$	$1.75 \cdot 10^5$ $1.90 \cdot 10^5$	— $1.95 \cdot 10^5$
	400	$2.16 \cdot 10^5$ $1.79 \cdot 10^5$	$3.02 \cdot 10^5$ $2.61 \cdot 10^5$	$3.31 \cdot 10^5$ $3.25 \cdot 10^5$	— $3.68 \cdot 10^5$	$2.84 \cdot 10^5$ $3.70 \cdot 10^5$
423	380	$2.73 \cdot 10^4$ $3.02 \cdot 10^4$	$6.44 \cdot 10^4$ $3.90 \cdot 10^4$	$9.09 \cdot 10^4$ $4.43 \cdot 10^4$	$9.23 \cdot 10^4$ $4.94 \cdot 10^4$	$1.14 \cdot 10^5$ $4.72 \cdot 10^4$
	360	$1.25 \cdot 10^5$ $6.30 \cdot 10^4$	$1.80 \cdot 10^5$ $7.60 \cdot 10^4$	$1.64 \cdot 10^5$ $8.30 \cdot 10^4$	$1.67 \cdot 10^5$ $9.26 \cdot 10^4$	$1.76 \cdot 10^5$ $8.62 \cdot 10^4$
	340	$2.27 \cdot 10^5$ $1.30 \cdot 10^5$	$3.05 \cdot 10^5$ $1.47 \cdot 10^5$	$2.89 \cdot 10^5$ $1.56 \cdot 10^5$	$2.77 \cdot 10^5$ $1.71 \cdot 10^5$	$2.19 \cdot 10^5$ $1.57 \cdot 10^5$
	320	$2.90 \cdot 10^5$ $2.67 \cdot 10^5$	— $2.88 \cdot 10^5$	$3.70 \cdot 10^5$ $2.94 \cdot 10^5$	— $3.17 \cdot 10^5$	$2.45 \cdot 10^5$ $2.88 \cdot 10^5$
448	300	$4.18 \cdot 10^4$ $4.91 \cdot 10^4$	$6.28 \cdot 10^4$ $4.76 \cdot 10^4$	$4.46 \cdot 10^4$ $4.53 \cdot 10^4$	$4.67 \cdot 10^4$ $4.74 \cdot 10^4$	$3.04 \cdot 10^4$ $4.18 \cdot 10^4$
	280	$9.36 \cdot 10^4$ $9.74 \cdot 10^4$	$9.47 \cdot 10^4$ $8.94 \cdot 10^4$	$9.82 \cdot 10^4$ $8.25 \cdot 10^4$	$7.52 \cdot 10^4$ $8.56 \cdot 10^4$	$6.50 \cdot 10^4$ $7.31 \cdot 10^4$
	260	$2.26 \cdot 10^5$ $1.94 \cdot 10^5$	$1.35 \cdot 10^5$ $1.68 \cdot 10^5$	$1.44 \cdot 10^5$ $1.49 \cdot 10^5$	$1.47 \cdot 10^5$ $1.53 \cdot 10^5$	$1.17 \cdot 10^5$ $1.30 \cdot 10^5$
	240	$3.44 \cdot 10^5$ $3.83 \cdot 10^5$	$2.74 \cdot 10^5$ $3.15 \cdot 10^5$	$3.07 \cdot 10^5$ $2.68 \cdot 10^5$	$2.88 \cdot 10^5$ $2.75 \cdot 10^5$	$1.91 \cdot 10^5$ $2.29 \cdot 10^5$
473	240	$4.95 \cdot 10^4$ $4.12 \cdot 10^4$	$3.11 \cdot 10^4$ $3.29 \cdot 10^4$	$2.82 \cdot 10^4$ $2.74 \cdot 10^4$	$2.81 \cdot 10^4$ $2.77 \cdot 10^4$	$2.41 \cdot 10^4$ $2.28 \cdot 10^4$
	220	$8.04 \cdot 10^4$ $7.89 \cdot 10^4$	$7.56 \cdot 10^4$ $5.93 \cdot 10^4$	$4.23 \cdot 10^4$ $4.81 \cdot 10^4$	$5.83 \cdot 10^4$ $4.82 \cdot 10^4$	$4.56 \cdot 10^4$ $3.90 \cdot 10^4$
	200	$1.92 \cdot 10^5$ $1.51 \cdot 10^5$	$1.60 \cdot 10^5$ $1.08 \cdot 10^5$	$1.20 \cdot 10^5$ $8.50 \cdot 10^4$	$1.17 \cdot 10^5$ $8.37 \cdot 10^4$	$1.01 \cdot 10^5$ $6.65 \cdot 10^4$
	180	$3.74 \cdot 10^5$ $2.90 \cdot 10^5$	$2.25 \cdot 10^5$ $1.97 \cdot 10^5$	$1.67 \cdot 10^5$ $1.49 \cdot 10^5$	$1.70 \cdot 10^5$ $1.45 \cdot 10^5$	$1.70 \cdot 10^5$ $1.14 \cdot 10^5$

Notes. 1) Experimental values are listed in the first rows and theoretical values are listed in the second rows. 2) The values of ε_p are in percent.

where τ_0, U_0, and γ are coefficients and R is the universal gas constant. According to [1, 2], the coefficients τ_0, U_0, and γ have a definite physical meaning, which was interpreted in [1] and refined in [2]. The action of thermal fluctuations is considered with reference to a certain group of atoms rather than to a single atom. The notion of dilaton, a region with a characteristic linear dimension Λ, is introduced. In accordance with the dilaton model for crack nucleation, the coefficients τ_0, U_0, and γ are determined by the period of thermal oscillations of the dilaton $\tau_{0d} = \tau_0$, its initial activation energy $U_{0d} = U_0$, and the activation volume of dilaton formation $\gamma_d = \gamma$. The coefficient γ is assumed to be proportional to the compliance of atomic bonds and the overstress (overload) factor in the regions of local failure.

With a decrease in the local overstress, the coefficient γ decreases and, as follows from Eq. (1), the life increases. From this viewpoint, preliminary plastic deformation, which is accompanied by relaxation of internal

TABLE 2

The Effect of Preliminary Plastic Strain (ε_p) on Local Damage (ω_*) and Activation Volume (γ_*) for Failure of D16 AT Alloy Specimens

ε_p, %	ω_*	γ_*, kJ/(mole · MPa)
0	0	0.1277
2	0.5706	0.1176
4	0.7598	0.1110
6	0.7853	0.1090
8	0.8557	0.1057

stresses, in particular, by preferred failure of the most overstressed zones in the material, should decrease the coefficient γ, thus increasing the residual life. At the same time, the failure process accompanying plastic deformation leads to damage to the material and, hence, decreases the residual life. The total result of these interrelated and counteracting processes can be different. One can see from Table 1 that preliminary plastic deformation results in an increase in residual life in some cases and a decrease or no change in others, depending on test conditions.

Indeed, if preliminary plastic deformation introduces damage ω_*, it should be taken into account in the formula for the residual life. Moreover, allowance should be made for the dependence of γ on the preliminary plastic strain ε_p and the introduced damage ω_*. Denoting the residual life by τ_* and the activation volume by $\gamma_* = \gamma(\varepsilon_p, \omega_*)$, we write the following formula for the residual life:

$$\tau_* = (1 - \omega_*)\tau_0 \exp\Big(\frac{U_0 - \gamma_*\sigma}{RT}\Big). \tag{2}$$

Here $\omega_* = 0$ in the absence of damage and $\omega_* = 1$ for the onset of microfailure of the material (macrocrack formation). To determine the coefficients U_0 and γ_* and the local damage ω_*, we use experimental data from Table 1.

For constant loading, we set $\sigma \approx \sigma_0 = \text{const}$, where σ_0 is the initial stress, and write (2) as

$$U(\sigma) = U_0 - \gamma_*\sigma = RT \ln\{\tau_*/[(1 - \omega_*)\tau_0]\}. \tag{3}$$

The initial fracture activation energy U_0 is assumed to be independent of plastic strain and accumulated damage. This assumption is based on the results of life studies [4, 5] of three aluminum alloys, where the quantity U_0 remained constant for each alloy under various loading conditions, and on the results of life studies of aluminum alloys in various structural states [1, 6].

We first determine the value of U_0 using test data for the specimens in the initial state for $\omega_* = 0$. Taking the values of σ_0, T, and τ ($\tau_0 = 10^{-13}$ sec [3–5]) from Table 1, we obtain a set of values of $U(\sigma) = RT \ln(\tau/\tau_0)$ and construct the relation $U(\sigma) = U_0 - \gamma\sigma$, in which the regression coefficients U_0 and γ are determined by the least-squares method. As a result, for D16 AT, we obtain $U_0 = 190.1$ kJ/mole, which is close to the values obtained for other batches and semiproducts of this material [4, 5]. Then, specifying the same values of U_0, for each ε_p, we determine ω_* and γ_* from the values of σ_0, T, and τ_* in Table 1 by solving Eq. (3) using the method of successive approximations.

The calculated local damage ω_* and coefficient γ_* as functions of the preliminary plastic strain ε_p are listed in Table 2 and plotted in Fig. 2. One can see from Fig. 2 that, with an increase in the preliminary plastic strain, the local damage ω_* (curve 1) rapidly reaches a value of $\omega_* \approx 0.8$, after with its growth rate decreases and then increases again during further deformation of the material. This trend is characteristic of failure at high stresses [9]. The damage growth is accompanied by a decrease in the activation volume γ_* (curve 2) since failure is a relaxation process and the most overstressed volumes of the material fail first [1, 2].

In Table 1, the residual lifes τ_* calculated by Eq. (2) using the obtained values of γ_* and ω_* (the second row) are compared with experimental data. With allowance for the scatter of the experimental data, the results are in satisfactory agreement.

Estimating the Bulk Damage from Variation of the Natural Frequency of Specimens. In contrast to the local damage in a material, which characterizes the microcrack concentration in a local volume responsible for failure, the bulk damage is determined by the total microcrack concentration in the entire specimen. The rate of increase of the local damage determines the time of occurrence of a macrocrack, whereas the bulk damage affects

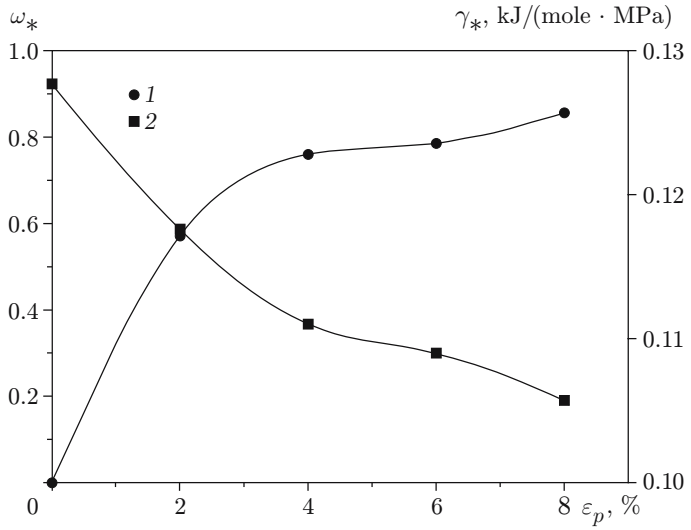

Fig. 2. Local damage ω_* (1) and activation volume γ_* (2) versus preliminary plastic strain ε_p for D16 AT alloy specimens (the curves are spline interpolations of the tenth order).

the propagation rate of the macrocrack [10]. One method for obtaining qualitative estimates of the bulk damage is to determine the variation of the natural frequency of specimens [11] by recording the variation of the compliance of the tested material due to the occurrence of discontinuities.

We give experimental data which show the kinetics of variation of the material compliance during failure at constant load and temperature. Circular specimens of diameter 8 mm made from a 1201 T1 alloy sheet were processed by milling fillets of radius 10 mm near the thread heads so that each specimen had two flat working parts of nearly rectangular cross section 3 mm thick. The specimens were subjected to life tests under uniaxial tension at constant temperature and load. The tests were carried out using ZSt-2/3 machines. Using the data of previous tests [4], from formula 1 we determined the following two loading regimes with approximately the same life: $T = 423$ K for an initial stress $\sigma_0 = 270$ MPa (the first regime) and $T = 473$ K for $\sigma_0 = 180$ MPa (the second regime). Three specimens with two working parts were tested for each regime. The bulk damage accumulation was determined form the variation of the tensile elastic modulus, which, in turn, was found from the variation of the natural frequency of flexural vibrations of the specimens. The vibration frequency was determined with a ChZ-57 digital frequency meter as the average of several repeated measurements with an error of $\pm 0.03\%$ for a probability of 0.95.

Before the tests, we measured the natural frequencies of specimens for each working part. During the tests, the frequencies were measured at an interval of $1.44 \cdot 10^5$ sec. The specimens were unloaded, removed from the furnace, and cooled to room temperature. The vibration frequencies of the specimens were measure in a special setup having a large mass and suspended on long steel strings. The setup was designed in such a manner that a specimen was fixed vertically, so that the vibration direction was perpendicular to the suspension and have little effect on the measurement results.

After $5.76 \cdot 10^5$ sec of testing, we measured the actual residual creep strain ε_p in the minimum cross section of the working parts of the specimens as the sum of the true transverse strains:

$$\varepsilon_p = \ln(b_0/b) + \ln(h_0/h),$$

where b_0, b and h_0, h are the initial and final widths and thicknesses of the working parts of the specimen, respectively. The true residual strain ε_* after failure was determined in a similar manner. If one of the working parts of a specimen failed before $5.76 \cdot 10^5$ sec had elapsed, further measurements were not performed and the specimen was fixed in grips of the testing machine by means of an adapter to determine only the residual life of the second working part and the residual strain in its minimum cross section after failure. The test and measurement results for all specimens are summarized in Table 3.

TABLE 3

Data on Specimens Made from a 1201 T1 Alloy Sheet
Tested under Constant Loading at Elevated Temperature

Test regime	Specimen No.	Initial state: F_0, Hz	Natural frequency of flexural vibration of specimens F (Hz) at various times (sec)				ε_p after $5.76 \cdot 10^5$ sec	ε_*	τ, 10^5 sec
			$1.44 \cdot 10^5$	$2.88 \cdot 10^5$	$4.32 \cdot 10^5$	$5.76 \cdot 10^5$			
423 K 270 MPa	1.1	257.22	254.88	253.64	252.81	250.62	0.022	0.065	8.418
	1.2	251.29	248.20	246.82	246.41	243.48	0.068	0.075	7.086
	2.1	264.39	262.24	261.31	261.31	260.50	0.006	0.092	8.226
	2.2	258.12	255.95	254.86	254.27	253.80	0.032	0.109	6.984
	3.1	247.76	244.63	242.27	—	—	—	0.137	3.900
	3.2	251.17	248.05	246.24	—	—	—	0.112	5.304
473 K 180 MPa	4.1	255.58	254.29	252.52	250.44	248.65	0.045	0.065	15.588
	4.2	258.72	257.70	256.59	254.81	251.58	0.0285	0.107	18.300
	5.1	256.62	256.39	256.35	254.20	252.09	0.019	0.098	19.098
	5.2	250.82	249.94	249.65	247.50	245.77	0.058	0.085	17.352
	6.1	215.53	209.78	—	—	—	—	0.255	2.070
	6.2	243.18	240.68	—	—	—	—	0.134	18.225

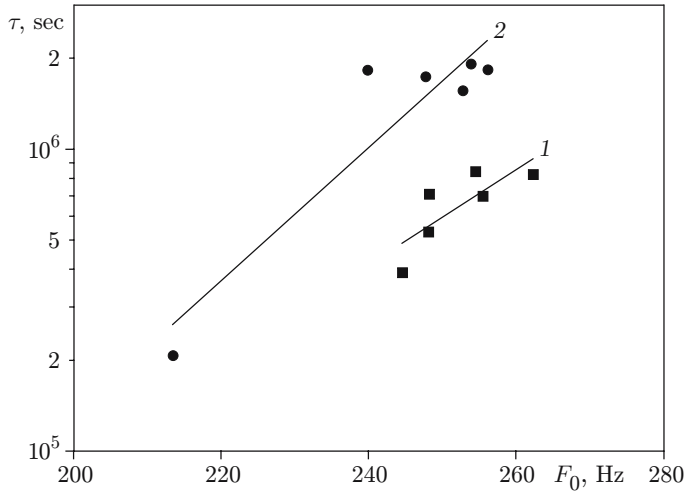

Fig. 3. Statistical dependence of the life τ of 1201 T1 alloy specimens on the squared natural frequency of their flexural vibrations F_0 in the initial state: curve 1 refers to 423 K and 270 MPa and curve 2 refers to 473 K and 180 MPa.

Figure 3 gives the statistical dependence of the logarithm of the specimen life τ and the squared frequency of their natural vibrations F_0 in the initial state. The curves represent the linear-regression equations. The difference in natural frequency between individual specimens in the initial state is due to the presence of initial defects in the material, which have affect the scatter of life values.

The formula for the natural frequency of flexural vibrations of a cantilevered beam [12] implies that the squared natural frequency F is proportional to the flexural rigidity of the beam and inversely proportional to the fourth power of its length. Because of the damage arising during failure, the tensile elastic modulus E decreases and becomes equal to E'.

Let us find the relationship between the variation of the tensile elastic modulus and variation of the squared natural frequency of flexural vibrations of the beam. The relative variation of the tensile elastic modulus, called the modulus defect, is denoted by δ and given by

$$\delta = (E' - E)/E.$$

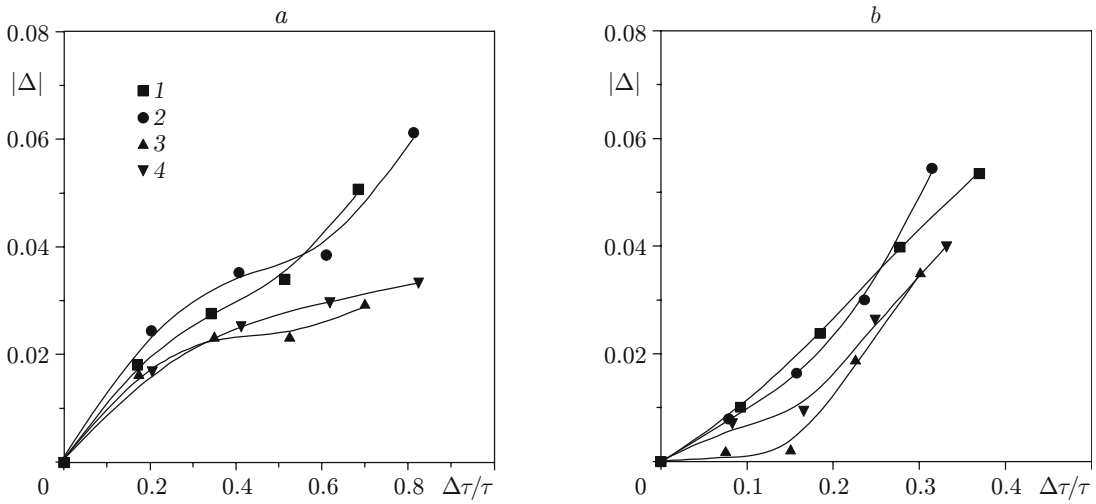

Fig. 4. Relative variation of the squared natural frequency of flexural vibrations $\Delta = (F^2 - F_0^2)/F_0^2$ of 1201 T1 alloy specimens versus the relative failure time $\Delta\tau/\tau$ (points 1 to 4 refer to the specimen numbers in Table 3; the curves are spline interpolations of tenth order): (a) 423 K and 270 MPa (curves 1, 2, 3, and 4 refer to specimen Nos. 1.1, 1.2, 2.1, and 2.2, respectively); (b) 473 K and 180 MPa (curves 1, 2, 3, and 4 refer to specimen Nos. 4.1, 4.2, 5.1, and 5.2, respectively).

TABLE 4

Calculated Values of the Tensile Modulus Defect δ in Specimens Made from 1201 T1 Alloy Sheet after $5.76 \cdot 10^5$ sec of Testing under Constant Loading and Elevated Temperature

Test regime	Specimen No.	ε_p, %	$\Delta\tau/\tau$	$\lvert\delta\rvert$, 10^{-2}
423 K 270 MPa	1.1	2.20	0.6843	8.65
	1.2	6.80	0.8129	7.67
	2.1	0.60	0.7002	5.44
	2.2	3.20	0.8247	4.49
473 K 180 MPa	4.1	4.50	0.3695	7.68
	4.2	2.85	0.3148	8.97
	5.1	1.90	0.3016	5.72
	5.2	5.80	0.3320	4.08

In constructing the relation between the flexural rigidity of the working part of the beam and the tensile elastic modulus defect, we take into account the shift of the neutral axis due to bending, assuming that the quantity E remains unchanged during compression. Expressing the tensile modulus defect and the relative variation of the squared natural frequency of the beam in terms of the relative shift of the neutral axis due to bending and eliminating this variable, we obtain the following relation between the relative variations of the modulus and frequency

$$\delta = -4(1 - \sqrt{1 + \Delta}\,)/(2 - \sqrt{1 + \Delta}\,)^2,$$

where $\Delta = (F^2 - F_0^2)/F_0^2$. For small values of Δ, we can use the approximate formula

$$\delta \approx 2\Delta. \tag{4}$$

Figure 4 shows the relative variation of the squared frequency of vibrations Δ versus the relative failure time $\Delta\tau/\tau$ for two test regimes. The quantity δ defined by formula (4) should be refined since the vibration frequency decreases not only due to the discontinuities formed but also due to the creep strain ε_p accumulated during failure. To find an approximate expression for the correction taking into account the variation of δ due to increasing creep strain, we use the statistical dependence represented by the curve in Fig. 5. The linear regression function is used because of the smallness of $|\Delta|$. The values of the tensile modulus defect $|\delta|$ with the correction for the strain ε_p calculated by the straight-line equation plotted in Fig. 5 are listed in Table 4. The defect of the tensile modulus is taken as an approximate qualitative estimate of the bulk damage of the material.

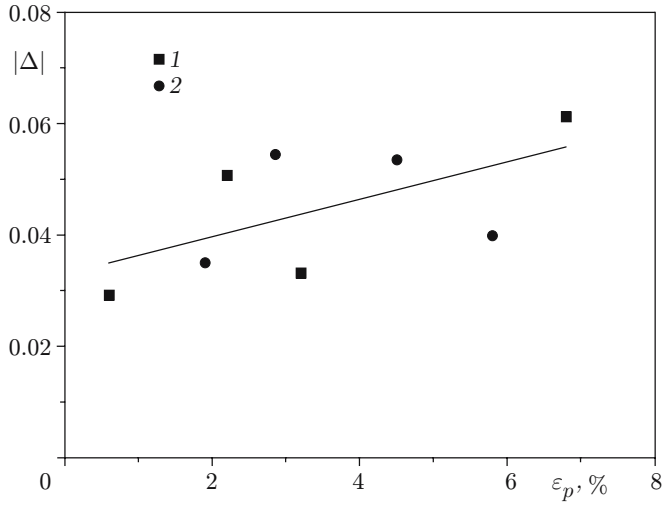

Fig. 5. Statistical dependence of the relative variation of the squared natural frequency of flexural vibrations of 1201 T1 alloy specimens on the plastic strain in the working part after $5.76 \cdot 10^5$ sec of testing: points 1 refer to 423 K and 270 MPa and points 2 to 473 K and 180 MPa.

Conclusions. Special techniques and methods are required to determine qualitative values of the local and bulk damage levels. In the present paper, a method was proposed to indirectly estimate the damage level from its effect on the residual life of specimens (local damage) and on the tensile elastic modulus defect (bulk damage). To determine the local damage, an additional parameter ω_* characterizing the conditional local damage is included in the life equation (1). Experimental data on the residual life of D16 AT alloy subjected to preliminary plastic strain ε_p were used to determine the parameters ω_* and $\gamma_*(\varepsilon_p, \omega_*)$ [see Eq. (2)]. The calculation using Eq. (2) agrees with the results of experiments, in which the residual life of specimens subjected to preliminary plastic deformation increases in some cases and decreases or remains unchanged in other cases, depending on test regime (Table 1). An approximate estimate of the bulk damage of 1201 T1 alloy specimens in creep was obtained by the method of determining the elastic-modulus defect. The modulus defect depends on the loading conditions and the stage of failure (Table 4). To gain more detailed information on the local and bulk damage levels in materials, it is necessary to obtain new experimental data using the well-known and new methods.

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