Journal of Applied Mechanics and Technical Physics, Vol. 42, No. 4, pp. 725–730, 2001

KINETIC APPROACH TO PREDICTION OF THE LIFE OF ALUMINUM ALLOYS UNDER VARIOUS THERMAL–TEMPORAL LOADING CONDITIONS

M. G. Petrov and A. I. Ravikovich

UDC 539.4:620.1

Results of investigation of the life of D16 T, AK4-1 T1*, and* 1201 T1 *aluminum alloys are generalized on the basis of the kinetic concept of failure. The life is studied under creep at constant loads and loads increasing with different rates and at different temperatures. The temperature is varied within the range of* 473–77 K*, and the duration of tests ranges from fractions of a second to ten thousand hours. Information on the effect of internal-stress relaxation on the life of alloys is obtained. A method for predicting the life with allowance for relaxation processes in solids is verified experimentally.*

Introduction. At present, the theories of material strength can be divided into two main groups: the limit-state theories and the kinetic failure theories, which take into account the duration of loading. The limit-state theories agree with experimental data only in particular cases of failure. Within the framework of kinetic theories, various fracture criteria are used. In practice, different concepts of strength of a material and units of its working capacity (duration of loading, number of cycles or blocks of loads, accumulated residual plastic strain, etc.) are used, depending on the conditions and character of loading.

As the potentialities and complexity of experimental methods grew, it became necessary to combine solutions of particular problems of the short- and long-term strength, creep and fatigue, and thermocyclic strength. This approach, which gives a new interpretation of fracture and deformation of loaded solids, was proposed and developed at the Leningrad scientific school headed by S. N. Zhurkov in the 1950s. According to this approach, a solid body is a physical medium in which the action of an external force depends on the interaction between atoms performing heat motion. Nonuniformity of heat motion plays an important role. Failure is treated as an irreversible accumulation of submicro- and microcracks that occur as a result of thermofluctuation rupture of interatomic bonds in a mechanically stressed material [1]. It is assumed that, for any type of loading, the main characteristic is the life, i.e., the time from the moment a load is applied to the moment when macrocracks occur.

In reality, deformation and failure of structural alloys are complex processes and call for special theoretical and experimental investigations. In the present paper, we give results on the life of D16 T, AK4-1 T1, and 1201 T1 aluminum alloys used in aircraft industry.

Experimental Technique. According to the kinetic failure theory [1], the relation between the life τ, the acting stress σ, and the absolute temperature T for constant stresses has the form

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

where τ_0, U_0, and γ are coefficients and R is the universal gas constant. For time-dependent stresses $\sigma(t)$, the life is determined from the formula

$$\int\limits_0^\tau \frac{dt}{\tau_0 \exp\{[U_0 - \gamma\sigma(t)]/RT\}} = 1, \tag{2}$$

which is derived with the use of the principle of linear summation of damages.

Chaplygin Siberian Aviation Institute, Novosibirsk 630051. Translated from Prikladnaya Mekhanika i Tekhnicheskaya Fizika, Vol. 42, No. 4, pp. 190–196, July–August, 2001. Original article submitted December 14, 2000; revision submitted February 20, 2001.

TABLE 1

Initial Data and Test Results for Aluminum Alloys at Constant Loads (points 3 in Figs. 1–3)

D16 T (Fig. 1)			AK4-1 T1 (Fig. 2)			1201 T1 (Fig. 3)		
T, K	σ_0, MPa	τ, sec	T, K	σ_0, MPa	τ, sec	T, K	σ_0, MPa	τ, sec
398	420	$5.76 \cdot 10^4$	403	240	$4.73 \cdot 10^7$	398	300	$5.76 \cdot 10^5$
398	400	$2.26 \cdot 10^5$	423	240	$3.16 \cdot 10^6$	398	280	$4.83 \cdot 10^6$
423	380	$2.95 \cdot 10^4$	423	200	$3.48 \cdot 10^7$	398	260	$2.36 \cdot 10^7$
423	360	$1.29 \cdot 10^5$	423	160	$6.96 \cdot 10^7$	423	240	$1.44 \cdot 10^7$
423	340	$2.32 \cdot 10^5$	423	140	$1.04 \cdot 10^8$	433	280	$2.72 \cdot 10^4$
423	320	$3.19 \cdot 10^5$	448	240	$8.57 \cdot 10^5$	433	260	$1.22 \cdot 10^5$
423	240	$8.78 \cdot 10^6$	448	200	$5.26 \cdot 10^6$	433	250	$4.02 \cdot 10^5$
448	300	$4.20 \cdot 10^4$	448	160	$2.13 \cdot 10^7$	433	240	$1.46 \cdot 10^6$
448	240	$4.23 \cdot 10^5$	—	—	—	433	220	$1.44 \cdot 10^7$
448	200	$2.14 \cdot 10^6$	—	—	—	448	200	$1.52 \cdot 10^7$
448	160	$1.35 \cdot 10^7$	—	—	—	448	180	$4.06 \cdot 10^7$
448	140	$1.66 \cdot 10^7$	—	—	—	473	160	$1.83 \cdot 10^7$
473	180	$3.82 \cdot 10^5$	—	—	—	473	140	$4.26 \cdot 10^7$

TABLE 2

Initial Data and Test Results for Aluminum Alloys at Loads Increasing with a Constant Rate

Point Nos. in Figs. 1–3	D16 T (Fig. 1)			AK4-1 T1 (Fig. 2)			1201 T1 (Fig. 3)		
	T, K	σ_*, MPa	t_*, sec	T, K	σ_*, MPa	t_*, sec	T, K	σ_*, MPa	t_*, sec
4	293	644	$4.53 \cdot 10^7$	293	702	$7.25 \cdot 10^6$	293	563	$9.30 \cdot 10^6$
	—	—	—	293	698	$3.71 \cdot 10^7$	293	545	$4.62 \cdot 10^7$
5	—	—	—	293	679	$1.17 \cdot 10^6$	—	—	—
	—	—	—	293	682	$3.01 \cdot 10^4$	—	—	—
6	—	—	—	373	657	$2.76 \cdot 10^4$	—	—	—

In Eqs. (1) and (2), the coefficients τ_0, U_0, and γ are the physical constants. The preexponential factor $\tau_0 \approx 10^{-13}$ sec characterizes the period of atomic oscillations in the solid. The coefficient U_0, which has the dimension of energy, depends on the rupture energy of interatomic bonds, and it is assumed to be a structurally independent parameter. It is called the initial energy of fracture activation. The coefficient γ, which is referred to as the activation volume, is a structurally dependent parameter. According to [1], the coefficient γ depends on the internal local stresses and it is an integral characteristic of nonuniformity of the load distribution over atomic bonds. It is determined as the product of the atomic volume by the coefficient of overstresses at the sites where local fracture occurs. The lower the local overstress, the stronger the material, the longer its life, and the smaller γ.

Investigations of the life of various materials (polymers, metals, and alloys) [2–5] and of the residual life after preliminary plastic strain [6] show that fracture is accompanied by internal-stress relaxation, which can be found from the change in the coefficient γ determined by mechanical tests.

The test program was developed so as to study the entire range of the activation volume from $\gamma_{\min}$ to $\gamma_{\max}$. For example, to determine $\gamma_{\max}$, we performed impact tests in which the internal stresses had no time to relax during fracture and also tests at cryogenic temperature where the relaxation was almost absent. To determine intermediate values of γ and $\gamma_{\min}$, we performed break tests with different strain rates of specimens at normal and elevated temperatures, creep tests at a constant load and elevated temperature, and creep tests under a long-time load increasing at a constant rate at normal and elevated temperatures.

In the life experiments, we used standard smooth specimens under uniaxial tension. The loading conditions were varied within the temperature range from 473 to 77 K, and the test duration ranged from fractions of a second to ten thousand hours.

TABLE 3

Initial Data and Test Results for the D16 T Aluminum Alloy at Different Strain Rates $\dot{\varepsilon}$ [sec^{-1}]

T, K	$\dot{\varepsilon} = 100$			$\dot{\varepsilon} = 1.25$			$\dot{\varepsilon} = 3 \cdot 10^{-3}$		
	Point Nos. in Fig. 1	σ_*, MPa	t_*, 10^{-3} sec	Point Nos. in Fig. 1	σ_*, MPa	t_*, 10^{-1} sec	Point Nos. in Fig. 1	σ_*, MPa	t_*, sec
77	—	—	—	—	—	—	19	812	45
123	—	—	—	—	—	—		747	34
223	—	—	—	—	—	—	18	694	52
293	16	763	1.53	15	706	1.13	17	658	62
373	14	722	1.40	13	651	0.96	7	634	39
423		662	1.29	12	652	0.90	9	610	41
473		686	0.98	11	612	0.84	10	592	38

The specimens were made of the serial quenched and naturally aged D16 T alloy (bar and sheet), serial quenched and artificially aged AK4-1 T1 alloy (bar), and 1201 alloy (slab). After quenching and tension with a 2.5% residual strain, the 80 mm-thick 1201 slab produced by forging and rolling was artificially aged to the state T1.

Long-term loading tests were performed with the use of ZSt 2/3 machines. The specimens were tested under creep conditions at constant loads and elevated temperature and under conditions of slow stepwise increase in load with a constant rate at normal and elevated temperatures. Short-term loading tests were performed at normal, elevated and cryogenic temperatures. The loading at a rate recommended by the Russian standard GOST for mechanical tests was performed with the use of a GURM-10 hydraulic breaking machine equipped with a removable electric furnace and a cryogenic chamber into which liquid nitrogen was injected. Shock loading was performed on a high-speed rotation machine SM-10. We tested up to 10 specimens in the same regimes of impact loading; in other cases, we tested three to five specimens. Temperature and load deviations from the specified values met the GOST standard. The initial stresses σ_0 (in constant-load tests), the strain rate $\dot{\varepsilon}$, the temperature T, the rupture stress σ_* in the specimen neck, and the rupture times τ and t_* for constant and uniformly increasing stresses, respectively, are listed in Tables 1–4 (τ, t_*, and σ_* are the averaged experimental data for specimens tested in identical regimes).

The experimental data obtained for different regimes of loading were processed by the following method. For creep tests at a constant load, formula (1) with $\sigma = \sigma_0 = \text{const}$ becomes the expression

$$U(\sigma) = U_0 - \gamma\sigma = RT \ln(\tau/\tau_0), \tag{3}$$

which is called the force relation for the fracture activation energy. For $\tau_0 = 10^{-13}$ sec and given σ, T, and τ, the values of U can be determined. If γ is constant for identical experimental conditions, the U–σ dependence must be linear. The line was plotted by statistic processing of the fracture activation energy by the least squares procedure. Extrapolating the resulting straight line to the ordinate axis (as $\sigma \to 0$), we found the initial fracture activation energy U_0; the activation volume γ was determined from the slope of the line.

To process the experimental data obtained for the stress varied with time, we used Eq. (2). Let the stress increase with a constant rate w from zero up to the moment t_* when the specimen fails at a stress σ_*. Substituting $\sigma(t) = wt$, where $w = \sigma_*/t_*$, into (2), we obtain

$$B\sigma_* = -\ln\Big[At_* \frac{1 - \exp(-B\sigma_*)}{B\sigma_*}\Big], \tag{4}$$

where $A = \nu_0 \exp[-U_0/(RT)]$ and $B = \gamma/(RT)$ ($\nu_0 = 1/\tau_0$). The product $t_*[1 - \exp(-B\sigma_*)]/(B\sigma_*)$ in (4) is the equivalent rupture time τ_* for constant stresses $\sigma = \sigma_*$. To determine τ_* from Eq. (4), we calculate B from experimental values of σ_*, t_*, and T by an iterative method. In this case, the coefficients U_0 and τ_0 have the same values as for $\sigma = \text{const}$. In processing the experimental data, τ and σ in (3) are replaced by the values of τ_* and σ_*.

Figures 1–3 show the fracture activation energy obtained by processing experimental data versus stress for D16 T, AK4-1 T1, and 1201 T1 alloys, respectively. Strength lines 1 and 2 refer to the minimum ($\gamma_{\min}$) and maximum ($\gamma_{\max}$) values of the activation volume. The points enumerated in Figs. 1–3 are listed in Tables 1–4.

TABLE 4

Initial Data and Test Results for the AK4-1 T1 and 1201 T1 Aluminum Alloys for the Strain Rate $\dot{\varepsilon} = 3 \cdot 10^{-3}$ sec^{-1}

T, K	Point Nos. (see Figs. 2 and 3)	AK4-1 T1 (Fig. 2)		1201 T1 (Fig. 3)	
		σ_*, MPa	t_*, sec	σ_*, MPa	t_*, sec
77	19	741	30	720	22
77		—	—	708	37
77		—	—	701	54
123		699	52	—	—
293	17	666	21	567	232
343	8	—	—	572	195
373	7	—	—	598	204
398		—	—	576	212
448	10	—	—	521	194
473		—	—	454	178

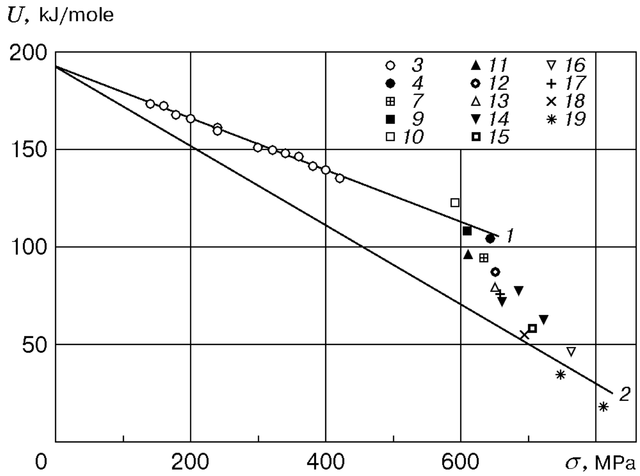

Fig. 1

Discussion of Results. It follows from the given dependences of the fracture activation energy on the stress that some of the values of $U(\sigma)$ lie far from the straight line. One can see from Fig. 1 that the values of $U(\sigma)$ for the D16 T alloy (points 3, 4, 9, and 10) obtained in test Nos. 3, 4, 9, and 10,respectively, are close to line 1. In this case, fracture occurs under conditions favorable to relaxation (long-time testing and elevated temperature). This suggests that the internal-stress relaxation is completed earlier than the fracture. As a result, the coefficient γ acquires a constant value close to $\gamma_{\min}$. For test Nos. 15, 16, 18, and 19, the values of $U(\sigma)$ (points 15, 16, 18, and 19, respectively) lie close to line 2. In these cases, the fracture occurred almost without internal-stress relaxation, and the coefficient γ took the constant value $\gamma_{\max}$.

In the region between curves 1 and 2 (see Fig. 1), interaction and competition between the fracture and stress-relaxation processes are manifested. In this case, the coefficient γ acquires intermediate values within the range from $\gamma_{\min}$ to $\gamma_{\max}$, depending on test conditions. The same dependence of the activation volume γ on relaxation is observed for the AK4-1 T1 and 1201 T1 alloys (see Figs. 2 and 3). The parameters U_0, $\gamma_{\min}$, and $\gamma_{\max}$ for each alloy are listed in Table 5.

An analysis of the experimental dependence of the coefficient γ on relaxation allows us to determine

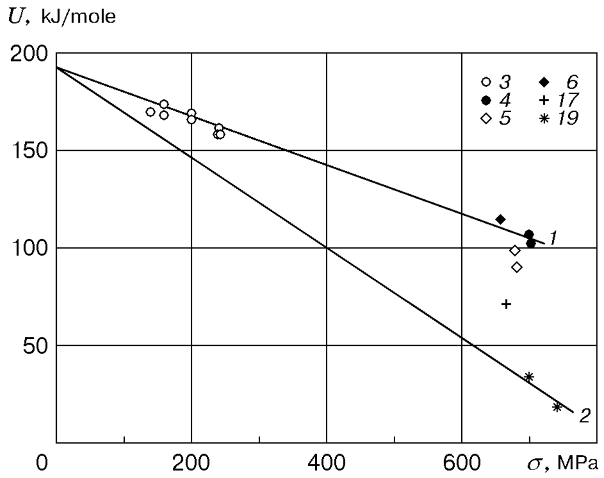

Fig. 2

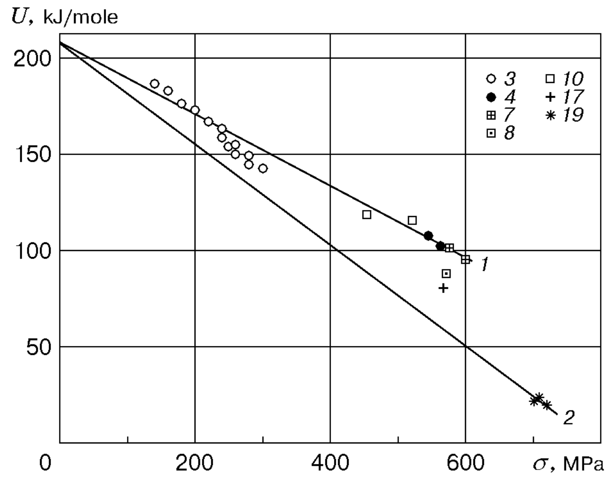

Fig. 3

TABLE 5

Activation Parameters of Fracture of Aluminum Alloys

Alloy	U_0, kJ/mole	$\gamma_{\min}$, kJ/(mole · MPa)	$\gamma_{\max}$, kJ/(mole · MPa)
D16 T	193	0.135	0.203
AK4-1 T1	193	0.126	0.234
1201 T1	208	0.188	0.264

temperature–temporal test regimes, where the relaxation process does not occur or is completed earlier than fracture. The knowledge of regions where $\gamma = \text{const}$ for various σ_*, τ_*, and T makes it possible to reduce the number and duration of experiments on determining the parameters in Eqs. (1) and (2). To obtain the general idea of the dependence of the fracture activation energy on stresses, it is sufficient to test monotonically loaded specimens with a constant loading rate and different temperature for each specimen. Thereafter, one can easily choose loading regimes for more detailed investigations.

According to Eqs. (1) and (2), the stress and temperature enter the exponent and equally influence the life. These equations imply that the rupture stress of the specimen increases as the temperature decreases. This is supported by experimental data (see Tables 1–4). An increase in the loading rate or strain rate also increases the stress (see Tables 3 and 4) since the duration of loading is reduced. For example, in the case of an impact load, higher stresses are required to rupture a specimen in a short time. The same specific features are typical of the flow stresses of a material, which are referred to as "yield points." The flow stresses can be determined with the use of rheological models [7].

The dependence of the "rupture" stress on temperature and loading duration (i.e., dependence of life on stress and temperature) is rather complex in the range of variation of γ from $\gamma_{\min}$ to $\gamma_{\max}$, where fracture and internal-stress relaxation interact. On the one hand, an increase in the duration of loading decreases the "rupture" stress; on the other hand, long-term loading of a material favors more complete relaxation and, hence, makes the material stronger. These processes are characterized by different activation parameters; as a result, the dependence of γ on test conditions becomes more complex. In these cases, one should use rheological models that describe structural changes in the material characterized by the change in the activation volume [7]. Modeling the relaxation processes allows one to predict the life of a material for arbitrary changes in stresses and temperature [5, 7].

In summary, the kinetic approach to the strength problem allows one to analyze the processes at the microlevel and to determine the parameters of these processes from results of mechanical tests.

The authors are grateful to Yu. I. Ustinovshchikov for providing data on impact tests and to A. P. Kuznetsov and N. I. Zhichina for performing most of the creep tests.

REFERENCES

1. V. R. Regel', A. I. Slutsker, and É. E. Tomashevskii, *Kinetic Nature of the Strength of Solids* [in Russian], Nauka, Moscow (1974).
2. I. V. Razumovskaya, Yu. T. Korabel'nikov, G. M. Bartenev, and K. V. Panferov, "The life and relaxation of polymers," *Mekh. Polimer.*, No. 4, 629–632 (1969).
3. V. A. Stepanov, V. V. Shpeizman, and L. V. Zhoga, "Dependence of the strength of brittle solids on temperature and time," *Fiz. Met. Metalloved.*, **42**, No. 5, 1064–1074 (1976).
4. M. G. Petrov, "Strain–fracture relation for loaded solids," *Prikl. Mekh. Tekh. Fiz.*, No. 6, 122–126 (1976).
5. V. A. Stepanov, N. N. Peschanskaya, and V. V. Shpeizman, *Strength and Relaxation in Solids* [in Russian], Nauka, Leningrad (1984).
6. N. A. Moshkin and A. I. Ravikovich, "Application of the kinetic strength concept to prediction of the life of light alloys after preliminary plastic strain," *Prikl. Mekh. Tekh. Fiz.*, No. 1, 128–132 (1978).
7. M. G. Petrov, "Rheological properties of materials from the point of view of physical kinetics," *Prikl. Mekh. Tekh. Fiz.*, **39**, No. 1, 119–128 (1998).