TREATMENT

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EFFECT OF STRAIN AGING ON PHYSICAL AND MECHANICAL PROPERTIES OF AUSTENITIC CHROMIUM-NICKEL STEEL

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The processes occurring in chromium-nickel steel 08Kh18N10T due to strain aging are studied. The tensile mechanical properties are determined. Relaxation tests and metallographic and x-ray diffraction analyses are performed. The internal energy state of the steel is estimated with the help of acoustic emission.

Key words: stainless steel, strain aging, mechanical properties, structure, acoustic emission.

INTRODUCTION

Some traditional ways for improving the mechanical properties of steels and alloys have almost exhausted themselves of have proved to be inefficient at the present time. Specifically, the possibilities of alloying are limited by scarceness of some elements.

On the other hand, the tightening requirements of the industry on the quality of steels and alloys with special properties and high strength characteristics make specialists search for expedient methods for fabricating simple structural and corrosion-resistant materials.

Promising ways for raising mechanical characteristics consist in fuller use the possibilities of the structure-phase method of hardening, of combined deformation, deformation-heat, radiation, and other processes.

A great number of reviews, monographs, and original publications has been devoted to the regular features of variation of mechanical properties in steels of type 18-10 of austenitic class as a result of martensitic transformation [1 – 6]. These structural materials possess a number of positive characteristics. In the first turn, they are adaptable to manufacture, i.e., are satisfactorily deformable at high temperatures, withstand cold bending, flaring, straightening, etc. Such steels are relatively heat-resistant and refractory, possess high resistance to total corrosion and corrosion cracking, are well weldable, and resistant to intercrystalline corrosion at optimum alloying. A typical feature of steels of this class after heat treatment is a low stability of austenite that transforms into martensite in the process of hardening from high temperatures. Subsequent plastic deformation, especially at a low temperature, causes further transformation of austenite into martensite.

In the process of martensitic transformation the state of retained austenite changes substantially due to volume and phase hardening [5 – 11]. As a result of low-temperature plastic deformation of hardened austenitic steel the nonuniformly distributed microstresses are ordered and the location of defects in the volume of the metal changes [8].

The observed strengthening of the steel is chiefly a result of mechanical hardening. In martensitic transformation the neighbors of any atom in austenite remain neighbors of the same atom in martensite. Dislocations in the $\gamma \rightarrow \alpha$ transformation by martensitic mechanism do not disappear but are "transferred" from the initial phase to the new phase, i.e., the martensite inherits the structure of strained austenite. The very high density of dislocations in martensite, which are fixed by carbon atoms and carbide segregations, is responsible for the high strength of the steel. However, the use of the mentioned treatment process is hindered by the necessity to use powerful equipment for pressure or tensile treatment because it is necessary to subject the steel to a high deformation (at least 50%) in order to ensure high strength. Another substantial drawback is the low resistance of strongly hardened steel to brittle fracture. As the density of dislocations in martensite, which cause considerable hardening, increases; the resistance to crack propagation (a very important charac-

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teristic of a structural material) does not change or even decreases.

The aim of the present work consisted in studying the effect of strain aging on the strength characteristics of chromium-nickel steel of type 18-10.

METHODS OF STUDY

We studied austenitic chromium-nickel steel 08Kh18N10T containing (wt.%) up to 0.08 C, 18 Cr, up to 2.0 Mn, up to 0.8 Si, 0.7 Ti, 9.0 – 11.0 Ni, < 0.02 S, < 0.035 P.

Complex-geometry specimens with working sizes 3.2 × 2.0 × 14.0 mm were cut from a sheet 3 × 710 × 1420 mm in size along the rolling axis.

The specimens were annealed in a vertical tube furnace at 1050 – 1100°C for 60 min. After the production, mechanical grinding, and annealing the specimens were divided into three batches and treated in the following modes:

(*1*) water quenching from 1050 – 1080°C (according to GOST 5582–75);

(*2*) quenching (according to GOST 5582–75), plastic deformation at – 196°C with residual strain of up to 40%, heating at a rate of 0.0875 – 0.7 K/sec from – 196°C to a temperature ranging within about 130 – 830°C, cooling with the furnace;

(*3*) quenching (according to GOST 5582–75), plastic deformation at – 196°C with residual strain of up to 20%, heating at a rate of 0.175 K/sec from – 196°C to a temperature ranging within 300 – 800°C, and tempering at these temperatures at a stress $\sigma_i = (0.1 - 0.75)\sigma_{0.2}$ for 1 h.

Fluctuation of temperature in short-term and long-term tests did not exceed ± 3 and ± 5°C, respectively.

Mechanical characteristics were determined in tensile tests (until failure) of specimens of all the batches at 20°C with deformation rate 10^{-3} sec^{-1}.

The metallographic study was performed using a MIM-10 light microscope. The x-ray diffraction study of the specimens was performed using a DRON-1 diffractometer in chromium radiation with automatic recording of interference line (311) of austenite on diagram tape. The relative error of measurement of crystal lattice parameter computed for the conditions of the experiment was $\Delta a/a = 0.016\%$.

After the heat treatment we performed tests for stress relaxation at the appropriate temperature in a tensile testing machine with 30-min duration of the recording. The relaxation effect $\Delta\sigma_{rel}$ was determined from the decrease in the stress from the initial value $\sigma_i = 0.9\sigma_{0.2}$ or $0.5\sigma_{0.2}$ to some final σ_t attained in the specified period of time.

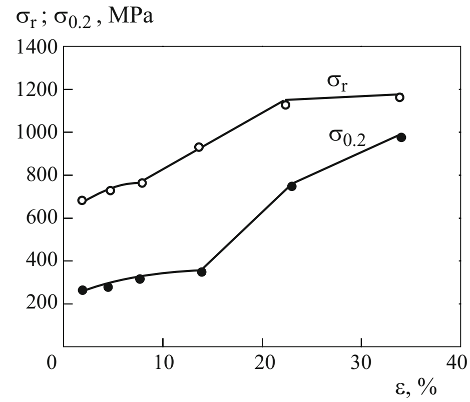

Fig. 1. Dependence of strength characteristics ($\sigma_{0.2}$ and σ_r) of steel 08Kh18N10T on the strain level ε at – 196°C (treatment by regime *2*, $v_h = 0.175$ K/sec, $t_{temp} = 500$°C).

The microhardness *HV* was determined with the help of a PMT-3M device at a load of 1 N.

The acoustic emission (AE) was measured using a piezoelectric transducer with gain factor of 40 dB in the process of tensile deformation of the specimens at 20°C. The level of noises reduced to the inlet was 10 μV. The sensitivity of the installation was 10^{11} V/m; the operating range was 0.1 – 0.8 MHz. The input signal of the AE transducer was transmitted to the inlet of the primary amplifier with controlled gain factor of up to 60 dB and then detected in a computer with built-in analog-to-digital converter with readout frequency of 3 MHz. The informative parameters of AE were the AE activity ($\dot{N}_\varepsilon = dN/d\varepsilon$), the total value ($N_\Sigma$), and the total energy of the AE signal (W_Σ). The averaging time for plotting the histogram of variation of a single pulse W_p as a function of the degree of strain was 0.56 sec.

RESULTS AND DISCUSSION

The table presents the strength characteristics of the steel after different modes of treatment. It can be seen from the data of Fig. 1 and of Table 1 that after the martensitic $\gamma \to \alpha$ transformation that occurs in quenching and plastic deformation at – 196°C (mode *2*) the strength characteristics in-

TABLE 1. Mechanical Characteristics of Steel 08Kh18N10T after Treatment by Different Modes

Treatment mode		σ_r, MPa	$\sigma_{0.2}$, MPa
Quenching from 1060 – 1080°C in water at 20°C	(*1*)	640	225
Mode 1 + deformation at – 196°C with residual degree ≤ 20%, tempering at 500°C, 1 h	(*2*)	1020	640
Mode 1 + deformation at – 196°C with residual degree ≤ 20%, tempering at 500°C, 1 h at stress $\sigma_i = 0.5\sigma_{0.2}$	(*3*)	1750	1060

Note. The modes of treatment are numbered in parentheses.

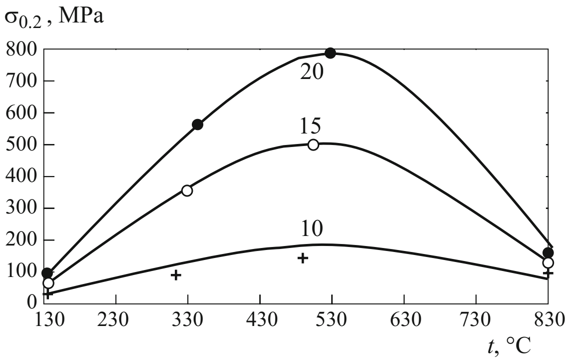

Fig. 2. Dependence of the yield strength of steel 08Kh18N10T on the tempering temperature (treatment by mode *2*, $v_h = 0.175$ K/sec). The degree of deformation at – 196°C is given at the curves.

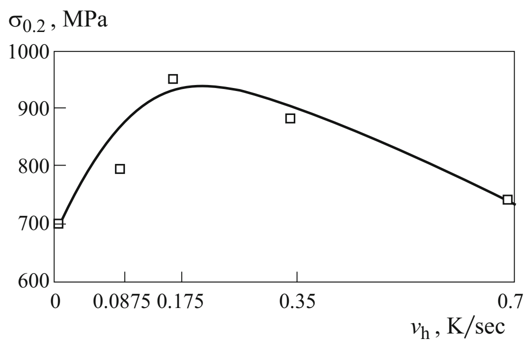

Fig. 3. Dependence of the yield strength of steel 08Kh18N10T on the rate of heating (treatment by mode *2*, $t_{temp} = 500$°C, $\varepsilon = 20\%$).

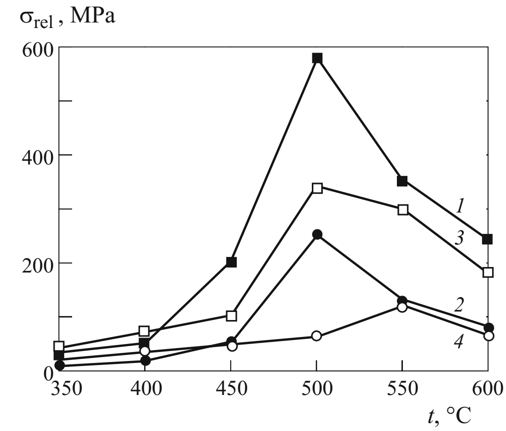

Fig. 4. Dependence of stress relaxation $\Delta\sigma_{rel}$ of steel 08Kh18N10T on the test temperature after treatment in the following modes: *1*, *2*) quenching, deformation at – 196°C at $\varepsilon \le 20\%$, tempering at 500°C; *3*, *4*) quenching, deformation at – 196°C at $\varepsilon \le 20\%$, tempering at 500°C at a load $\sigma_i = 0.5\sigma_{0.2}$, cooling in air; *1*, *3*) tests for relaxation at $\sigma_i = 0.9\sigma_{0.2}$; *2*, *4*) $\sigma_i = 0.5\sigma_{0.2}$.

crease substantially. The yield strength $\sigma_{0.2}$ depends on the degree of deformation, on the temperature, and on the rate of heating after the deformation at – 196°C. The dependences of $\sigma_{0.2}$ on the temperature and on the heating rate have a well manifested maximum (Figs. 2 and 3). The values of the tempering temperature of 500°C and of the heating rate $v_h = 0.175$ K/sec corresponding to this maximum can be treated as optimum parameters of the production process. In this case the strain increases by 20%.

Specimens from the third batch after quenching and plastic deformation at – 196°C were tempered at a stress $\sigma_i = (0.1 - 0.75)\sigma_{0.2}$ at an optimum temperature of 500°C. As a result of such treatment the strength and ductility characteristics improved with respect to those of the third batch due to the processes of strain aging.

The growth in the mechanical characteristics after treatment by mode *3* is connected with relaxation of internal stresses and ordering of the structure [12 – 14].

Metallographic studies showed that the specimens strain-aged at $t < 500$°C (mode *3*) did not contain carbide segregations. Formation of second phase was observed after aging at a temperature equal to or higher than 500°C. The lattice parameter of the steel after strain aging at about 500°C did not change.

Low-temperature plastic deformation and subsequent tempering under stress caused development of dislocation structure in the martensite phase, which was intense in the range of 480 – 520°C [15]. In the process of strain aging the dislocation structure changed actively at the places of the action of local internal stresses that promoted acceleration of diffusion processes in heating; in its turn this caused intense aging of martensite and thus intensified the additional strengthening of the steel [16].

Low-temperature plastic deformation of quenched austenitic steel causes growth of nonuniformly distributed microstresses.

The degree of nonuniformity of the microstresses can be judged upon after active loading is stopped (Fig. 4). The susceptibility to relaxation $\Delta\sigma_{rel}$ is connected with the degree of nonuniformity of the distribution of elastic energy in the material under load and therefore can serve a qualitative characteristic of nonuniformity of internal stresses [17].

Taking into account that the carbon content in the steel is not high, we may expect intense motion of dislocations in the mentioned temperature range. In combination with high dislocation density this promotes creation of conditions for relaxation by microplastic displacements.

It follows from the results obtained that the relaxation effect depends on the process of treatment of the steel and of articles from it [18], i.e., analysis of the results of evaluation of relaxation effect makes it possible to choose the optimum treatment regime.

The processes occurring in the steel in strain aging were also studied by methods of acoustic emission. Since as a result of thermomechanical treatment the content of martensite phase in the studied steel is about 80 – 90%, the most probable source of AE in the elastic deformation range is the dislo-

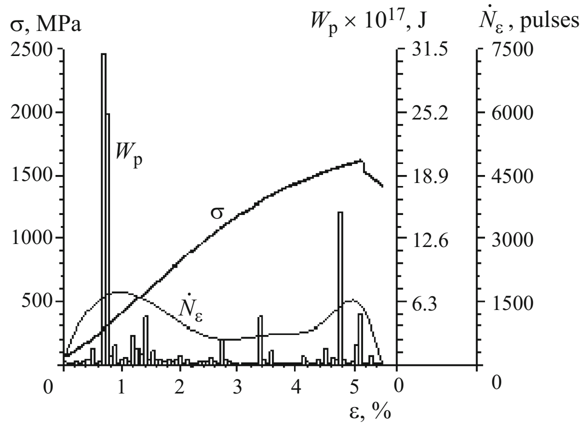

Fig. 5. Dependence of the acoustic emission activity $\dot{N}_\varepsilon$, the mean energy of single pulse W_p, and the stress σ on the degree of tensile deformation at 20°C for steel 08Kh18N10T (treatment by mode *3*, $t_{temp} = 500$°C, $\sigma_i = 0.5\sigma_{0.2}$).

cation process occurring during strain aging of the martensite.

The mean energy of a single pulse detected after deformation of a specimen by Δε was determined by the formula

$$W_p = \frac{W(\varepsilon)}{\dot{N}(\varepsilon)}. \tag{1}$$

Computation showed that $W_p \sim 10^{-16}$ J. The obtained value of a single pulse ranges within $10^{-10} - 10^{-19}$ J, which corresponds to dislocation processes and processes of formation of microcracks.

Figure 5 presents a dependence of the mean energy of single pulse W_p computed by Eq. (1) on the degree of deformation. As a rule, a marked growth in W_p is caused by the activity of AE connected with the nonuniformity of the distribution of internal imperfections in the crystal lattice.

It should be noted that the nonmonotonic behavior of the activity of AE ($\dot{N}_\varepsilon$) is also caused by the change in the strain mechanism in the test process (curve $\dot{N}_\varepsilon$ in Fig. 5) [19]. The value of the activity of AE in the range of elastic strain is the highest in the case of stress tempering at $\sigma_i = 0.1\sigma_{0.2}$ and $\sigma_i = 0.15\sigma_{0.2}$ (Fig. 6). Similar dependences have been obtained for metals and alloys with fcc and bcc structures [20, 21].

The value of the activity of AE in the elastoplastic range decreases monotonically upon growth in the stress.

Such behavior of the activity of AE in the elastic range as a function of the tempering stress is associated with the proportion of the processes of strain aging of martensite, i.e., formation of carbides, growth in the density of mobile dislocations and their fixation by impurity atoms, and relaxation of internal stresses.

The change in mechanical characteristics correlates with the change in some parameters of AE (Fig. 7). The growth in the activity and in the total value of AE in the range of elastic strain at tempering stress $\sigma_i = 0.1\sigma_{0.2}$ and $\sigma_i = 0.15\sigma_{0.2}$ is explainable primarily by intensification of the processes of cracking of carbides at growing dislocation density at the boundaries of the forming carbide phase and increase in the stresses that appear due to mismatch between the initial and segregating phases. It is obvious that the similar behavior of

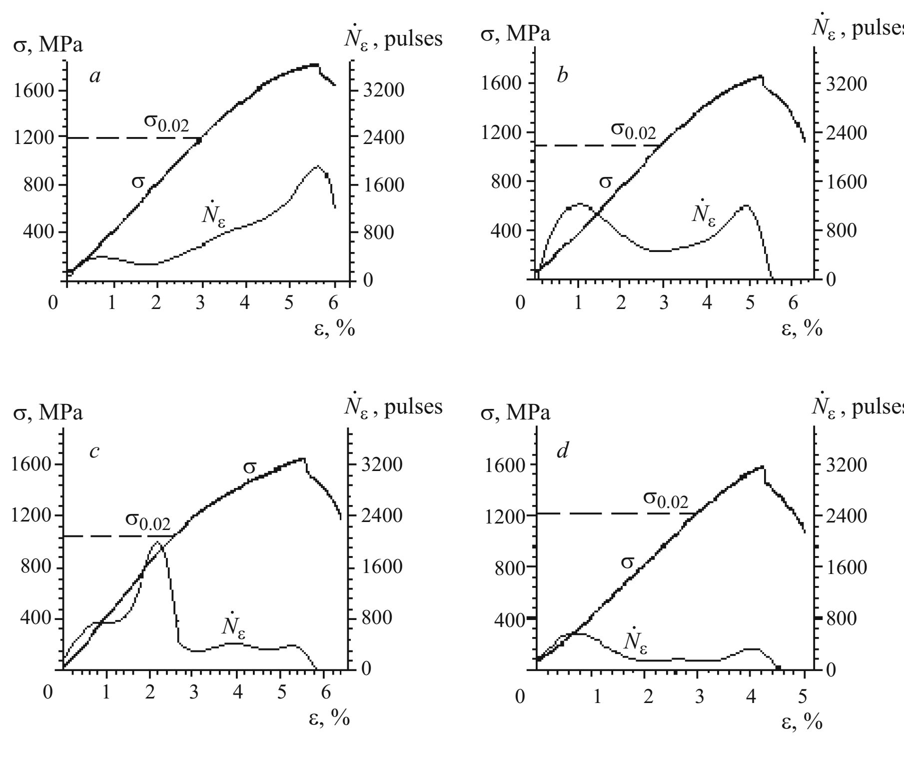

Fig. 6. Dependences of the activity of acoustic emission $\dot{N}_\varepsilon$ and stress σ on the degree of tensile deformation of steel 08Kh18N10T in the elastoplastic range after treatment by mode *3* ($t_{temp} = 500$°C): *a*) without load in tempering; *b*) $\sigma_i = 0.1\sigma_{0.2}$; *c*) $\sigma_i = 0.15\sigma_{0.2}$; *d*) $\sigma_i = 0.2\sigma_{0.2}$.

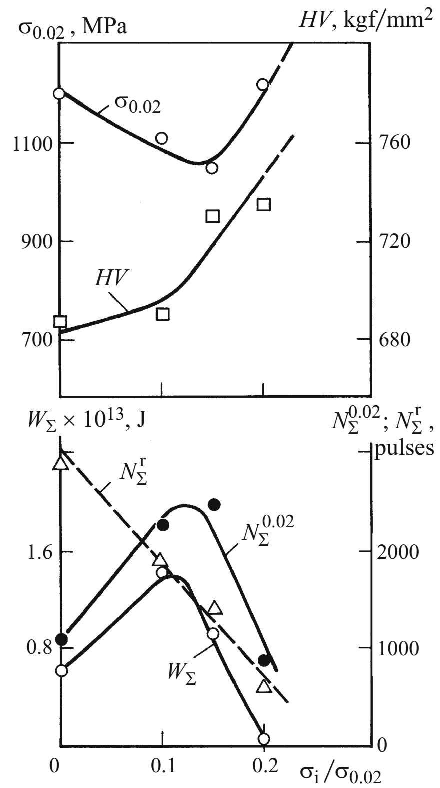

Fig. 7. Dependences of mechanical characteristics and AE parameters of steel 08Kh18N10T on the relative values of load in the process of strain aging at 550°C (treatment by mode *3*): $N_\Sigma^{0.02}$, N_Σ^{r}) total activity of AE in the elastic and plastic ranges, respectively.

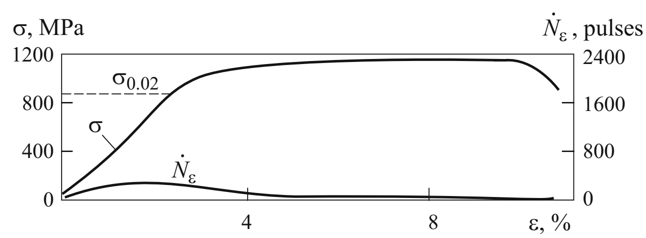

Fig. 8. Dependence of the AE activity and stress on the degree of tensile deformation of steel 08Kh18N10T after strain aging at 650°C (treatment by mode *3*).

the total energy of the AE signal is connected with the same processes. Further growth in the parameters of mechanical characteristics and decrease in the parameters of AE is most probably a result of relaxation of internal stresses in the structure of the material [19].

This finds reflection in the decrease in the value of the relaxation effect $\Delta\sigma_{rel}$ upon growth in σ_i in tempering (Fig. 4). The monotonic growth in the microhardness is connected with the process of relaxation of internal stresses and elevation of the content of martensite phase under conditions of stress growth in tempering (Fig. 7).

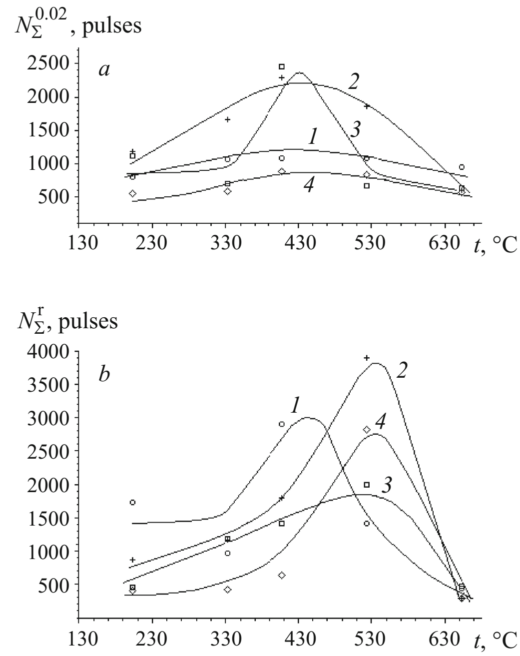

Fig. 9. Dependence of the total AE in elastic ($N_\Sigma^{0.02}$) and plastic (N_Σ^{r}) ranges on the tempering temperature (treatment by mode *3*): *1*) without load; *2*) $\sigma_i = 0.1\sigma_{0.2}$; *3*) $\sigma_i = 0.15\sigma_{0.2}$; *4*) $\sigma_i = 0.2\sigma_{0.2}$.

The decrease in the level of activity of AE in the elastoplastic range as a result of strain aging at 650°C (Fig. 8) as compared to the activity after stress tempering at lower temperatures (Fig. 9*a*) is explainable by finish of the inverse $\alpha \to \gamma$ transformation and, as a consequence, by a change in the mechanism of plastic strain. The high level of AE activity in the other cases seems to be caused by the complex morphology of martensite and the presence of phase boundaries over which considerable internal stresses are concentrated [11, 17]. Under these conditions the change in the activity of AE can be a result of transition of dislocations through the boundaries of breakage of elastic moduli, development of slip lines, relative displacement of grains, formation and development of microcracks and other processes. Relaxation of the mentioned stresses as a result of stress tempering causes monotonic decrease in the activity of AE and in the total amplitude of the AE signals in the elastoplastic range. However, these parameters depend considerably on the processes of nucleation and propagation of cracks. This occurs the most vividly in the temperature range of 410 – 520°C. Here the tendency of lowering of AE upon growth in the stress not exceeding the conventional yield limit at the given temperature can be disturbed, obviously, by redistribution of the roles of ductile and brittle fracture, because the processes brittle fracture are more intense sources of AE. In the temperature range mentioned the stress-strain curves have shorter strain regions corresponding to progressive decrease in the load and preceding failure of the specimen, and σ_r attains a maximum value (Fig. 9*b*).

CONCLUSIONS

1. Growth in mechanical characteristics of austenitic steel plastically deformed at low temperature as a result of tempering under stress by an optimum mode promotes decrease in the relaxation and acoustic parameters.

2. The regularities observed are connected with the stable state of the structure of the treated steel due to relaxation of stresses, growth in the dislocation density and other processes occurring in strain aging of martensite.

3. The variation of the level of activity of acoustic emission $\dot{N}_\varepsilon$ and of its total value N_Σ^{r} has been shown to depend on the tempering parameters (applied stress and temperature). This makes it possible to judge on the presence and value of internal stresses in the studied metal.

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