Temperature dependent mechanical properties and thermal activation plasticity of nanocrystalline and coarse grained Ni-18.75 at.% Fe alloy

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Abstract. Mechanical properties of Ni-18.75 at.% Fe in coarse grained (average grain size 15 µm) and nanocrystalline (average grain size 22 nm) states were studied in uniaxial compression in the temperature range 4.2-350 K. Temperature dependences of the flow stress, strain rate sensitivity and activation volume of plastic deformation were measured. The thermal activation analysis of the experimental data has been fulfilled for the plastic deformation value of 2 %. It was shown that plastic deformation in temperature range from 35 to 350 K in both studied structural states has the thermally activated type. Comparative analysis of low temperature thermal activation plastic deformation was carried out for the alloy in coarse grained and nanocrystalline states. Empirical estimates of parameters of the dislocation interaction with local barriers and internal stress value were obtained for the both studied structural states. Analysis of the results indicates that different mechanisms control the thermal activation plasticity of the Ni-18.75 at.% Fe alloy in coarse grained and nanocrystalline states. Possible mechanisms, which control plasticity of the studied states, are discussed.

1. Introduction
The low-temperature peculiarities and mechanisms of plastic deformation of bulk nanocrystalline (NC) materials with averaged grain size of less than 100 nm are subject of lively discussion [1-17]. The question on the grain size dependence [15,18-19] of the deformation mechanisms controlling the mechanical properties of polycrystalline materials is also deeply discussed in literature. In the work [20] an attempt was made to elucidate the mechanism governing plastic deformation in the NC state for the Ni-18.75%Fe alloy (average grain size is ~ 23 nm) produced by electrodeposition method. It was found that in the entire temperature range 300 - 4.2 K the strength of the alloy in the NC state is an order of magnitude greater than the strength typical for polycrystalline FCC materials (including the strength of coarse-grained (CG) Ni-18.75%Fe alloy). Assuming a thermally activated nature of the process of plastic deformation, the values of the activation volume V of plastic deformation in the first stage of deformation curves have been evaluated in Ref. [20]. The values obtained (V ~ 11b^3 for 300 K and V ~ 3b^3 for 77 K) appeared less by an order of magnitude than those obtained for normal CG FCC materials [1,21].

However, because of the lack of strict theoretical description of the processes occurring during the low-temperature plastic deformation of NC materials, the unambiguous conclusions on the physical mechanisms of plastic deformation for these materials, particularly, for the Ni-18.75%Fe alloy, have not been made yet.

It is known that one of the effective methods of determining the physical mechanisms of plastic deformation in crystalline materials is the method of thermal activation analysis [22,23]. In order to apply the method correctly, the comprehensive experimental data on temperature and strain rate dependences of deforming stress in a wide temperature range are necessary.

At the same time, to reveal the mechanisms controlling the low-temperature plastic deformation of bulk NC materials, it is of great importance to carry out the thermal activation analysis of the process of plastic deformation for a single material but for its different, NC and CG, states.
Therefore, the aim of the present work is to study the regularities of low-temperature plastic deformation of nanocrystalline Ni-18.75 at.% Fe alloy in NC and CG states, to obtain dependence of strain rate sensitivity in a wide temperature range (4.2 – 350 K), and to perform the thermal activation analysis of the data obtained for the process of plastic deformation.

2. Materials and methods
We studied the NC Ni-18 wt.% Fe (18.75 at.% Fe) alloy produced by electrodeposition (average grain size \(d \approx 23\) nm). The CG state of the alloy (\(d \approx 35\) µm) was reached by annealing of the NC state at 800°C for two hours. The procedure of producing the NC alloy as well as investigation of its microstructure are described in detail in Ref. [5,24-26]. The alloy has a FCC lattice, and Fe atoms form a substituted solid solution in the Ni matrix. The content of such impurities as sulfur did not exceed 0.018 at.% (0.01 wt.%) [5].

Samples in the form of rectangular prisms 1.5 × 1.8 × 3 mm in size were deformed by uniaxial compression with initial strain rate \(\dot{\varepsilon}_a = 3.3 \cdot 10^{-4}\) s\(^{-1}\). Stiffness of the deformation machine was about \(7 \cdot 10^6\) N/m. The experiments were carried out at 4.2 K (liquid helium), 77 K (liquid nitrogen), at intermediate temperatures 77 – 290 K (that were obtained by blowing nitrogen through the shower surrounding the sample) and at 4.2 – 77 K (obtained by cooling the sample in helium vapour). The temperatures above 290 K were obtained by heating the sample in hot air.

The obtained diagrams “load-time” were reconstructed into “stress \(\sigma – \) plastic strain \(\varepsilon\)” – \(\sigma(\varepsilon)\) curves, for which we determined the conventional yield stresses \(\sigma_y\), maximal reached stresses \(\sigma_f\) and plastic strain \(\varepsilon_f\) corresponding to these stresses. The magnitude of \(\sigma\) was determined as ratio of load to initial cross-sectional area of the sample, the magnitude of \(\varepsilon\) was calculated as the ratio between the change in sample length due to plastic deformation and its initial length. We also calculated the strain curves in the coordinates of “shear stress \(\tau – \) plastic strain \(\varepsilon\)” – \(\tau(\varepsilon)\). Stress \(\tau = 0.5\sigma\) was chosen as the shear stress. We assume that in the Ni-18.75 at.% Fe alloy the slip system \{111\}<110> of total dislocations, typical for FCC crystals, is operating (the magnitude of the Burgers vector is \(b = 0.25\) nm).

Compressive strain rate jump tests were carried out at different temperatures \(T\) and by increasing the strain rate from \(\dot{\varepsilon}_a = 3.3 \cdot 10^{-4}\) s\(^{-1}\) to \(\dot{\varepsilon}_a = 1.3 \cdot 10^{-3}\) s\(^{-1}\).

To construct the temperature dependence of the strain rate sensitivity \(\Delta \tau/\Delta \ln \dot{\varepsilon}_a\), the data were obtained, in the first stage of strain hardening curve at \(\varepsilon \approx 2\%\). According to the standard formula [22], the values of activation volume for the process of plastic deformation were calculated from the strain rate sensitivity \(\Delta \tau/\Delta \ln \dot{\varepsilon}_a\):

\[
V(T) = kT \frac{\Delta \ln \dot{\varepsilon}_a}{\Delta \tau(T)}
\]

where \(k\) is the Boltzman constant.

Let us note that to exclude dependence of the activation volume on the strain and the strain rate [12-14], the values of \(V\) for all studied temperatures were evaluated by increasing always the same strain rate \(\dot{\varepsilon}_a = 3.3 \cdot 10^{-4}\) s\(^{-1}\) by a factor of 4 at the first stage of the strain hardening curves at the same value of strain \(\varepsilon \approx 2\%\).

3. Experimental results
3.1. Type of strain curves at different temperatures.
Typical deformation curves for the Ni-18.75%Fe are shown in Fig.1 in coordinates “shear stress – plastic strain \(\tau(\varepsilon)\)” in different structural states and temperatures of investigation. The initial stage of the deformation curves obtained in different states has a so-called “parabolic character” smoothly extending into linear region.
Stages of deformation curves as well as significant difference between the values of strain hardening, obtained in NC and CG states, are clearly visible in Fig.1 and in the plots of the dependence of strain hardening rate on the plastic strain $\varepsilon$ (Fig. 2a, NC; 2b, CG).

Fig.1 Typical deformation curves shown in coordinates “shear stress – plastic strain $\tau(\varepsilon)$” for NC and CG states of the Ni-18.75%Fe at different temperatures. The inset shows a jump-like deformation curve for the CG state at 4.2 K.

Fig.2 Dependence of strain hardening rate $d\tau/d\varepsilon$ on deformation for NC (a) and CG (b) states at different temperatures.

Note that the two-stage form of the curves $\tau(\varepsilon)$ is typical for many polycrystalline materials [2]. The main reason for the presence of parabolic hardening in the initial stage of curves $\tau(\varepsilon)$ is associated with the fact that under the influence of high internal stress some of the grains deform before others, and then a gradual spread of deformation from grain to grain [10,19] is followed. The stage of parabolic hardening stops when all the grains are involved in the
process of plastic deformation. However, as can be seen from Fig.2, the length of the parabolic hardening depends on the structural state: for the NC state it is greater than that for the CG.

In the alloy studied, a high plasticity is observed regardless of its structural state (Fig.1). However, the maximal value of plasticity $\varepsilon_f$ can be measured only in the NC state as due to compressive deformation the specimens disintegrated into two parts. In the CG state, the specimens take a barrel-like form in the course of the deformation by compression and do not fail.

The values $\varepsilon_f$ for the NC state are non-monotonic functions of the temperature (Fig.1). In the range 230 - 300 K the failure of specimens took place at the values of plastic deformation $\varepsilon_f \sim 5\text{-}10\%$; with a reduction in the temperature down to 77 - 45 K plasticity considerably increased and eventually attained maximum ($\varepsilon_f \sim 30\%$); with a further decrease in the temperature ($17 - 4.2$ K) the samples failed at $\varepsilon_f \sim 4\text{-}8\%$. The observed increase of $\varepsilon_f$ with a reduction of temperature from room to liquid nitrogen (77 K) may be associated with a rise of twinning activity in the studied NC alloy with decreasing temperature. Note that nucleation of twinning dislocations should proceed much easier (especially at low temperatures) due to the presence of partial dislocations which have less energy of stacking fault in the alloy than in the pure Ni. Decreased plasticity at temperatures under 17 K is obviously associated with a transition from the homogeneous (smooth) character of plastic deformation to that of inhomogeneous one (jump-like).

Note that for the NC the jump-like plastic deformation the in low-temperature range starts immediately after reaching the yield point and continues along the entire deformation curve. For the CG state, the jump-like plastic deformation was observed at 4.2 K, starting from the deformation $\sim 7\%$. The value of jump ($\Delta \tau$) heavily depends on the structural state of the alloy.

For the NC state, at $T = 17$ K, these values are about $\Delta \tau \sim 0.05$ GPa at $\varepsilon \sim 0.5\%$ reaching 0.1 GPa at $\varepsilon \sim 7\%$. For the CG state, at 4.2 K, the values of $\Delta \tau$ at $\varepsilon \sim 7\%$ are two orders of magnitude less ($\Delta \tau \sim 0.001$ GPa). The jump-like plastic flow points to a localized, avalanche-like process of plastic deformation development which is implemented by shift bands. Values of $\Delta \tau$, greater than those obtained in the CG state, suggest that the processes of localized plastic deformation are much more pronounced than in the NC state.

It is significant that high plasticity is observed in the alloy under study irrespective of its structural state (Fig.1). However, at the temperatures below 30 K, in the NC state, when the smooth character of plastic deformation changes into the jump-like, the magnitude of plasticity tends to decreasing.

3.2. Temperature dependences of conventional yield stress $\tau_{02}(T)$, deforming stress $\tau_{0}(T)$ and shear strain rate sensitivity of the alloy in different structural states.

Fig.3 shows the temperature dependence of the conventional yield stress $\tau_{02}(T)$ for different structural states. The increase in the value $\tau_{02}$ with decreasing temperature is a demonstration of the thermally activated character of the process controlling the plastic deformation in the both structural states.

The temperature dependence of the flow stress $\tau_{0}(T)$ and the strain rate sensitivity $\Delta \tau(T)/\Delta \ln \dot{\varepsilon}$, measured at strain $\varepsilon \sim 2\%$ for the NC and CG states, are shown in Fig.4. It should be noted that at the temperatures below 17 K (when the deformation curves take on the jump-like character) measurements of $\Delta \tau(T)/\Delta \ln \dot{\varepsilon}$ have not been conducted. A maximum of $\Delta \tau(T)/\Delta \ln \dot{\varepsilon}$ is observed in the graphs for the both structural states: for NC at $T \approx 300$ K and for CG $\approx 200$ K.

It must be emphasized that the similar maxima on the temperature dependence $\Delta \tau(T)/\Delta \ln \dot{\varepsilon}$ were observed in a number of CG metals and alloys with the FCC and HCP lattices [18,23,27].
Fig. 3 Temperature dependences of yield stress for the NC (□) and CG (●) states.

Fig. 4 Temperature dependences of flow stress $\tau(T)$ (●), strain rate sensitivity of deforming stress $\Delta \tau(T)/\Delta \ln \dot{\varepsilon}_u$ (□) at strain $\varepsilon \sim 2\%$ for NC (a) and CG (b) states of the alloy under study. The solid lines represent the theoretical dependences obtained from the formula (7) (curve 1) and that of (8) (curve 2).

In Fig. 5 are shown the temperature dependences of the activation volume $V$ of the process of plastic deformation, calculated from the formula (1), using the data on the strain rate sensitivity $\Delta \tau(T)/\Delta \ln \dot{\varepsilon}_u$ at strain $\varepsilon \sim 2\%$ for different structural states. It is seen that in the CG state (Fig. 5b) the magnitude of activation volume decreases monotonically from $318b^3$ at $T = 345 K$ to $32b^3$ at $T = 23 K$. For the NC state (Fig. 5a), the values of the activation volume were less by an order of magnitude, $18b^3$ at $350 K$, and decreased monotonically to $7b^3$ with decrease in the temperature down to $T = 35 K$. The values $V$ obtained for the both structural states, NC and CG, agree closely with the values, obtained earlier for the FCC and HCP materials [2,6,28].

The decrease of the activation volume $V$ for the process of plastic deformation with decreasing temperature is characteristic of the thermal activation plasticity.
4. Analysis of experimental results

4.1. Thermally activated plastic deformation

For NC and CG structural states the form of the temperature dependences of the deforming stress $\tau_2(T)$ and strain rate sensitivity $\Delta \tau(T)/\Delta \ln \dot{\varepsilon}_a$ (Fig.4) obtained at $\varepsilon \sim 2\%$ in the temperature range $35 \text{ K} \leq T \leq 350 \text{ K}$ makes possible assuming of thermally activated character of the plastic deformation process for these states. At present time, different microscopical mechanisms of plastic deformation, associated with thermally activated dislocation motion, in NC metals and alloys with grain sizes from 15 to 100 nm, are widely discussed in literature [15,18,29]. Among other mechanisms we can set two main ones. The first mechanism is thermally activated dislocation motion over local barriers within the grains that coincides with a micromechanism of plastic deformation in course-grained materials. The second mechanism is the process of thermally activated nucleation of dislocations at grain boundaries which can be described as follows: a dislocation originated at the grain boundary is curved under the action of stress applied between the points of fixation such as steps at the grain boundary, impurity barriers at grain boundaries, grain triple junctions, etc. If the magnitude of stresses acting on the dislocation is insufficient for extremely forceful separation from the points of fixation, then the release of dislocations can be activated thermally. Strongly curved dislocation loop under the influence of thermal fluctuations at some point overcomes the potential barrier and begins free movement, easily passes through the grain and is absorbed at the opposite boundary. The macroscopic plastic deformation of the sample occurs by movement of a large number of dislocations in many grains.

The two mentioned mechanisms of thermal activation plasticity are based on similar basic acts of plastic deformation: a curved dislocation segment is pinned to the edges of the local barriers and overcomes them due to thermal fluctuations. This allows for the data obtained in the present study to be analyzed by classical thermally activated analysis developed earlier and used to analyze the effects associated with thermally activated overcoming of impurity atoms by dislocations [20,23,30-31].

When describing the thermally activated process of plastic deformation, the interconnection between strain rate $\dot{\varepsilon}$, effective stress $\tau^*$ and temperature $T$ can be represented as a classical Arrhenius relation:
\dot{\varepsilon} = \dot{\varepsilon}_0 \exp \left[ \frac{-H(\tau^*)}{kT} \right], \quad (2)

where \dot{\varepsilon}_0 is the pre-exponential factor; \tau^* = \tau - \tau_i is the effective stress which equals the difference between the deforming stress \tau and the characteristic value of long-range internal stress \tau_i; \ H(\tau^*) is the effective enthalpy of activation.

The \ H(\tau^*) function depends on the parameters of local barriers and the statistical distribution of the barriers by dislocations. In most cases the function \ H(\tau^*) is well approximated by the analytical expression, which explicitly defines the dependence of \ H on \ \tau^* \cite{31}:

\begin{equation}
H(\tau^*) = H_0 \left[ 1 - \left( \frac{\tau^*}{\tau_c} \right)^{p} \right]^{q}, \quad 0 \leq p \leq 1, \ 1 \leq q \leq 2. \quad (3)
\end{equation}

Here, \ H is the energy parameter characterizing the interaction of dislocation with the local barriers; \ \tau_c is the critical stress of activation-free overcoming of local barriers; \ p and \ q are numerical parameters; the value of \ q depends on the shape of the barriers, and parameter \ p is dependent upon the barriers’ properties and their statistics.

The following theoretical expressions can be derived from Eqs. (2) and (3):

\begin{equation}
\tau(T) = \tau_i + \tau_c \left[ 1 - \left( \frac{T}{T_0} \right)^{1/q} \right]^{1/p}, \quad (4)
\end{equation}

\begin{equation}
\left( \frac{\partial \tau^*}{\partial \ln \dot{\varepsilon}} \right)_T = \frac{\tau_c}{pqA} \left( \frac{T}{T_0} \right)^{1/q} \left[ 1 - \left( \frac{T}{T_0} \right)^{1/q} \right]^{(1-p)/p}, \quad (5)
\end{equation}

where \ T_0 = H_0/kA, \ A = \ln(\dot{\varepsilon}_0/\dot{\varepsilon}). \ Eq. (5) is obtained under the assumption that \ \dot{\varepsilon}_0 \ is independent of \ \tau^*.

The above parameters \ \tau_c, \ \tau_i and \ H_0 should be temperature-dependent due to temperature dependence of static constants. In order to simplify the analysis, as in Ref. \cite{20,23,32} we suggest that the temperature dependence of elastic constants leads to the same temperature dependence of parameters \ \tau_c, \ \tau_i and \ H_0:

\begin{equation}
\tau_c(T) = \tau_c \eta(T), \quad \tau_i(T) = \tau_i \eta(T), \quad H_0(T) = H_0 \eta(T) \quad (6)
\end{equation}

where \ \eta(T) is a dimensionless function of temperature, which describes the temperature dependence of the elastic modulus of the material under investigation. The analytical form of this function can be chosen, for example, based on the criterion of optimal approximation registered for the elastic modulus in a given temperature range \cite{23}. Parameters \ \tau_{c0} = \tau_c(0), \ H_{00} = H_0(0) \ are constants, and \ \tau_{i0} \ can, in general, depend on temperature.

Using Eq. (6) and introducing the notation \ T_{oo} = H_{oo}/kA, \ Eqs. (4) and (5) may be represented as follows:

\begin{equation}
\tau(T) = \eta(T) \left\{ \tau_{i0} + \tau_{c0} \left[ 1 - \left( \frac{T}{\eta(T) T_{oo}} \right)^{1/q} \right]^{1/p} \right\}, \quad (7)
\end{equation}
From Eqs. (7) and (8) an expression for parameter \( A \) can be derived:

\[
A = \frac{T}{\eta(T)} \left( \frac{\partial \tau^*}{\partial \ln \dot{\varepsilon}} \right)^{-1} \left( \eta(T) \frac{d\tau}{dT} - \tau(T) \frac{d\eta}{dT} - \eta^2(T) \frac{d\tau_{i0}}{dT} \right),
\]

Eq. (9) allows to analyze the temperature dependence of the parameter \( A \) and to obtain the empirical estimates of its magnitude. An important point is that the temperature independence of the parameter \( A \) is one of criteria of applicability of expressions (2) and (3) for a description of the process of thermally activated plastic deformation [23, 32].

Results of calculation of the parameter \( A \) from Eq. (9) for NC and CG states are shown in Fig. 6. In order to evaluate the value of parameter \( A \) the corresponding experimental dependences \( \tau_{c}(T) \) and \( \Delta\tau(T)/\Delta\ln \dot{\varepsilon} \) were used at \( \varepsilon \approx 2\% \), Fig. 4. For the dependence \( \eta(T) \) an analytical expression described in [20] was also used. It was assumed that the parameter \( \tau_{i0} \) was temperature independent.

![Graph](image)

**Fig. 6** Temperature dependences of parameter \( A = \ln(\dot{\varepsilon}_{i0}/\dot{\varepsilon}) \) for NC (●) and CG (□) states.

It can be seen from Fig. 6, that for the both structural states in the temperature range 55 – 300 K the average values of parameter \( A \) practically coincide and are temperature independent (\( A = 17.55 \) for NC and 17.8 for CG states, respectively). Athermicity of empirical values of parameter \( A \) for the NC state as well as for the CG one is evidence of the validity of the assumption on the temperature independence of parameter \( \tau_{i0} \) and lends support to the applicability of the Eqs. (2) (where \( \dot{\varepsilon}_{0} = \text{const} \)) and (3) to describe the process of thermally activated plastic flow for different structural states.

We compared experimental temperature dependences \( \tau_{c}(T) \) and \( \Delta\tau(T)/\Delta\ln \dot{\varepsilon} \) (Fig. 4) to theoretical dependences (7) and (8) and conducted joint approximation of these functions by least squares method, which allowed in self-consistent manner to find the values of all theoretical parameters: \( p, q, \tau_{i0}, H_{oo}, T_{oo}, S_{o} \). The corresponding values of evaluated theoretical parameters age given in the first row of Table. Experimental dependences of flow
stress $\tau_0(T)$, strain rate sensitivity $\Delta \tau(T)/\Delta \ln \dot{\varepsilon}_\alpha$, and their approximate theoretical curves obtained by means of Eqs. (7) and (8), and by using the values of theoretical parameters from the Table are shown in Fig.4 for different structural states.

Table. Values of theoretical parameters for NC and CG states of Ni-18.75%Fe alloy, obtained by an analysis of experimental data.

|       | $p$  | $q$  | $\tau_{i0}$, GPa | $\tau_{c0}$, GPa | $H_{oo}$, eV | $T_{oo}$, K | $A$  |
|-------|------|------|------------------|------------------|-------------|------------|------|
| NC    | $2/3$| 1,12408 | 0,99005          | 0,67478          | 0,787       | 514,1      | 17,76 |
| CG    | $2/3$| 1,71207 | 0,13940          | 0,07059          | 0,669       | 442,7      | 17,55 |

Under the above assumptions, the following theoretical dependence of the activation volume of plastic deformation on temperature is derived from Eqs. (3) and (8):

$$V(T) = -\frac{\partial H}{\partial \tau} = \frac{pqH_{oo}T}{\eta(T)\tau_{c0}T_{oo}} \left( \frac{T}{\eta(T)T_{oo}} \right)^{1/q} \left[ 1 - \left( \frac{\tau}{\tau_{c0}} \right) \right]^{1/q} \frac{1}{(p-1)/p}. \quad (10)$$

The theoretical dependences (10) for NC and CG states calculated with the use of the values from Table are represented in Fig.5 by solid lines. It is apparent that in the entire temperature range $23 \text{ K} \leq T \leq 350 \text{ K}$ the theoretical dependence adequately approximates experimental data.

Thus, the data presented in Figs.4 and 5 show a good agreement between theoretical and experimental results over the entire range $23 \text{ K} \leq T \leq 350 \text{ K}$, which is an additional confirmation of adequacy of the theoretical model used. In addition, based on a set of data obtained, it can be concluded, that in the CG as well as in the NC states the process of thermally activated plastic deformation over the investigated temperature range may be described by one mechanism. Let us try to understand the difference between the mechanisms for the CG and NC states.

4.2. Discussion of mechanisms of thermally activated dislocation motion for CG and NC states.

The main differences between the mechanical properties of NC and CG states of the studied alloy are as follows:

(i) The magnitudes of the yield stress in the NC state are 10 times those in the CG state over the entire temperature range under study. With decreasing temperature from $300 \text{ K}$ down to $35 \text{ K}$ the magnitude of $\tau_{o2}$ increases by $\sim 1.5$ times for the both states.

(ii) The empirical values of internal stresses in NC alloy are about 7 times those obtained for CG alloy (under the assumption that this empirical value is temperature independent).

(iii) At the same temperature the process of deformation in CG and NC alloys goes on at different magnitudes of effective stresses. The magnitude $\tau^*$ in NC alloy is considerably greater than that in CG one: at $T = 300 \text{ K}$ the ratio between them is 36. With decrease in the temperature down to $35 \text{ K}$, the ratio monotonically decreases to 13.

(iv) At the same temperature the magnitude of the activation volume $V$ in the CG state is much greater than in the NC one. For $350 \text{ K}$ the activation volume $V = 318 b^3$ and $V = \ldots$
18b° for CG and NC states, respectively. There is also a considerable difference between
temperature dependences of the magnitude V. For the CG state a tenfold decrease in V (from
318b° to 32b°) is observed when temperature changes from 350 K down to 35 K. For the NC
alloy V shows more weak dependence on the temperature. With reducing temperature from
350 K down to 35 K, the values of V decrease ~ 2.5 times (from 18b° down to 7b°).
As the processes of plastic deformation in CG and NC states proceed at equal temperatures
but under different effective stresses, a comparison of the dependences V(τ°) would be
appropriate. Using the relation (3) results in the expression describing the dependence V(τ°)

\[
V(\tau°) = -\left( -\left( \frac{\partial H}{\partial \tau°} \right)_T \right) = \frac{pqH_0}{\tau°_c} \left( \frac{\tau°}{\tau°_c} \right)^{p-1} \left[ 1 - \left( \frac{\tau°}{\tau°_c} \right)^p \right]^{q-1} .
\]  

(11)

These dependences are shown in Fig.7 in double logarithmic coordinates. Theoretical
dependences V(τ°) derived from Eq. (11) for the temperature range 23 ≤ T ≤ 350 K are shown
by solid lines. Formal extrapolation of theoretical dependences V(τ°) to temperature range
23 ≤ T for the CG alloy and T ≤ 350 K for the NC alloy are shown by dashed lines.
Experimental data are indicated by dark and light symbols. It is evident from the figure that
these curves correspond to two different dependences V(τ°).
Different empirical values of parameters H_00 and q, given in Table, also point to a distinct
nature of local barriers controlling the process of thermally activated plastic flow. In the NC
state (q = 1.12408), the effective force barrier is steep and its shape is close to a square, but in
the CG state (q = 1.71207), the effective force barrier has more gently descending slopes.
Since in our case ethne values of the parameter p coincide (p = 2/3), then according to Eq.
(11), the dissimilarity of empirical values of parameter q for different states brings to different
functional dependence of the activation volume V on the dimensionless argument τ°/τ°_c, i.e. to
different dependences of V(τ°). Thus, the essential difference between the magnitudes of
activation volumes as well as their dependences on the temperature and effective stress
allows to assume that the microscopical mechanisms, controlling kinetics of thermally
activated plastic flow in CG and NC states, are of different physical nature.
alloy are shown by dashed lines; ○, O are experimental values for the CG and NC states of the alloy, respectively.

Based on the above data, we could estimate average area $S_0$ which corresponds to one local barrier for different structural states. The coincident values of the parameter $p = 2/3$ obtained for the CG and NC states agree with the Friedel statistics [33], describing the dislocation motion through a net of the randomly arranged local barriers.

For the Friedel statistics the length of dislocation segment $L_0$ that bends under action of the stress $\tau^*$ between two local barriers, can be expressed as:

$$L(\tau^*)=(2E_d/\tau^*)^{1/3},$$  \hspace{1cm} (12)

where $E_d$ is a linear tension of dislocation, $S_0$ is an average area per one local barrier. Let us express the activation volume in the terms of:

$$V = bwL,$$  \hspace{1cm} (13)

where $w = w(\tau^*)$ is an effective activation distance, characterizing the process of microscopical thermally activated plastic deformation caused by thermally activated dislocation motion through a system of local barriers.

Using Eqs. (12) and (13), we obtained the expression for the parameter $S_0$, that makes it possible to evaluate the density of local barriers, controlling the dislocation motion:

$$S_0 = \frac{\tau V^3}{2E_db^2w^3},$$  \hspace{1cm} (14)

Using Eq. (14) and standard estimate $2E_d = Gb^2$, $S_0$ can be expressed as:

$$S_0 = \frac{b^2\tau^*}{G}\left(\frac{b}{w}\right)^3\left(\frac{V}{b^3}\right)^3,$$  \hspace{1cm} (15)

Substituting the obtained values of parameters (at $T = 290$ K) for $w \geq b$ in Eq. (15) gives the following values of $S_0$ for different structural states:

$S_0 \leq 5.205 \cdot 10^{-17}$ m$^2$ ~ 830 b$^2$ (CG) and $S_0 \sim 9.48 \cdot 10^{-19}$ m$^2$ ~ 15b$^2$ (NC)

For different structural states were found significant differences between the magnitudes of activation volumes, their dependences on temperature and effective stress, and the values of the average area of one local barrier. On the basis of the fulfilled analysis, it can be concluded that the microscopical mechanisms, governing the kinetics of thermally activated plastic flow in the CG and NC states, have different physical nature.

In the NC state (with a large volume fraction of grain boundaries), the most likely controlling mechanism is a thermally activated process of generation of dislocations at grain boundaries (i.e. the release of dislocations from the points of fixation). In the CG state, the process of plastic deformation is controlled by a thermally activated dislocation motion through intragranular local barriers.

5. Conclusions

The peculiarities of plastic deformation of Ni-18.75 at.% Fe alloy were studied in a wide range of low temperatures (4.2 – 350 K) under uniaxial compression and in two states: NC and CG. For these states the temperature dependences of conventional yield stress were obtained and deformation curves were analyzed. Temperature dependences of deforming stress and strain rate sensitivity during plastic deformation of $\varepsilon \approx 2\%$ were studied in detail. Thermal activation analysis of the temperature dependences was carried out and parameters of plasticity and temperature dependences of the activation volume of plastic deformation were obtained. Empirical estimates of the parameters, characterizing the kinetics of macroscopical plastic deformation, and estimates of internal and effective stresses were obtained for different structural states.
The results of comparison analysis of plastic deformation parameters for different structural states revealed significant differences between the magnitudes of activation volumes, their dependences on the temperature and effective stress as well as the values of average area per one local barrier. The results suggest different microscopical mechanisms controlling kinetics of thermally activated plastic flow in the CG and NC states.

In the NC state the most probable controlling mechanism will be a thermally activated process of separation the dislocation from fixation points at grain boundaries. However, in the CG state the plastic deformation is controlled by a thermally activated process of motion of dislocations through the intragranular local barriers.

5. References
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