Manifestation of the Portevin–Le Chatelier Effect in the Kh20N80 Alloy

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Abstract—It has been found that the single-phase nickel alloy Kh20N80 demonstrates discontinuous flow in the range of deformation temperatures of 300–650°C. It is shown that the discontinuous flow of this alloy possesses signs of the Portevin–Le Chatelier effect, strong localization of deformation, and negative strain-rate sensitivity of the flow stress. Furthermore, three specific features of the discontinuous flow of the Kh20N80 alloy have been revealed: (1) the "normal" dependence of the critical degrees of the onset of the discontinuous flow of ε_{cr} on the strain rate; (2) the "inverse" temperature dependence of ε_{cr}; and (3) the positive temperature dependence of the flow stress. These special features appear to be caused by the formation of a short-range-order structure, which occurs in this alloy at temperatures of approximately 350–500°C.

INTRODUCTION

As is known, discontinuous flow (DF) (or serrated yielding) is a complex phenomenon connected with the periodic transitions between the macro-uniform and macro-non-uniform deformation [1] which are caused by the nucleation and propagation of deformation bands. In turn, the macro-localization of deformation is determined by different factors, including microstructural processes [2–6]. Since DF is observed in wide temperature and strain-rate ranges in different alloys and in pure metals, several micromechanisms which lead to the appearance of DF are usually distinguished [3]. The basic mechanism is known as the Portevin–Le Chatelier (PLC) effect; it consists in the dynamic strain aging of solid solutions, in which a periodic blocking of dislocations by atmospheres of impurity atoms and their unblocking occur [7]. It is known that the PLC effect is characterized by deformation localization and by a negative instantaneous strain-rate sensitivity of the flow stress. The PLC effect is frequently accompanied by acoustic emission [8]. The so-called "pseudo-PLC effects" are also distinguished, which are observed in the alloys, in which the precipitation of particles or the formation of a short-range-order structure occurs [6, 9]. In such cases, there can be observed anomalous dependences of the critical degrees of the start of discontinuous flow on the temperature and the strain rate [6].

In the literature, there exist numerous investigations of DF and, in particular, of the PLC effect in various aluminum [1, 7, 12] and copper [8] alloys, steels [13], and some intermetallic compounds [14–15]. In nickel superalloys, the phenomenon of DF has not been studied in detail [16, 17]. The complexity of studying DF and of determining the mechanisms of its appearance in multiphase nickel alloys is connected with the fact that their deformation behavior is significantly affected by the precipitation of strengthening second-phase particles. The role of coherent particles of a strengthening γ'' phase and its parameters in the manifestation DF was studied in detail in [17] on the example of an Inconel 718 alloy; the authors made a conclusion that DF is most likely to be connected with the superposition of deformation mechanisms such as passing through (cutting) or bypassing of particles by dislocations. However, the true reasons for DF in these alloys have not been clearly understood, since the role of the γ matrix containing no second-phase precipitates has not been clarified. In this connection, of great interest is any information on the deformation behavior and on the possibility of the manifestation of DF in the γ matrix of nickel alloys, which is represented by the single-phase alloy Kh20N80 (Nichrome). At present no such information exists in the literature. The study of the phenomenon of DF in Nichrome and an analysis of the mechanisms of its appearance with allowance for the well-known fact of the formation in Nichrome of short-range order (SRO) at definite temperatures [18, 19] will make it possible to estimate the influence of the solid solution on the DF in nickel alloys separately from the action of a mechanism such as the cutting of particles by dislocations.

This work is aimed at the study of the manifestations and an analysis of the possible mechanisms of discontinuous flow in the Kh20N80 alloy.

EXPERIMENTAL

The alloy Kh20N80 utilized in the work has the following chemical composition: Ni–21% Cr–0.6% Si–0.3% Mn–0.75% Fe–0.31% Al–0.08% Ti–0.35% Cu–0.05% C. To obtain a uniform coarse-grained structure with an average size of grains of 100 μm, we used a hot-rolled bar with a diameter of 40 mm and annealed it at 1025°C for 2 h.

The mechanical tests (by upsetting samples of the alloy with dimensions of $\varnothing$ 10 × 15 mm) were conducted on a Schenck universal dynamometer in wide temperature and strain-rate ranges with a step of 50 K (T = 300–700°C, $\dot{\varepsilon} = 10^{-5}$–$5 \times 10^{-2}$ s^{-1}). The degree of deformation of the samples was approximately 30%. To guarantee the uniformity of deformation during upsetting via a decrease of frictional forces, we used a graphite lubrication. The effect of the "barreling" of the deformed samples was insignificant. The coefficient m of the strain-rate sensitivity of the flow stress was determined as

$$m = d\log\sigma / d\log\dot{\varepsilon}, \qquad (1)$$

where $d\log\sigma = \log(\sigma_2/\sigma_1)$ and $d\log\dot{\varepsilon} = \log(\dot{\varepsilon}_2/\dot{\varepsilon}_1)$, and the stresses σ_2 and σ_1 were obtained from the strain–stress curves at ε = 20% and strain rates $\dot{\varepsilon}_2$ and $\dot{\varepsilon}_1$, respectively.

The electron-microscopic studies of the samples were conducted on a JEM-2000EX transmission electron microscope at an accelerating voltage of 160 kV. The foils were prepared by the method of jet polishing, using a 10% solution of perchloric acid in butanol as the electrolyte.

For studying the deformation relief, the samples with dimensions of 6 × 5 × 3 mm after electropolishing were upset in a vacuum in a special vacuum system of the IMASh 20-78 type[1] to the degree of deformation of 16 and 5%, respectively, at temperatures of 500 and 650°C at a strain rate of 7×10^{-4} s^{-1}. An indirect proof of the uniformity of deformation during upsetting is the fact that no essential difference in the deformation relief has been revealed between the center of sample and the surface of the butt end.

RESULTS

Temperature–Strain-Rate Range of the Manifestation of Discontinuous Flow

It was discovered during the mechanical tests that the deformation behavior of the Kh20N80 alloy is characterized by discontinuous flow. It manifests itself in the appearance of serrations in the flow-stress σ–degree of deformation ε curves. Figure 1a displays the temperature–strain-rate region corresponding to the discontinuous flow of the alloy. It is within the limits of temperatures of 300–650°C at strain rates $\dot{\varepsilon} = 10^{-5}$–$5 \times 10^{-2}$ s^{-1}.

The σ–ε curves demonstrate intensive strengthening up to high degrees of deformation. An analysis of the σ–ε curves showed the presence of different types of serration, which change depending on the temperature and the rate and degree of deformation (Fig. 1b). It is possible to distinguish isolated serrations (type A) and serrations with a wide plateau (type B). It is necessary to note that the amplitude of the decrease in the flow stress $\Delta\sigma$ at the serrationteeth grows with increasing temperature and decreasing strain rate and reaches 20 MPa. At low temperatures and high strain rates, there is observed only a slight waviness of the stress (curve *2*); with increasing temperature and decreasing deformation rate, a pronounced serration appears in the curves (curve *3*).

The fall of stress at each tooth was usually accompanied by acoustic emission. But acoustic emission is audible only in a temperature range of 450–650°C. It is interesting that with increasing temperature and decreasing strain rate the intensity of sound increases. At temperatures of 300–400°C, no sounds were audible in the entire strain-rate interval investigated.

Critical Degree of Deformation

The appearance of serrations in the curves is observed not immediately after the beginning of deformation, but after the deformation reaches a certain degree which was called the critical degree of deformation ε_{cr} [5]. The critical degree of deformation depends on the temperature and rate of deformation (Fig. 2). As is known, usually ε_{cr} increases with increasing strain rate and decreasing temperature of deformation [5]. This behavior is called "normal." But, frequently, "inverse" dependences of ε_{cr} on the temperature and rate of deformation are encountered [6].

As can be seen from Fig. 2, for the alloy under investigation there occurs an increase in ε_{cr} with increasing deformation rate, i.e., a "normal" strain-rate dependence of the critical degree of deformation is characteristic. But the change in ε_{cr} with increasing deformation temperature is nonmonotonic, which is usually called "inverse" dependence. However, in the dependences given in Fig. 2b it is possible to separate regions of the "normal" and "inverse" behavior. It is seen that the temperature intervals of "inverse" behavior are different for different deformation rates, and are approximately equal to 50 K, e.g., 350–400°C for $\dot{\varepsilon} = 10^{-5}$ s^{-1}; 400–450°C for $\dot{\varepsilon} = 7 \times 10^{-4}$ s^{-1}; and 450–500°C for $\dot{\varepsilon} = 10^{-2}$ s^{-1}. Note also that, with increasing rate of deformation, the regions of the "inverse" behavior of $\varepsilon_{cr}(T)$ are shifted toward higher temperatures and are within the limits of T = 350–500°C.

[1] These investigations were performed in cooperation with M.Kh. Mukhametrakhimov.

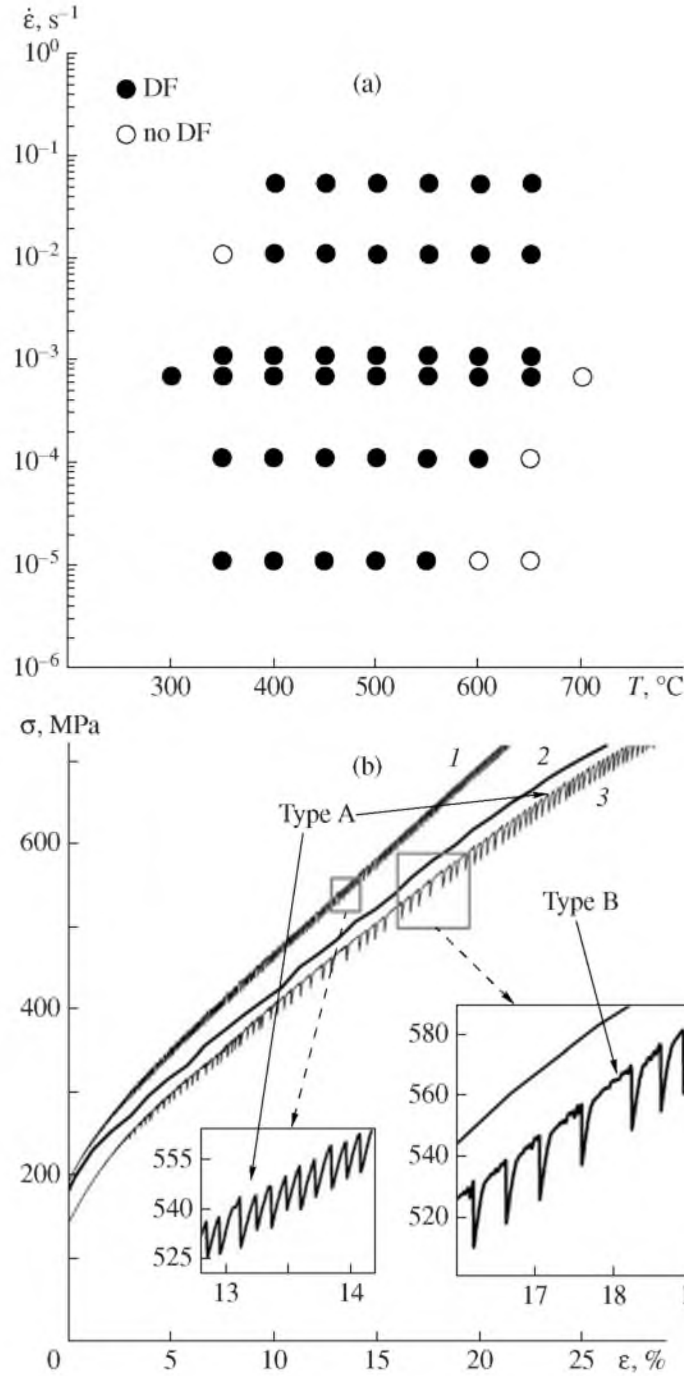

Fig. 1. (a) Temperature–strain-rate region of the manifestation of the discontinuous flow of the Kh20N80 alloy and (b) the characteristic forms of serration in the flow-stress–degree-of-deformation curve of the Kh20N80 alloy: (*1*) $T = 400°C$, $\dot{\varepsilon} = 10^{-4}$ s^{-1}; (*2*) $T = 500°C$, $\dot{\varepsilon} = 10^{-2}$ s^{-1}; and (*3*) $T = 500°C$, $\dot{\varepsilon} = 10^{-5}$ s^{-1}.

Dependence of the Flow Stress on Temperature and Deformation Rate

Figure 3a presents the dependences of the flow stress at the degree of deformation $\varepsilon = 20\%$ and at different rates of deformation $\dot{\varepsilon}$ (10^{-5}, 7×10^{-4}, 10^{-2} s^{-1}) on the deformation temperature. An analysis of these dependences at the intermediate strain rate 7×10^{-4} s^{-1} shows that at temperatures of 450–550°C there is observed a positive temperature dependence of the flow stress; i.e. σ_{20} manifests an anomalous increase (to 30%) with an increase in the deformation temperature and reaches a peak at $T = 550°C$. In the temperature

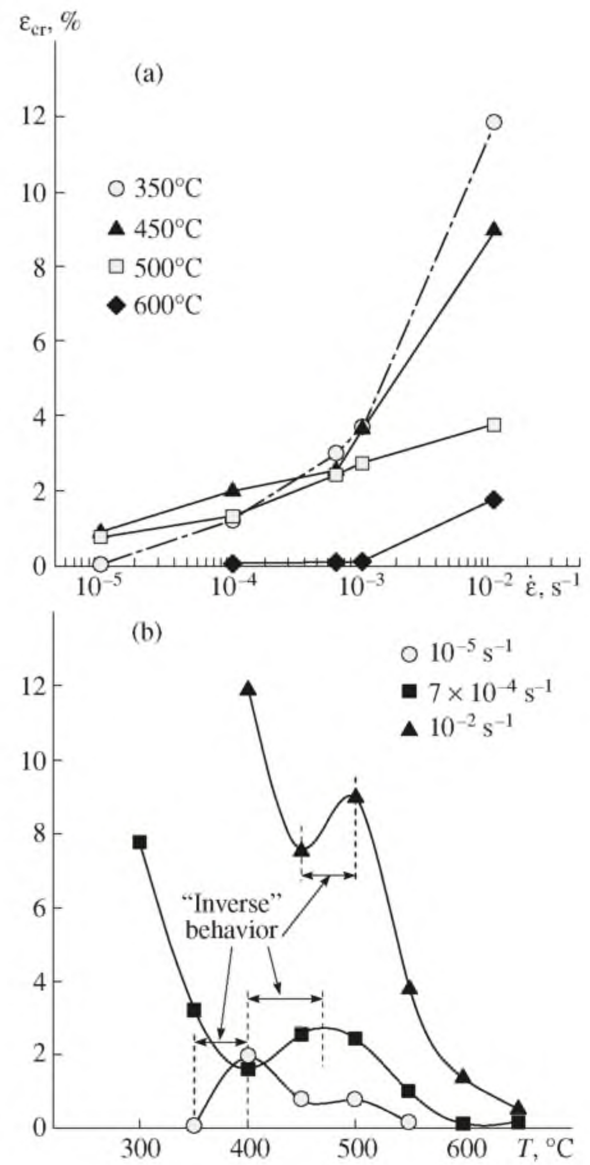

Fig. 2. Variation of the critical degree of deformation ε_{cr} required for the start of discontinuous flow with increasing (a) rate of deformation and (b) temperature of deformation.

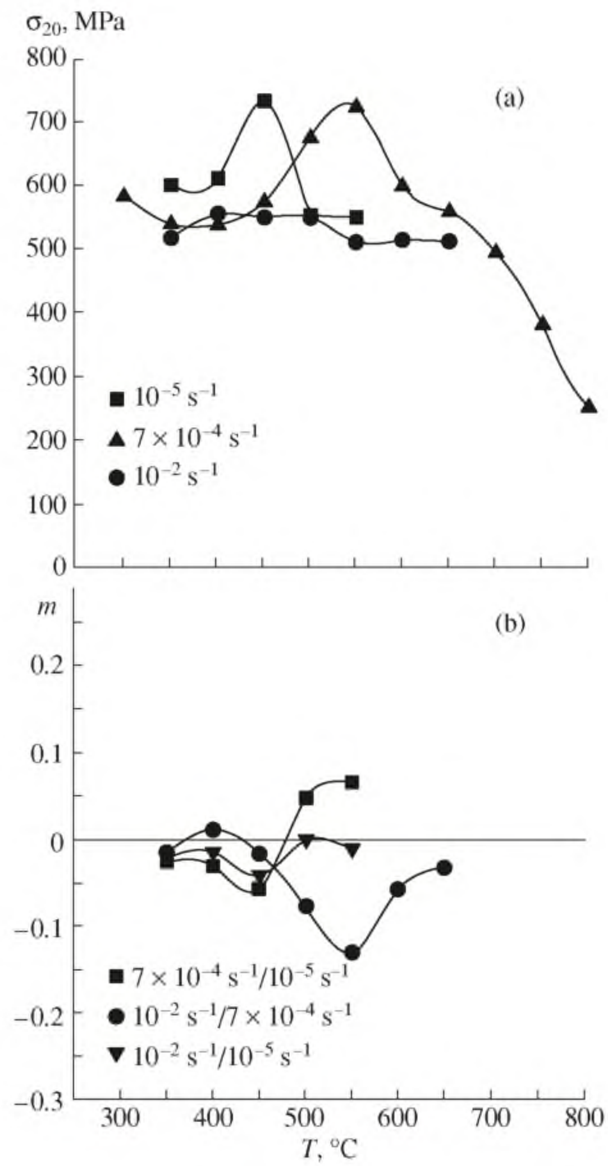

Fig. 3. (a) Dependence of the flow stress σ_{20} of the Kh20N80 alloy on the temperature of deformation and (b) the coefficient m of the strain-rate sensitivity of the flow stress at different temperatures and rates of deformation.

range of 550–650°C, the level of σ_{20} begins to be reduced, but it remains higher than at a temperature of 400°C. An analogous behavior of the $\sigma_{20}(T)$ dependence is also observed at the minimum strain rate of 10^{-5} s^{-1}. It should be noted that, in this case, an anomalous increase in the flow stress begins at a temperature of 400°C, and stress peak corresponds to the temperature of 450°C. As in the case with the $\varepsilon_{cr}(T)$ dependence (Fig. 2b), here also it is possible to observe a shift of the $\sigma_{20}(T)$ peak toward higher temperatures with an increase in the deformation rate.

Another type of dependence is observed at the high rate of deformation (10^{-2} s^{-1}); in this case, no clearly pronounced peak exists in the curve, but the level of the flow stress at T = 400–500°C is again higher than at T = 350°C.

An unusual dependence of σ_{20} on the deformation rate was also revealed, which can be seen in both Fig. 3a and Fig. 3b, where it is shown that the coefficient m is close to zero or even is negative.

Electron-Microscopic Study of the Alloy Structure

Figures 4a–4d display photographs of the microstructure of the Kh20N80 alloy and of the relief which appears at the lateral surfaces of the samples after deformation under temperature–strain-rate conditions

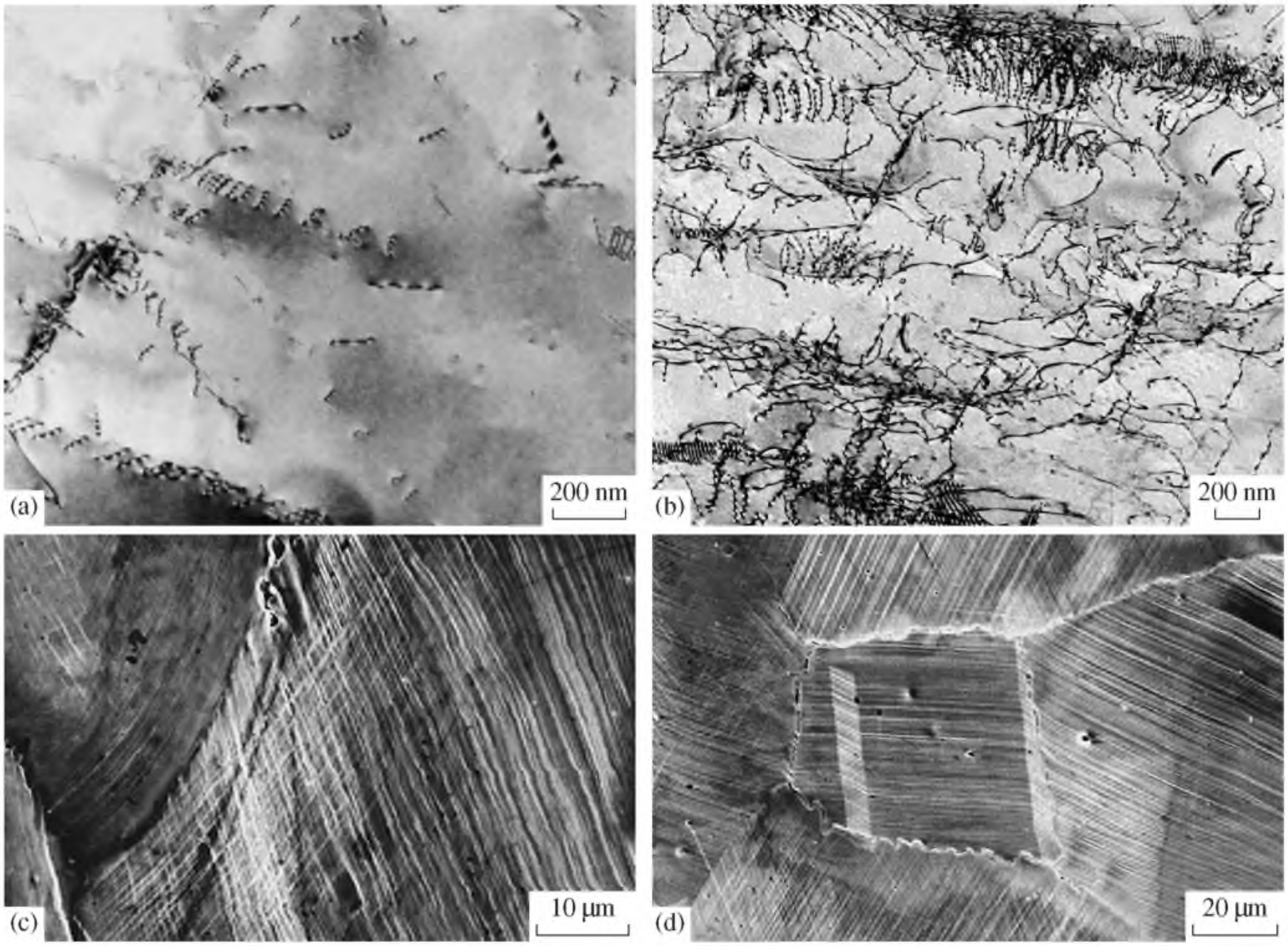

Fig. 4. (a, b) Fine structure of the Kh20N80 alloy after deformation at $T = 500°C$, $\dot{\varepsilon} = 7 \times 10^{-4}\ s^{-1}$ to (a) $\dot{\varepsilon} = 3\%$ and (b) $\varepsilon = 10\%$; and (c, d) deformation relief after deformation at (c) $T = 500°C$, $\dot{\varepsilon} = 16\%$, $\dot{\varepsilon} = 7 \times 10^{-4}\ s^{-1}$ and (d) $T = 650°C$, $\varepsilon = 5\%$, and $\dot{\varepsilon} = 7 \times 10^{-4}\ s^{-1}$.

which correspond to the manifestation of DF. An analysis of the fine structure (Figs. 4a, 4b) demonstrates that an enhanced dislocation density is characteristic of the deformed state. The dislocations form wide planar pileups, which is typical of materials with a low stacking-fault energy.

The study of the deformation relief (Figs. 4c, 4d) showed that during deformation under conditions corresponding to the development of DF there occurs a strong localization of plastic deformation on some crystallographic planes. In some grains there are distinctly visible rough slip bands, which develop predominantly in one direction and pass through the entire grain. The distance between the bands is approximately 1 μm. In other grains there are present second-slip lines located at an angle of about 60° to the primary bands, which indicates the action of a second system of intragranular dislocation slip. At a temperature of 500°C, the bands are rougher and become wavy as a result of the passage of secondary slip bands. There is also observed a pronounced waviness of grain boundaries.

DISCUSSION

The above results of the investigation of the deformation behavior and microstructural changes in the Kh20N80 alloy indicate that the revealed discontinuous flow of this alloy is characterized by a number of signs of the PLC effect which were observed earlier in other materials [1, 7, 8, 10, 11], namely,

—DF is observed in a specific temperature range (for this alloy, at temperatures of 350–650°C at strain rates of 10^{-5}–$5 \times 10^{-2}\ s^{-1}$) and is accompanied by acoustic emission;

— the flow stresses are only weakly sensitive to the deformation rate; the coefficient of strain-rate sensitivity m is close to zero or is negative;

— the plastic flow is strongly localized in deformation bands.

The signs of the PLC effect revealed in the Kh20N80 alloy make it possible to assume that on the microlevel the basic mechanism of plastic deformation which leads to the appearance of unstable flow is dynamic strain aging, i.e., a periodic blocking and unblocking of dislocations by solute atoms of the alloying element, i.e., by chromium [8, 12].

This conclusion agrees well with our previous results [24], where it was shown that the deformation behavior of the Kh20N80 alloy at temperatures lower than 650°C, i.e., lower than the upper temperature boundary of the manifestation of DF, can be considered as cold deformation; i.e., in this temperature interval the processes of a redistribution of dislocations are strongly hindered, which favors blocking dislocations. But at temperatures higher than 650°C the activation of the processes of dislocation redistribution due to climb prevents their blocking by impurity atoms, which results in the disappearance of this effect.

Besides the signs of the PLC effect enumerated above, the discontinuous flow of the alloy under investigation possesses three additional specific features:

(i) First, the $\varepsilon_{cr}(\dot{\varepsilon})$ dependence is "normal"; i.e., ε_{cr} grows with increasing deformation rate.

(ii) Second, the $\varepsilon_{cr}(T)$ dependences are nonmonotonic; they are divided into segments of "normal" and "inverse" behavior; the temperature intervals of the "inverse" behavior, in which ε_{cr} increases with increasing temperature, are different for different deformation rates and are within the limits of a temperature range of 350–500°C.

(iii) Third, the flow stresses demonstrate a positive temperature dependence in the region of temperatures of 350–550°C.

These specific features of DF in the Kh20N80 alloy, obviously, cannot be explained only by dynamic strain aging, i.e., by true PLC effect. Apparently, the DF in this alloy is substantially affected by the development of an SRO structure in a specific temperature interval. The appearance of an SRO in the Kh20N80 alloy, just as the precipitation of particles in other alloys [6, 9], can result in an instability of flow. On the one hand, dislocation slip through ordered zones leads to the destruction of atomic order, which decreases stresses. On the other hand, as a result of diffusion there are again formed regions of local ordering, which lead to an increase in stresses. The interaction of these processes of strengthening–softening can lead to the appearance of serrations in the diagram of deformation and, probably, to an enhancement of the effect of dynamic strain aging.

The anomalous strengthening of the Kh20N80 alloy at temperatures of 350–550°C is a direct consequence of the arising SRO structure of the Ni_2Cr type, which was mentioned earlier [18–23]. As is known, the local ordering characteristic of nichromes causes an increase in strength, microhardness, and electrical resistance.

According to [6], the combination of the "normal" $\varepsilon_{cr}(\dot{\varepsilon})$ dependence and "inverse" $\varepsilon_{cr}(T)$ dependence is characteristic of alloys in which there occurs a homogeneous precipitation of particles, which leads to an essential strengthening. In our case, the analogous character of these dependences for the Kh20N80 alloy agrees well with the conclusions of [6] and confirms the presence of processes of additional strengthening due to SRO. In particular, an increase in ε_{cr} with increasing temperature in specific intervals which are located within the temperature range of 350–500°C shows that under given conditions of deformation there are required higher flow stresses for the appearance of DF and consequently, for the destruction of regions of atomic order by moving dislocations.

Of significant importance is the fact that the strain rate exerts a substantial influence on the temperature interval of the appearance of SRO. In particular, as the strain rate increases, the temperature intervals of the "inverse" behavior of $\varepsilon_{cr}(T)$ and also the peaks of a positive dependence of $\sigma_{20}(T)$ are shifted toward higher values. Consequently, it can be assumed that with increasing deformation rate the temperature range of the appearance of SRO will be shifted toward higher values as well.

Based on the above data concerning the anomalous behavior of the flow stress and critical strain rates required for the onset of DF, it can be assumed that the overall temperature range of the appearance of SRO in the Kh20N80 alloy at all employed deformation rates is 350–500°C. However, the region of the manifestation of the PLC effect is somewhat wider: it is 300–650°C. Thus, the appearance of DF in the Kh20N80 alloy is substantially affected, apart from the basic mechanism—dynamic strain aging,—by the formation of an SRO structure, and it is precisely SRO that causes the specific features of the PLC effect revealed in the alloy under study.

CONCLUSIONS

It has been found that the single-phase nickel alloy Kh20N80 (Nichrome) demonstrates discontinuous flow in the range of deformation temperatures of 300–650°C at strain rates of $\dot{\varepsilon} = 10^{-5}$–$5 \times 10^{-2}$ s^{-1}. The signs of the Portevin–Le Chatelier effect revealed in the alloy behavior, i.e., a strong localization of deformation and a negative strain-rate sensitivity of the flow stress, make it possible to assume that it is the dynamic strain aging that is the basic mechanism of the appearance of discontinuous flow.

In addition, three specific features of the discontinuous flow of the Kh20N80 alloy have been revealed, namely, (1) the "normal" dependence of the critical degrees of deformation ε_{cr} (required for the start of discontinuous flow) on the strain rate; (2) the "inverse" temperature dependence of ε_{cr} in the temperature range of T = 350–500°C; and (3) a positive temperature dependence of the flow stress in the temperature range of 350–550°C. These features, apparently, are caused by the formation of short-range-order structure, which leads to the development of an additional mechanism of discontinuous flow at T = 350–500°C, which consists in the interaction of the processes of the diffusional formation of regions of atomic order and their destruction as a result of passage of dislocations. It is noted that with increasing deformation rate there is observed a tendency to a shift of the temperature interval of the appearance of short-range order toward higher temperatures.

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