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МАТЕРІАЛОЗНАВСТВО

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Strain Hardening of Low-Carbon Steel in the Area of Jerky Flow

Purpose. The aim of this work is to assess the effect of ferrite grain size of low-carbon steel on the development of strain hardening processes in the area of nucleation and propagation of deformation bands. **Methodology.** Low-carbon steels with a carbon content of 0.06–0.1% C in various structural states were used as the material for study. The sample for the study was a wire with a diameter of 1mm. The structural studies of the metal were carried out using an Epiquant light microscope. Ferrite grain size was determined using quantitative metallographic techniques. Different ferrite grain size was obtained as a result of combination of thermal and termo mechanical treatment. Vary by heating temperature and the cooling rate, using cold plastic deformation and subsequent annealing, made it possible to change the ferrite grain size at the level of two orders of magnitude. Deformation curves were obtained during stretching the samples on the Instron testing machine. **Findings.** Based on the analysis of stretching curves of low-carbon steels with different ferrite grain sizes, it has been established that the initiation and propagation of plastic deformation in the jerky flow area is accompanied by the development of strain hardening processes. The study of the nature of increase at dislocation density depending on ferrite grain size of low-carbon steel, starting from the moment of initiation of plastic deformation, confirmed the existence of relationship between the development of strain hardening at the area of jerky flow and the area of parabolic hardening curve. **Originality**. One of the reasons for decrease in Luders deformation with an increase of ferrite grain size of low-carbon steel is an increase in strain hardening indicator, which accelerates decomposition of uniform dislocations distribution in the front of deformation band. The flow stress during initiation of plastic deformation is determined by the additive contribution from the frictional stress of the crystal lattices, the state of ferrite grain boundaries, and the density of mobile dislocations. It was found that the size of dislocation cell increases in proportion to the diameter of ferrite grain, which facilitates the development of dislocation annihilation during plastic deformation. **Practical value.** Explanation of qualitative dependence of the influence of ferrite grain size of a low-carbon steel on the strain hardening degree and the magnitude of Luders deformation will make it possible to determine the optimal structural state of steels subjected to cold plastic deformation.

Keywords: ferrite grain size; dislocation; Luders deformation; strain hardening index; low-carbon steel

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Introduction

In the process of plastic deformation of metal material, the interaction of moving dislocations with defects at crystalline structure is accompanied by their deceleration up to a complete stop. To resume plastic flow, it is necessary to unblock dislocations, which becomes possible only due to an increase in resulting stress. In the literature this phenomenon is called strain hardening [1, 14]. To assess the degree of development of strain hardening processes, various techniques are used based on the analysis of deformation curves [7, 16]. In appearance, the deformation curves of metallic materials are divided into two types: with and without the area of jerky flow. At the area of jerky flow, the propagation of deformation is extremely inhomogeneous and provided due to the growth of deformation bands [9, 2]. In absolute terms, deformation degree in such a band coincides with the length of jerky flow area on the deformation curve. Such deformation in the literature is called Luders deformation. Deformation on section of jerky flow ends when the entire working part of the specimen is deformed by the amount of Luders deformation. In the case of second type of flow, the deformation curve has no section of jerky flow [13]. In this case, after the appearance of the first signs of plastic deformation, the area of uniform strain hardening immediately begins [3,10]. Regardless of the type of the deformation curve, the level of mobile dislocations density must be sufficient to maintain conditions for the continuity of deformation propagation. On other hand, already at the stages of appearance of the first signs of plastic deformation as a result of interaction of moving dislocations with defects at crystalline structure, it becomes necessary to constantly compensate for the continuous decrease in the number of mobile dislocations. In this case, the growth rate of dislocation density is determined by the structural state of alloy and the size of main structural element [1]. For single phase alloys and low carbon steels, the main structural element is grain size.

Purpose

The aim of this work is to assess the effect of ferrite grain size of low carbon steel on the development of strain hardening processes in the area of nucleation and propagation of deformation bands.

Methodology

Low-carbon steels with a carbon content of 0.06–0.1% C in various structural states were used as the material for the study. Samples for the study were a wire with a diameter of 1 mm. The structural studies of metal were carried out using an Epiquant light microscope. Ferrite grain size was determined using quantitative metallographic techniques. Different ferrite grain sizes were obtained as a result of combination of thermal and thermo mechanical treatments. Varying the heating temperature, cooling rate, using cold plastic deformation and subsequent annealing, made it possible to change the ferrite grain size at the level of two orders of magnitude. The strain hardening characteristics were determined from the analysis of tension curves plotted in logarithmic coordinates. Strain curves were obtained by stretching the samples on an Instron type testing machine at a strain rate of $10^{-3}\,s^{-1}$.

Findings

The strength properties of most metallic materials near the grain boundaries of matrix are approximately of the same order of magnitude as compared to the perfect crystalline structure [3]. On other hand, dependence of the applied stress on activation of a certain dislocation slip system and the structure of grain boundary itself indicate the development of complex structural changes upon its overcoming. On this basis, for single-phase alloys, the grain size (*d*) is considered to be the main structural element that determines conditions for the initiation and propagation of deformation.

For low-carbon steels, this characteristic is the ferrite grain size (Fig. 1). It was found experimentally that with decreasing *d*, the resistance to the onset propagation of plastic deformation increases. Based on the analysis of deformation curves of low-carbon steels (Fig. 2) and generalization of experimental data on the change in resistance to small plastic deformations(σ_0) (Fig. 2) and yield stress (σ_T) on the ferrite grain size (Fig. 3) obeys to the relation [11]:

$$\sigma_T = \sigma_i + k_y \cdot d^{-\frac{1}{2}}, \quad (1)$$

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where σ_i is some initial stress, k_y is the slope coefficient. Depending on the loading conditions and investigated processes of plastic deformation, the value of σ_i is quite often identified with the friction stress of crystal lattice [7], stress increase due to the presence of impurity atoms in the alloy [9], crystal lattice defects caused by embedded atoms [3], flow stress, and the yield point of a single crystal [1]. The value k_y characterizes the stress intensity from accumulation dislocations at slip plane near the grain boundary [11]. According to the results of these works, the fulfilment of relation (1) is confirmed; however, significant difficulties arise in explaining the nature of dependence σ_i and k_y on the structural state of steel. For example, the data [7] indicate absence phenomenon of pile-up dislocations near grain boundaries in alloys with a body-centred crystal lattice. In [5], it is believed that the k_y quantity is not constant, and according to the results of [6] the exponent at *d* is not equal to (-0,5).

Based on the generalization of indicated research results, it is proposed to divide the value σ_i into components:

$$\sigma_i = \sigma_i^T + \sigma_i^E , \quad (2)$$

where σ_i^E – athermal and σ_i^T – thermal components. The value of σ_i^T is a rather complex characteristic that determines dependence of σ_i on temperature, chemical composition of alloy [6], rate and degree of plastic deformation [7] according to the ratio:

$$\sigma_i^T = \alpha_1 \mu b \rho^{0,5} + \alpha_2 \mu b \lambda^{-1} + \frac{\alpha_3 x}{b \alpha_0^3} \quad (3)$$

where μ is the shear modulus, b is the Burgers vector, ρ is the total density of dislocations, λ is the distance between the particles (precipitates) of the second phase, x, α_1, α_2, α_3, α_0 is constants.

a

b

c

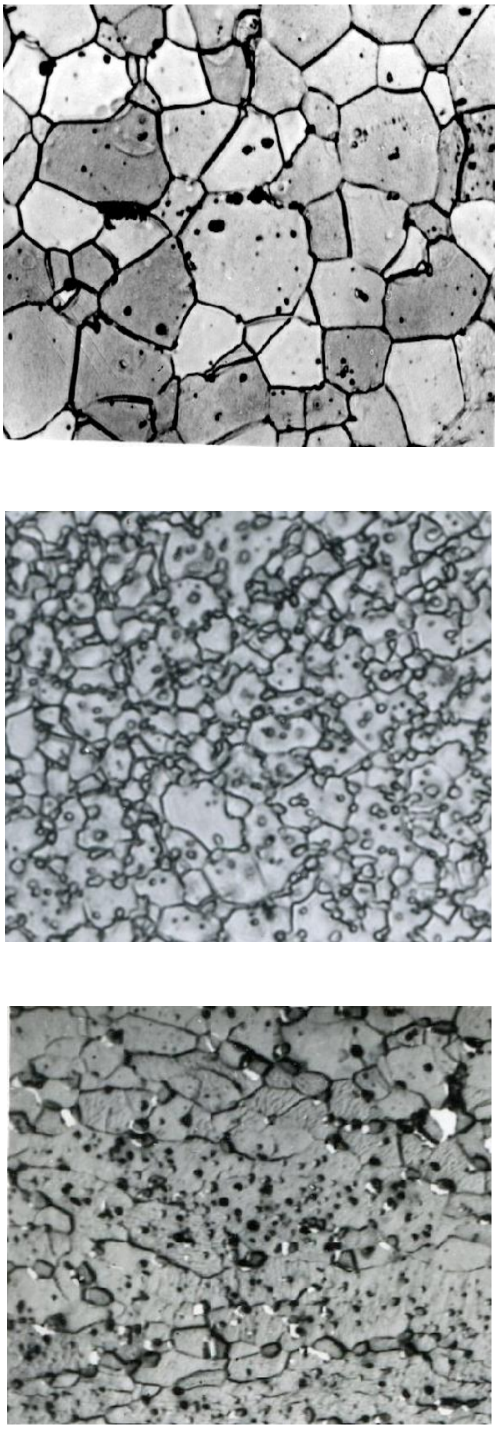

Fig. 1. Structure of low carbon steel after various heat treatments. Magnification:
a – 800, *b* – 2000, *c* – 2500

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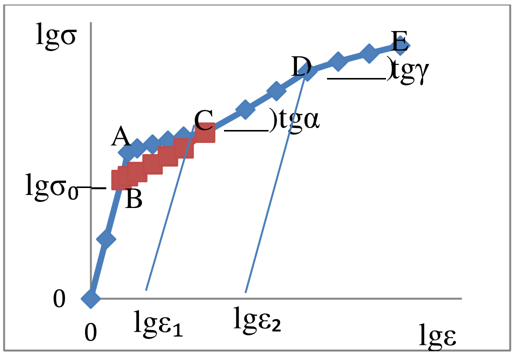

Fig. 2. Deformation curve in logarithmic coordinates

For single-phase metals and low-carbon steels, for which the influence of second phase particles is insignificant, and the role of second and third terms can be neglected, relation (3) takes form of Orowan's equation ($\sigma = \alpha \cdot \mu \cdot b \cdot \sqrt{\rho}$) [1].

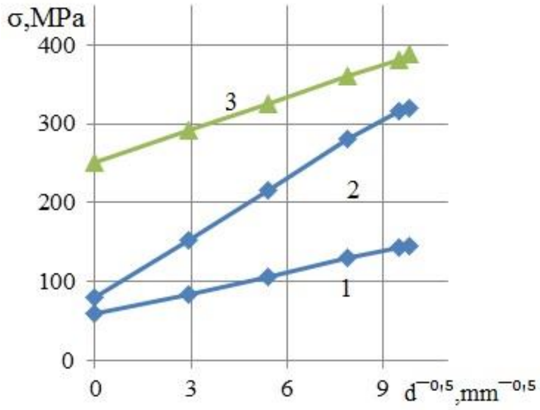

Fig. 3. Influence of ferrite grain size by ratios (1), for σ_0 (*1*), yield stress (*2*) and ultimate strength (*3*) of low-carbon steel with 0.06% carbon

Thus, the contribution of thermal component to the value σ_i for low-carbon steels can be significantly simplified and reduced to dependence on the dislocation density present in metal. A significant variety of the structural state of metallic materials is reflected in the conditions of the plastic flow initiation. Depending on the type of crystal lattice, under loading, metallic materials can have a jerky flow area, or the development of the processes of uniform strain hardening is immediately observed.

As follows from the analysis of numerous experimental data, relations at type (1) are equally successful both for steels with and without a yield area, for describing the yield stress and flow stress up to the ultimate strength (Fig. 3, curve 3). At the same time, a contradiction arises when using the same equation to describe the strength characteristics of metal with a different structural state. Indeed, a single mechanism is used to describe the behaviour of steels with different degrees of plastic deformation [4]. However, depending on the degree of plastic deformation, the mechanism has certain differences. For example, for steel with a yield point, the onset of flow is due to the formation of slip bands from generation of dislocations by the boundaries of ferrite grains [1]. While at absence of a physical yield point, development of slip is associated with the work of other sources of dislocations [7]. As for the absolute values σ_i and k_y, the differences found may be associated not only with the test conditions and chemical composition of the steel, but with differences in the initial structural state. For technical iron and low-carbon steel, in the range of ferrite grain sizes from 40 to 3,6 μ, a change in k_y from 18.2 to 22.2N/mm$^{3/2}$ was found [8]. Moreover, at the same time, with a sufficiently accurate coincidence between the calculated relation (1) and the experimental values of yield stress on ferrite grain size, the fulfillment of this dependence may be violated when *d* is replaced by the substructure parameter [1, 7]. Thus, for technical iron and low-carbon steel with subgrains sizes of 0.5–7.0 μ, to explain the nature of dependence of yield stress on *d* relation of the type (1) [8] was used, in which k_y is presented as a quantity dependent on dislocations density in subgrains. The exponent at *d* can also differ from the traditional value (–1/2), reaching the value (–3/4) [1].

Thus, characteristics of resistance to small plastic deformations for low-carbon steel are fairly well described by relation (1), when the role of the main structural element is assigned to the ferrite grain size, and other influences are minimal. With an increase in the dislocation density in the area of formation of the first signs of plastic deformation, the inevitable processes of their redistribution caused the development of an alternative method for determining the quantities σ_i and k_y. The proposed technique is based on the analysis of deformation curve. As a result of extrapolation of the area of uniform strain hardening of the CD section (Fig. 1) to zero plastic deformation, according to

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[15], the obtained value σ_0 was equated to σ_i from relation (1).

This identification is based on the concept of practically identical conditions for the appearance of the first signs of plastic deformation and rather close absolute values. According to the proposed method, after replacing σ_i by σ_0, and k_y by k_Y^1, relation (1) takes the form:

$$\sigma_T = \sigma_0 + k_Y^1 \cdot d^{-\frac{1}{2}}. \quad (4)$$

Considering that:

$$\sigma_T - \sigma_0 = k_Y^1 \cdot d^{-\frac{1}{2}}, \; k_Y^1 = \frac{\sigma_T - \sigma_0}{d^{-\frac{1}{2}}}. \quad (4a)$$

In contrast to σ_i and k_y from relation (1), quantities σ_0 (Fig. 2) and k_Y^1 (Fig. 4) are dependent on the ferrite grain size. Comparative analysis of the two methods for determining σ_i and k_y showed both coincidence in absolute values and their identical essence [15], and fundamental differences [1], up to inconsistency [8]. In fact, these techniques are based on several different mechanisms of plastic flow development at the initial stages of metal loading. In one case, it is the motion of free dislocations, in the other – it is unblocking or nucleation of additional dislocations upon deformation. Thus, the indicated methods for estimating parameters of deformation initiation and the initial stages of its propagation should be considered as complementary to each other [8].

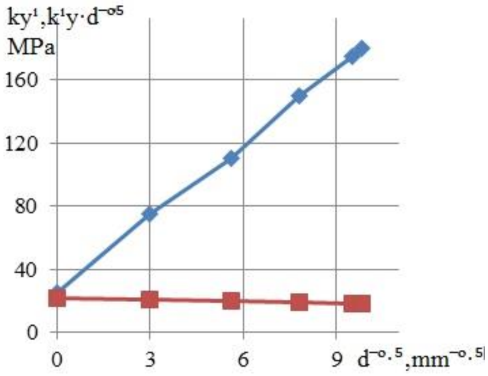

Fig. 4. The dependences $k_Y^1 \cdot d^{-0,5}$ – (♦) and $k_Y^1 \cdot d^{-0,5}$ – (■) on the ferrite grain size of low-carbon steel

It follows from above analysis that when the first signs of plastic deformation appear, resulting changes at internal structure in the metal are by their nature quite close to changes as a result of the development of strain hardening processes. Indeed, the evolution of dislocation structure during plastic deformation gives reason to believe that the development of strain hardening processes is observed already from the moment of appearance of the first signs of irreversible mobile dislocations. Detailed studies of changes in the dislocation structure on the ferrite grain size of low-carbon steel made it possible to determine two fundamental positions. First, these are rather close values of the subgrains size in comparison with the size of the dislocation cell of the deformed metal [1]. Second, the dislocation density at the beginning of the plastic flow is inversely proportional to the size of the ferrite grain. Taking into account the Orowan relationship of $\sigma_T \approx f(\sqrt{\rho})$ and $\sigma_T \approx f(1/\sqrt{d})$ according to (1), for the initial stages of plastic flow, we will write:

$$\rho \approx \frac{1}{d}. \quad (5)$$

The use of relation (5) makes it possible to assess the role of strain hardening processes at the stages of nucleation and propagation of deformation bands. Indeed, as shown in [9], the deformation into the front of Luders band with a width of λ varies from $\varepsilon = 0$ to ε_L (Fig. 5).

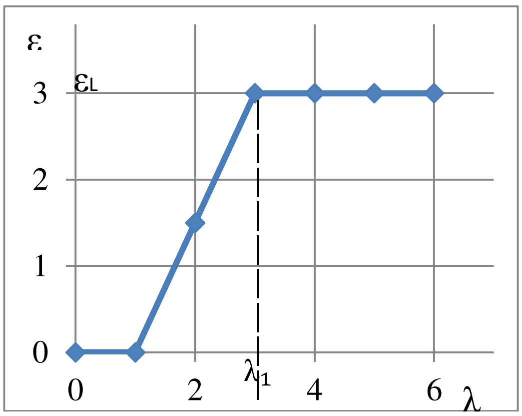

Fig. 5. Distribution diagram of the magnitude of plastic deformation along the Luders band front (ε_L is the Luders deformation, λ_1 is the width of band front in units of grain size)

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Considering that a certain amount (a) of grains is laid along the length, the gradient of deformation by the front of strain band will be:

$$\frac{d\varepsilon}{d\lambda}=\frac{\varepsilon_L}{a\cdot d} \tag{6}$$

where ε_L is the Luders deformation.

In essence, the value of $\frac{d\varepsilon}{d\lambda}$ should depend not only on the initial dislocation density, but also on the magnitude of its increment during the formation of the deformation band. According to the Orowan relation [9], the local strain rate ($\dot{\varepsilon}$) is generally estimated as follows:

$$\dot{\varepsilon}=\rho_m\cdot b\cdot k\cdot v, \tag{7}$$

where ρ_m is the density of mobile dislocations, b is the Burgers vector, k is the geometric factor, which is equal up to 1, and v is the velocity of dislocation movement. Compared with (7), $\dot{\varepsilon}$ can be represented in terms of $\frac{d\varepsilon}{d\lambda}$ for the front of deformation band:

$$\dot{\varepsilon}=v_L\cdot\frac{d\varepsilon}{d\lambda}, \tag{8}$$

where v_L is the speed of movement in the front of the deformation band. Assuming that v_L cannot be greater than v, for the initial conditions of propagation of plastic deformation, the following relation should be satisfied:

$$\nu_L=\nu \tag{9}$$

Equating (7) and (8) with each other, taking into account (6), finally obtain that for the formation of a deformation band of ε_L in steel with a ferrite grain size (d), the required density of mobile dislocations:

$$\rho_m=\frac{\varepsilon_L}{a\cdot b\cdot d}, \tag{10}$$

where a is the number of ferrite grains that fit on the width in the front of the deformation band. It was found experimentally [9] that the quantity a isactually a variable characteristic. It depends on the size of ferrite grain and can range from one to several d. The maximum value of a does not exceed 3. On this basis, one should take into account the existence of a certain gradient density of dislocations at transition from the peripheral sections in the front of the deformation band to the volumes of metal that have already undergone Luders deformation. The theoretical concepts of a continuous change at density of dislocations inside individual grains during the formation in the front of deformation band (Fig. 6, *a*) have not received practical confirmation [12]. In fact, from the distribution analysis of etching pits in individual ferrite grains, it was found that dislocations within one grain are distributed almost evenly. Thus, the density of dislocations inside an individual grain, during the formation in the front of the band, should be considered constant without gradients at distribution. As a result, structural changes at metal, within the front of deformation band, will be determined by a discrete change in density of dislocations at transition from one grain to another (Fig. 6, *b*).

a *b*

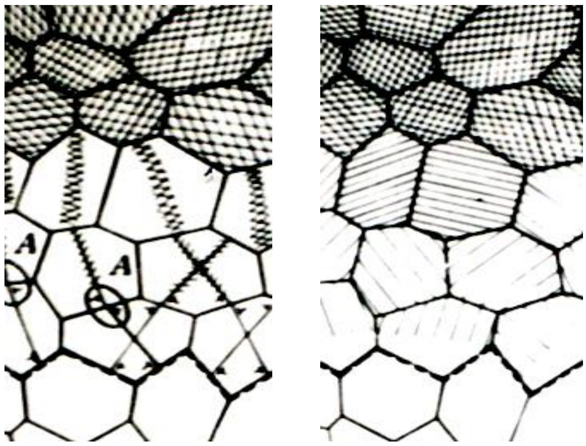

Fig. 6. The scheme of structure in the front of deformation band with a continuous distribution of dislocations within individual grains (*a*) and at a constant dislocation density (*b*) according to [10]

On this basis, development processes of strain hardening of metal at the propagation region of deformation bands have a discrete character. The resulting conclusion has a certain applied value. An example is the results of the development technology for obtaining alloys with ultra-fine grains and the search for areas of their further use [1]. In

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general, development of strain hardening processes during propagation of deformation bands presupposes an inevitable increase in deforming stress. Based on this, the increase in deforming stress after completion of jerky flow should depend on the value of ε_L.

To estimate the indicated increase of stress the equation [9] was used:

$$\sigma = \alpha \cdot \mu \cdot b \cdot \sqrt{\rho}\,. \qquad (11)$$

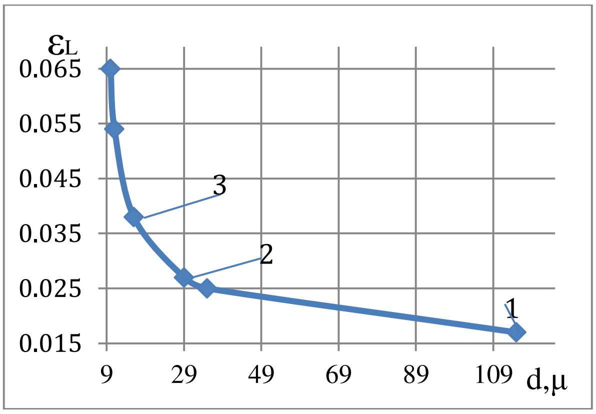
After substituting (10) into (11), we obtained a relation for estimating the stress required for the formation of a deformation band nucleus in a single-phase alloy or low-carbon steel:

$$\sigma_1 = \alpha \mu b \sqrt{\frac{\varepsilon_L}{abd}} \qquad (12)$$

The fulfillment of dependence (12) and the possibility of using it to describe the stress increase at the yield point is confirmed by experimental data for most low-carbon steels [9, 12]. Thus, at the moment of completion in the section of the jerky flow, a certain number of dislocations is already present in metal, which, depending on *d*, can be estimated from (12). At same time, the data indicating the lack of unambiguity at nature of influence of the ferrite grain size on characteristics of the work hardening of low-carbon steel in the indicated section of the deformation curve deserve some attention. In addition to a significant number of works showing an increase at coefficient and rate of strain hardening with an increase in *d* [1], there are data indicating a deviation from this ratio [8].

One of the explanations for these deviations at fulfillment of (1) is the absence of influence from substructure parameters in the initial state of low-carbon steel. As a confirmation of the above provisions, let us evaluate the hardening effect at region of deformation band formation depending on the ferrite grain size of low-carbon steel using relation (12). After substitution of experimental values ε_L in (12), for specific grain sizes of low-carbon steel ferrite in different structural states (Fig. 7, *a*), taking μ = 82260 MPa, $b = 2.48 \cdot 10^{-7}$ mm, $a = 2$, $\alpha = 0.3$, the calculated results (σ_1) are shown in Fig. 7, *b*.

a



b

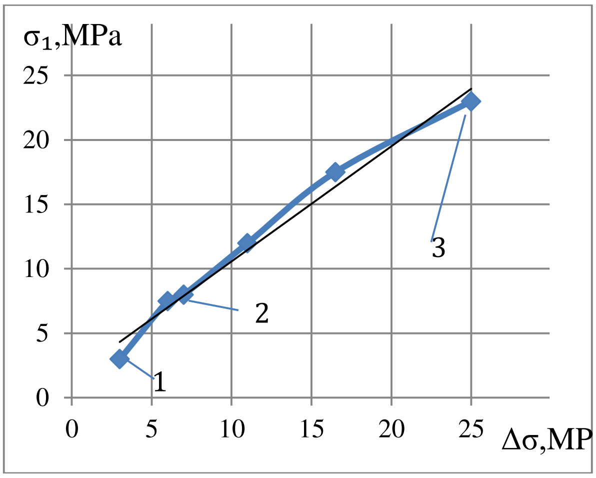


Fig. 7. Influence of ferrite grain size on steel with 0.06% C on ε_L – (*a*). The relationship between the calculated stress at the end of yield area (σ_1) according to (12) and obtained experimentally ($\Delta\sigma$), where $\Delta\sigma = \sigma_2 - \sigma_3$, σ_2, and σ_3, respectively, are the flow stress at the end and the beginning of the yield area (*b*). For structural states of steel:
1 – isothermal transformation at a temperature of 550°C, deformation by drawing by 25–90% and annealing at 680°C, 1 h; *2* – martensite quenching and tempering at 680°C, 1 h; *3* – normalization

Comparative analysis between the calculated by (12) values of the stress increase at the section of jerky flow (σ_1) and determined from the tension curve ($\Delta\sigma$), indicates a fairly good agreement (Fig. 7, *b*), with a maximum difference between them no more than 5%. Thus, the observed increase at dislocation density during propagation of the deformation band should be considered as confirmation of the fact of development of strain hard-

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ening processes. Indeed, if return to the analysis of deformation curve (Fig. 2), then extrapolated part of the curve (*BC*) can be considered as a kind of continuation of the *CD* section region of homogeneous strain hardening.

In this case, as a result of formal substitutionof the current values of deforming stress and values of plastic deformation for the section of *BC* curve (Fig. 2) into the ratio:

$$\frac{d\sigma}{d\varepsilon}=\frac{m\cdot(\sigma-\sigma_0)}{\varepsilon}, \qquad (13)$$

it is possible to evaluate the development of strain hardening processes until reaching region with parabolic hardening (*CD*, Fig. 2). For the region beginning formation of the first deformation band (point *B* in Fig. 2), the current value of deforming stress should correspond to the value σ_0, at absence of macroscopic plastic deformation (2). After substituting indicated values into relation (13) for point *B* at deformation curve, to obtain:

$$\frac{d\sigma}{d\varepsilon}=\frac{m\cdot(\sigma-\sigma_0)}{\varepsilon}=\frac{m\cdot 0}{0} \quad (14)$$

Uncertainty in definition $\frac{d\sigma}{d\varepsilon}$ according to (14), after transformation takes the form:

$$\lim_{\substack{\sigma\to\sigma_0\\ \varepsilon\to 0}}\frac{d\sigma}{d\varepsilon}=\lim\frac{m\cdot(\sigma-\sigma_0)'}{\varepsilon'}=m \qquad (14a)$$

The fulfillment of relation (14a) confirms effect from change of structure during the formation of the embryo of band of the deformation on subsequent development of strain hardening processes in the region with a developed substructure.

Originality and practical value

One of the reasons for decrease at Luders deformation with an increase in the ferrite grain size of a low-carbon steel is an increase in strain hardening coefficient, which accelerate decomposition of uniform distribution of dislocations in the front of the deformation band. The flow stress during the initiating of plastic deformation is determined by the additive contribution of frictional stress of the crystal lattice, by state of the grain boundaries of ferrite, and density of mobile dislocations. It was found that the size of dislocation cell increases in proportion to the diameter of ferrite grain, which facilitates annihilation of dislocations during plastic deformation. Explanation of influence of the ferrite grain size of a low-carbon steel on the coefficient of strain hardening and magnitude of Luders deformation will make it possible to determine the optimal structural state of steels intended for cold plastic deformation.

Conclusions

1. With an increase in the ferrite grain size in a low-carbon steel, the decomposition likelihood of uniform dislocations distribution into a structure with a certain periodicity increases. This leads to a decrease in Luders deformation.
2. Based on the analysis of structural changes in low-carbon steel, the nucleation and propagation of the deformation band is accompanied by the development of strain hardening processes.

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МАТЕРІАЛОЗНАВСТВО

Деформаційне зміцнення низьковуглецевої сталі в області переривчастої течії

Мета. Основною метою роботи є оцінка впливу розміру зерна фериту низьковуглецевої сталі на розвиток процесів деформаційного зміцнення в області зародження й поширення смуг деформації. **Методика**. Як матеріал для дослідження були низьковуглецеві сталі з вмістом вуглецю 0,06–0,1 % С в різному структурному стані. Зразками для дослідження взято дріт діаметром 1 мм. Структурні дослідження металу здійснено з використанням світлового мікроскопа «Епіквант». Розмір зерна фериту визначено за методикою кількісної металографії. Різний розмір зерна фериту отримано в результаті поєднання термічних і термомеханічних обробок. Варіювання температур нагрівання, схем і швидкостей охолодження з використанням холодної пластичної деформації й подальшого відпалу дозволило змінювати розмір зерна фериту на рівні двох порядків значень. Криві деформації отримано в результаті розтягування зразків на випробувальній машині типу «Інстрон». **Результати**. На основі аналізу кривих розтягування низьковуглецевих сталей із різним розміром зерна фериту встановлено, що зародження й поширення пластичної деформації в області переривчастої течії супроводжується розвитком процесів деформаційного зміцнення. Дослідження процесу приросту густини дислокацій залежно від розміру зерна фериту низьковуглецевої сталі від моменту зародження пластичної деформації підтвердило існування зв’язку між розвитком деформаційного зміцнення в області переривчастої течії й області кривої з параболічним зміцненням. **Наукова новизна**. Однією з причин зменшення деформації Людерса зі збільшенням розміру зерна фериту низьковуглецевої сталі є зростання показника деформаційного зміцнення, що пришвидшує розпад рівномірного розподілу дислокацій у фронті смуги деформації. Напруження течії під час зародження пластичної деформації визначається адитивним внеском від напруження тертя кристалічних грат, стану меж зерен фериту й густини рухомих дислокацій. Установлено, що пропорційно діаметру зерна фериту відбувається збільшення розміру дислокаційної комірки, що полегшує анігіляцію дислокацій під час пластичної деформації. **Практична значимість**. Пояснення якісної залежності впливу розміру зерна фериту низьковуглецевої сталі на показник ступеня деформаційного зміцнення й величину деформації Людерса дозволить визначити оптимальний структурний стан сталей, які піддають холодній пластичній деформації.

Ключові слова: розмір зерна фериту; дислокація; деформація Людерса; показник деформаційного зміцнення; низьковуглецева сталь

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