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On the applicability of hardening mechanisms to low-carbon and low-alloy steels

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ABSTRACT

On the basis of experimental studies, the approximate contribution of various hardening mechanisms to the yield point of low-carbon and low-alloy steels is estimated. It has been established that for hot-rolled steels (St.3sp and St5ps), solid-solution and grain-boundary hardening (54.0% and 29.0, %) make the greatest contribution to the yield point. The predominant strengthening mechanism of low-alloy steel 10HNDP is solid solution, a high proportion of which in this steel is explained by the resistance to moving dislocations from the side of dissolved atoms of Ni, Cu, P, and Cr in α -Fe. In low-alloy steel 16G2AF, along with these hardening components, the role of precipitation hardening is noticeable (20.0%). It is shown that thermomechanical treatment of steel grade St.5ps leads to an increase in the value of dislocation hardening up to 27.0% due to an increase in the density of dislocations and the retention of most of the dislocations in the rolled stock during accelerated cooling of hot-deformed austenite. It is noted that solid solution hardening with alloying with cheap alloying elements (Mn, Si), as well as dislocation and dispersion hardening through the use of thermomechanical treatment in combination with the addition of carbide and nitride-forming elements V (C, N).

Keywords: hardening mechanisms, yield stress, thermomechanical treatment, accelerated cooling, dislocation density, phase components.

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Introduction

The structural strength of low-carbon and low-alloy steels with a ferrite-perlite structure can be characterised by the yield strength of steel and the temperature of transition from viscous to fragile. Therefore, knowledge of individual hardening mechanisms allows for an indicative quantitative assessment of the steel yield strength and compare the calculated values of the yield strength with experimental ones. The purpose of this article is to

quantify the yield strength of low-carbon and low-alloy steels widely used in construction and engineering in terms of structure parameters after various technological treatments.

Experimental part

Initial data for quantifying steel strength are data on its chemical composition, distribution of elements between phases and quantitative parameters of the structure (grain size, ratio of

phase and structural components, their size and distribution, nature of the dislocation structure, volume fraction and size of dispersed particles). Such a calculation is based on the quantitative ratios established for each strengthening mechanism. Note that the above calculations seem to be indicative semi-quantitative, since there are a number of assumptions and simplifications in the theory of the strengthening mechanisms themselves. In addition, the complex distribution of dislocations in real steels and alloys is difficult to be strictly quantified. However, such calculations are necessary to identify the role of individual strengthening mechanisms in the formation of individual steel properties. The solution of such problems allows you to approach the solution of the main problem of physical and applied materials science - the quantitative relationship between the structure and properties of steels and alloys.

The main characteristics of steels to determine their structural strength are strength and propensity for brittle destruction [1]. Steel strength is estimated by the lower yield strength by the known Hall-Petch ratio, which for stretching conditions is as follows:

$$\sigma_y = \sigma_i + k_h \cdot d^{-1/2} \quad (1)$$

Where σ_i is the friction voltage of the lattice when dislocations move inside the grains;

k_h - coefficient characterising the contribution of grains to hardening;

d – is the diameter of the grain.

With sufficient accuracy, this ratio is applicable to ferrite steels with grain size from 0.3 to 400 μm ; it follows that the lower yield strength of the material increases with the decrease in grain size [2]. The tendency of steel to fragile destruction is assessed by the temperature of transition from viscous to fragile, which is defined as the ratio of viscous destruction area to the initial design section. The lower the transition temperature from

viscous to fragile, the more reliable the material is, so they tend to use a material whose transition temperature is lower than the operating temperature [3].

It is believed that the contribution of certain hardening factors to the total strengthening additive σ_i in the Hall-Petch strengthening can be represented as a sum:

$$\sigma_i = \sigma_o + \sigma_{s.s} + \sigma_p + \sigma_d + \sigma_{p.h} \quad (2)$$

The lower limit of steel yield followed from equations (1) and (2) is characterized by the lattice friction stress- α -Fe σ_o , solid solution strengthening $\Delta\sigma_{s.s}$, hardening due to pearlite formation σ_p , strain hardening σ_d , dispersion hardening $\sigma_{p.h}$ and strengthening due to grain boundaries $\sigma_{gb} \cdot d^{-1/2}$. The share of the contribution of individual hardening factors to the total lower yield stress of steel is not the same and depends on the type of alloying elements and the degree of alloying, its presence and dispersion of the hardening phases, the use of heat treatment, and other reasons.

This paper proposes an analysis of the effectiveness of various mechanisms of hardening of low-carbon and low-alloy steels of grades St3sp, St5ps, 10KhNDP, 16G2AF. They are used in construction and differ not only in chemical composition, but also in the applied heat treatment. The magnitude of the individual hardening factors, as well as their contribution to the total lower yield stress of these steels, are determined by the well-known empirical formulas of Table 1 [[3], [4]]. The coefficients required for the calculation are taken from these literature sources. The values of the lower yield stress of the studied steels calculated in this way were compared with the data according to GOST-380, GOST 19282, GOST 5781, GOST 10884.

Table 1 - Hardening coefficients and calculation formulas for quantitative evaluation

| No | Hardening factors | Calculation formula of hardening |
|----|--|---|
| 1 | Lattice friction stress α -Fe | $\sigma_o = 2 \cdot G \cdot 10^{-4}$ |
| 2 | Solid solution alloying | $\Delta\sigma_{ss} = \sum_{i=1}^n k_i \cdot c_i$ |
| 3 | Hardening due to the formation of pearlite | $\Delta\sigma_p = 2,4\% \cdot \Pi$ |
| 4 | Deformation hardening | $\Delta\sigma_d = 0,5 \cdot G \cdot b \cdot \rho^{1/2}$ |
| 5 | Dispersion hardening | $\Delta\sigma_{p.h} = (9,8 \cdot 10) / \lambda \cdot \ln(2\lambda)$ |
| 6 | Intergranular (substructural) hardening | $\Delta\sigma_{gb} = k_h \cdot d^{-1/2}$ $\Delta\sigma = k_s \cdot I^{-m}$ |

Determination of structural parameters (perlite content in steel, diameter of ferrite grains, size and volume fraction of carbonitride phase, etc.) for quantitative assessment of the lower yield limit was carried out by quantitative metallography methods on the NeoPhot 21 research microscope and the UEMV-100V electron microscope. The average length of a straight section crossing the grain in the microgrind plane was used as the diameter of ferrite (d) grain. The volume fraction of dispersed particles (1) and their diameter (D) in low-alloy steel 16G2AF were determined by electron microscopy, and the interparticle distance (λ) was determined by a known ratio:

$$\lambda = D \cdot (P/6f)^{1/2}$$

The proportion of the perlite component is determined by the Rosiwal method. According to it, the area of the structural component is calculated by the lengths of the straight section falling on each of the structural components in accordance with the evaluation criterion. Strengthening of thermomechanically hardened steels St.5ps is determined by X-ray structural analysis by the shape of diffraction lines. The density of dislocations is quantified by translucent electron microscopy of thin foils.

Discussion of the results

The lattice friction stress - α -Fe (Peierls-Nabaro stress) is estimated by the formula:

$$\sigma_o = 2 \cdot 10^{-4} \cdot G$$

где G – iron shear modulus (G = 84000 МПа).

However, this estimate depends largely on the content of impurities in the metal [3]. For high purity iron [$<10^{-7}$ (C+N)], the experimental values obtained ($\sigma_o=18-21$ МПа) almost coincide with the theoretically calculated ($\sigma_o = 17$ МПа). α -Fe-based steels take the value of $\sigma_o = 30$ МПа in calculations.

Other additions of hardenings ($\Delta\sigma_s, \sigma_p, \sigma_d, \sigma_{p,h}, \sigma_{gb}$) taking into account known assumptions, quantified for the steels under study, are given in Table 1 [[3], [4]]. For the convenience of comparing and analysing the effectiveness of various hardening mechanisms, the results of calculations are presented in the form of pie and column charts, histograms (Fig. 1-5). In carbon steels St3sp, St5ps (hot-rolled state), the main components of quenching are solid-solid and grain-boundary hardening, which account for St3sp steel 54% and 29%, respectively (Fig. 1). They are equal in absolute value: 140.5 МПа and 89.9 МПа.

Table 2 - Initial data for the quantitative assessment of the lower yield strength of the studied steels

| № | Characteristics of steel | Steel grade | | | | |
|---|---|-------------|--------------------|----------------------|--------|---------|
| | | St3sp | St5sp (hot rolled) | St5sp (heat-treated) | 16G2AF | 10KHNDP |
| 1 | The content of alloying elements in α -Fe, % | | | | | |
| | Mn | 0.52 | 0.65 | 0.65 | 1.5 | 0.45 |
| | Si | 0.21 | 0.11 | 0.11 | 0.45 | 0.27 |
| | P | 0.04 | 0.04 | 0.04 | 0.035 | 0.095 |
| | V | - | - | - | 0.11 | - |
| | Ni | - | - | - | - | 0.45 |
| | Cr | - | - | - | - | 0.65 |
| | Cu | - | - | - | - | 0.40 |
| | (C+N) | 0.015 | 0.015 | 0.015 | 0.015 | 0.015 |
| 2 | Hardening phase | - | - | - | V(CN) | - |
| 3 | Perlite fraction, % | 22 | 35 | 26 | 17 | 14.3 |
| 4 | Grain size, d мм | 0.056 | 0.051 | 0.033 | 0.014 | 0.028 |
| 5 | Volume fraction of dispersed particles (f), % | - | - | - | 0.096 | - |
| 6 | Size of dispersed particles (D), нм | - | - | - | 30 | - |
| 7 | Interparticle distance (X), нм | - | - | - | 765 | - |
| 8 | Dislocation density (ρ), cm^{-2} | 10^3 | 10^8 | 10^{10} | 10^8 | 10^8 |

Note: Experimental determination of dislocation density and volume fraction of phase components is a difficult task, so the data are taken from reliable literary sources.

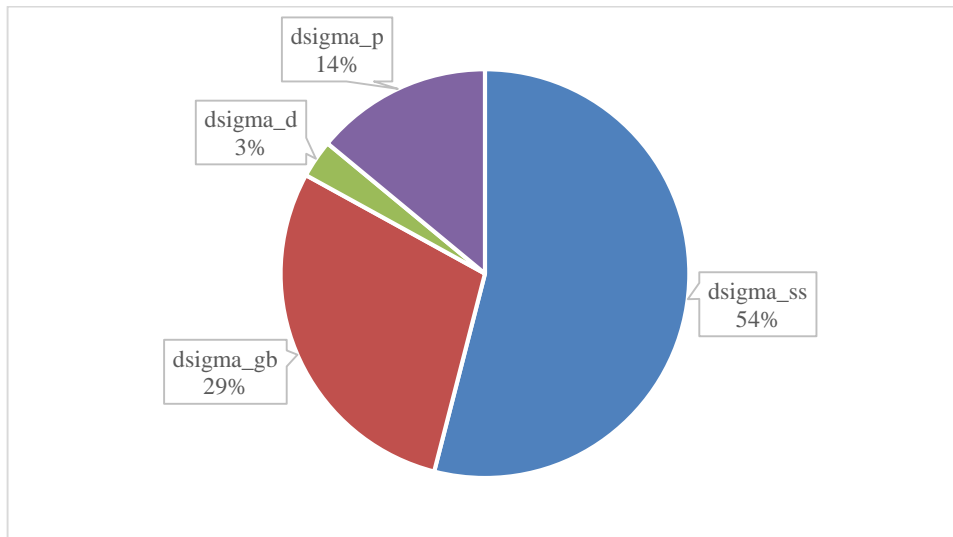


Figure 1 - Pie chart of hardening components for St3sp

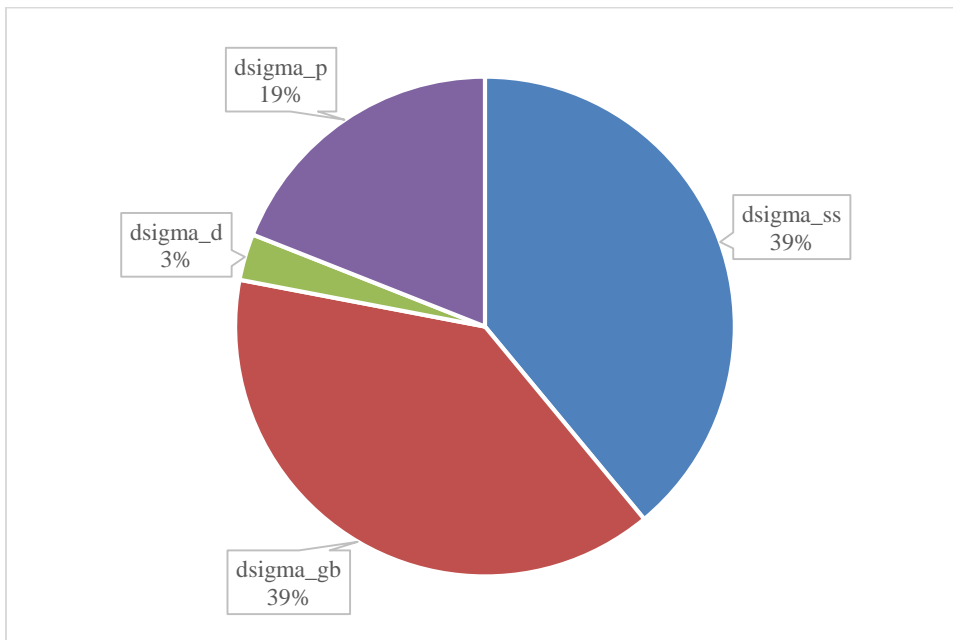


Figure 2 - Pie chart of St5ps hardening components (hot-rolled)

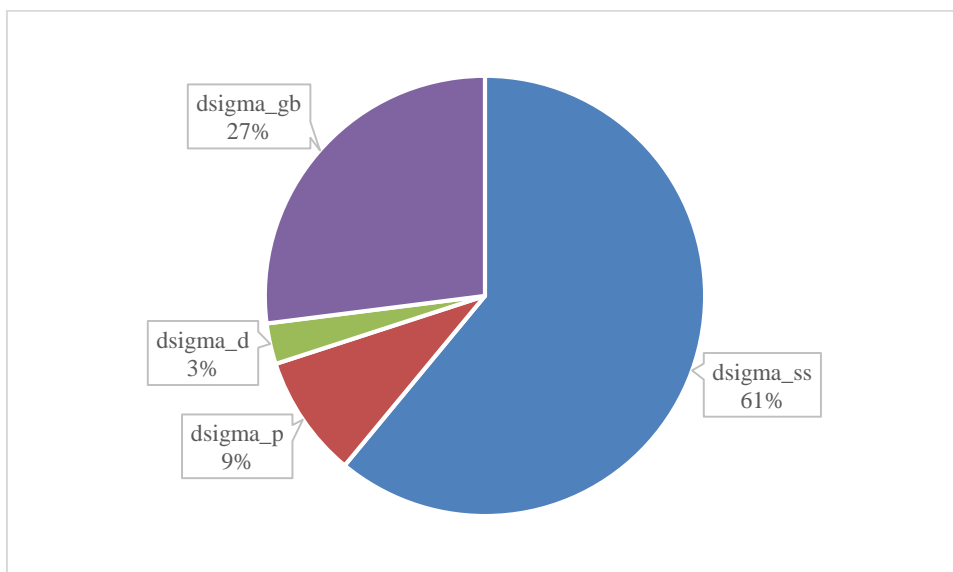


Figure 3 - Pie chart of the distribution of hardening components for 10KHNDP

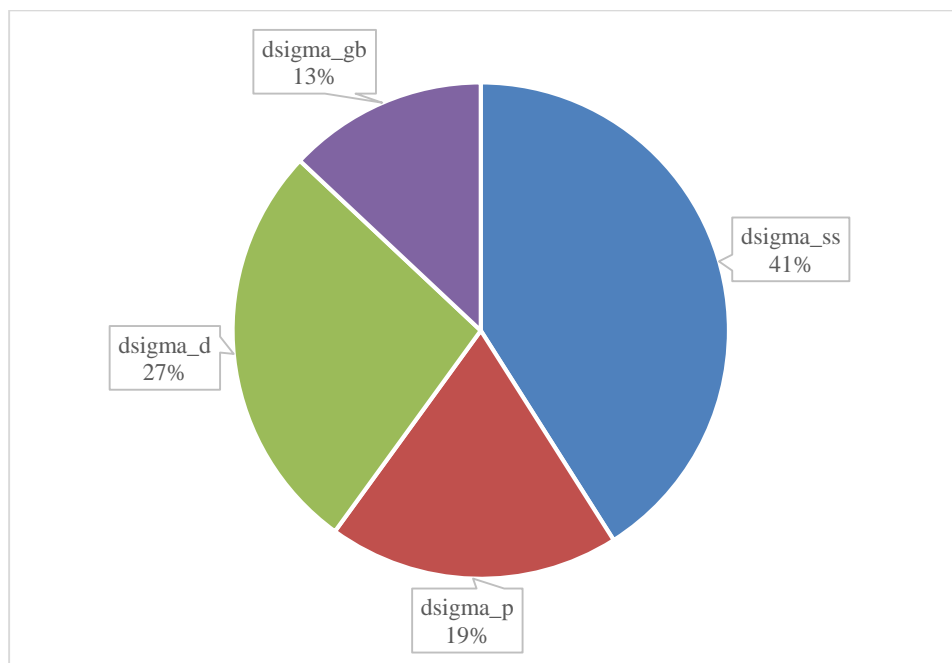


Figure 4 - Pie chart of St5ps hardening components (heat-strengthened)

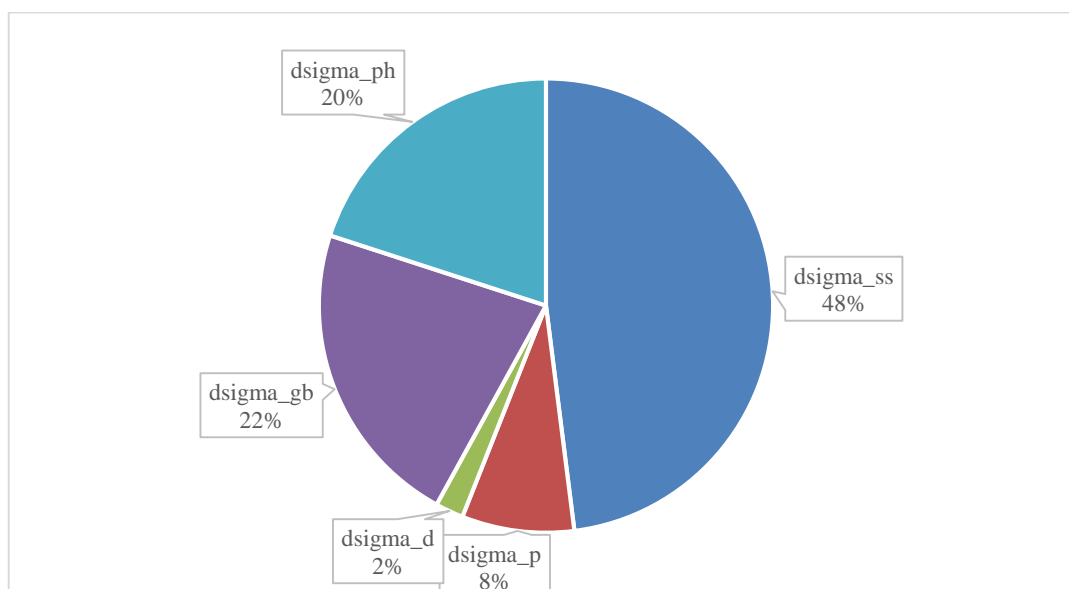


Figure 5 - Pie chart of hardening components for 16G2AF steel

When St5ps steel is subjected to thermomechanical treatment, deformation (location) hardening makes a significant contribution to the overall hardening. If the share of deformation hardening of St5ps steel cooled in calm air from the end temperature of rolling 1050 °C (hot-rolled state) is 3% in this steel, then in thermomechanically treated steel according to the intermittent hardening scheme followed by high-temperature mode (in a thermally hardened state), the proportion of deformation hardening increases to 27%, $\Delta\sigma_d = 104$ MPa (absolute value). This is due to the fact that during thermomechanical processing, recrystallisation processes are

suppressed by sharp cooling and therefore a significant part of the dislocations arising from hot rolling of austenite are recorded.

Thus, the dislocation structure of hot-deformed austenite is inherited by the formed martensite in the process of phase austenitic-martensitic transformation [5].

The formation of martensite crystals is achieved simultaneously with the grinding of austenite grain during thermomechanical processing.

The predominant mechanism of hardening low-alloy steel 10CNP is solid solution (Fig. 3). The high proportion of solid-solid hardening in 10CNP steel is due to the resistance of moving dislocations on the

part of dissolved atoms Ni, Cu, P and Cr in α -Fe, provided that the hardening of the solid solution is caused by a difference in atomic diameters in the lattice, alloying element and their elastic modules. Ferrite hardening coefficients of these elements:

$$K_{Ni}^F = 30; K_{Cu}^F = 40; K_P^F = 690; K_{Cr}^F = 30.$$

Indicating the effectiveness of this hardening mechanism and its applicability, it should be emphasised that there is probably some optimal degree of alloying of α -Fe, since saturation of α -Fe with substitution atoms causes only dangerous elastic deformation of the lattice and reduction of the impact toughness of the alloy [6].

In low-alloy steel 16G2AF, the role of dispersion hardening is noticeable, equal to 20% (Figure 5), $\Delta \sigma_{p.h} = 94.0$ MPa. From Table 2, this steel forms a carbonitride phase V(C, N), which strengthens ferrite by the Orowan mechanism, will form dispersed carbonitride. It is assumed that the carbonitride phase V(C, N) is incoherent with the lattice (α -Fe) and, as a result, the dislocations envelope non-coherent discharges V(C, N).

However, there are statements [[7], [8]] believe that in low-alloy steels small carbonitride particles released directly from the lattice may be coherently linked to it [9].

The influence of disperse phases on grain size is reflected in the efficiency and prospects of dispersion hardening [10]. Table 2 shows that a smaller grain $d=0.014$ mm is formed in 16G2AF steel, in the structure of which there is a carbonitride phase V(C, N), which has an embryo effect in the formation of new austenite grains during the transition through critical points Ac1 and Ac3 [11]. In addition, the carbonitride phase inhibits the growth of austenite grain when further heated to the dissolution temperature of these phases in austenite. Undissolved carbides and nitrides, as well as those released from austenite before the beginning of $\gamma \rightarrow \alpha$ transformation, serve as embryo centres for the formation of new ferrite grains [12]. All this leads to a noticeable grinding of ferrite grain in low-alloy steels with dispersed strengthening phases. Thus, dispersed particles of the carbonitride phase V (C, N) in steel cause additional grain boundary strengthening [13]. For the first time, this feature of hardening of carbonitride phases by dispersed particles is specified in article [14].

Ferrite is the main phase and structural component in low-carbon and low-alloy steels. Its share in these steels reaches 90÷95% [15]. When applying the load, deformation begins to manifest

itself in ferrite, this is due to the fact that perlite is a "barrier" for such deformation. A certain contribution to the overall hardening (in the yield strength) is made by hardening from pearlite components [16]. From the above figures 1-4 it can be seen that the share of hardening from perlite formation is about 10-20%, according to the absolute estimate of $\Delta \sigma_p = 75$ MPa for hot-rolled steels St3sp and St5ps [17].

Comparison of the calculated values of the steel yield strength with its value in the relevant GOSTs shows a satisfactory difference: for 16G2AF steel after normalisation, this difference is 17.8% (GOST 19282) and for St.5ps. (hot-rolled state) the difference is 12.0% (GOST 5781). After VTMO, the difference between the calculated value of the yield strength and the value according to GOST 10884 is 13.4%. These data indicate the applicability of quantification of the steel yield strength by the parameters of the formed structure after certain treatments and provide reliable information on existing hardening mechanisms [18].

It should be noted that non-metallic inclusions can affect the mechanical properties of steels [19]. However, their volume fraction in the studied steels does not exceed 0.1%, they do not have a strengthening effect and, therefore, in this work the behaviour of non-metallic inclusions was not taken into account [20].

Conclusion

The contribution of different hardening mechanisms to reducing the yield strength of low-carbon and low-alloy steels varies. For hot-rolled steels, the greatest contribution is made by solid-soluble and intergrain hardening (54% and 29% St3sp, 61% and 27% 10CNP), and 16G2AF steel, along with these hardening components, has a noticeable role of dispersion hardening (22%).

Thermomechanical treatment of St5ps leads to an increase in the value of dislocation hardening to 27% due to an increase in the density of dislocations and the preservation of most of the dislocation with accelerated cooling of hot-deformed austenite.

As effective and promising ways to increase the strength of low-alloy steels, it is necessary to consider solid-solution hardening with alloying with cheap alloying elements (Mn, Si), as well as dislocation and dispersion hardening by applying thermomechanical processing in combination with micro-alloying additives of carbide and nitride-

forming elements (V, A). A comparison of the calculated values of the yield strength of steel with its value in the corresponding GOST-ah shows a satisfactory difference, which indicates the applicability of a quantitative assessment of the yield strength of steel according to the parameters of the formed structure after certain treatments and provides reliable information about the existing hardening mechanisms.

Analysis of the data of quantitative assessment of the yield strength of carbon and low-alloy steels

by structure parameters shows that the main mechanisms of their hardening are solid-solution hardening by alloying with relatively cheap alloying elements (Mn, Si), as well as dislocation and dispersion hardening using hardening heat treatment and micro-alloying of steel with carbide and nitride-forming elements V (C, N).

Conflict of interests. On behalf of all authors, the correspondent author declares that there is no conflict of interests.

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Көміртегі аз және легирленген болатқа қатаю механизмдерінің қолданылуы

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| <p>Мақала келді: 12 қазан 2021 Сараптамадан өтті: 22 қараша 2021 Қабылданды: 03 ақпан 2022</p> | <p>ТҮЙІНДЕМЕ Эксперименттік зерттеулер негізінде әр түрлі қатаю механизмдерінің көміртегі аз және легирленген болаттардың түсу нүктесіне шамамен үлесі бағаланады. Ыстық илектелген болаттар үшін (Ст3сп және Ст5пс) қатты ерітінді мен дән шекарасының беріктігі (54,0% және 29,0, %) шығыс нүктесіне үлкен үлес қосатыны анықталды. 10ХНДП аз легирленген болаттың басым күшейту механизмі-бұл қатты ерітінді, оның болаттағы үлесі α-Fe-де Ni, Si, P және Cr атомдарының еріген жылжымалы дислокациясына төзімділігімен түсіндіріледі. Төмен легирленген 16Г2АФ болатта осы қатайтатын компоненттермен бірге жауын-шашынның қатаюының рөлі байқалады (20,0%). Ст5пс болат сыныбының термомеханикалық өңдеуі дислокация тығыздығының жоғарылауы мен дислокацияның көп бөлігін прокат өнімдерінде ұстау есебінен дислокацияның қатаю мәнінің 27,0% дейін өсуіне әкелетіні көрсетілген ыстық деформацияланған аустенит. Арзан легирлеуші элементтермен (Mn, Si) легирлеумен қатты ерітіндінің қатаюы, сондай-ақ V (C, N) карбиді мен нитрид түзуші элементтерді қосумен термомеханикалық өңдеуді қолдану арқылы дислокация мен дисперсияның қатаюы байқалады. Түйін сөздер: қатаю механизмдері, түсу кернеуі, термомеханикалық өңдеу, тездетілген салқындату, дислокация тығыздығы, фазалық компоненттер.</p> |
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О применимости механизмов упрочнения к малоуглеродистым и низколегированным сталям

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АННОТАЦИЯ

На основе экспериментальных исследований оценен ориентировочный вклад различных механизмов упрочнения в предел текучести малоуглеродистых и низколегированных сталей. Установлено, что для горячекатаных сталей (Ст.3 сп и Ст5пс) наибольший вклад в предел текучести дают твердо-растворное и зерно-граничное упрочнения (54,0% и 29,0, %). Преобладающим механизмом упрочнения низколегированной стали 10ХНДП является твердорастворный, высокая доля которого в этой стали объясняется сопротивлением движущимся дислокациям со стороны растворенных атомов Ni, Cu, P и Cr в α -Fe. В низколегированной стали 16Г2АФ наряду с этими слагаемыми упрочнения заметна роль дисперсионного упрочнения (20,0%). Показано, что термомеханическая обработка стали марки Ст.5пс приводит к росту величины дислокационного упрочнения до 27,0 % за счет роста плотности дислокаций и сохранения большей части дислокаций в прокате при ускоренном охлаждении горячедеформированного аустенита. Отмечено, что в качестве эффективных и перспективных способов повышения прочности низколегированных сталей является твердорастворное упрочнение с легированием дешевыми легирующими элементами (Mn, Si), а также дислокационное и дисперсионное упрочнение путем применения термомеханической обработки в сочетании с добавками карбидо- и нитридообразующих элементов V (C, N).

Ключевые слова: механизмы упрочнения, предел текучести, термомеханическая обработка, ускоренное охлаждение, плотность дислокаций, фазовые составляющие.

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