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


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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 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 ferrito-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:

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

Where $\sigma_i$ 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 $\mu$m; 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 $\sigma_i$ in the Hall-Petch strengthening can be represented as a sum:

$$\sigma_i = \sigma_o + \sigma_{SS} + \sigma_p + \sigma_d + \sigma_{ph}$$  \hspace{1cm} (2)

The lower limit of steel yield followed from equations (1) and (2) is characterized by the lattice friction stress $\sigma_o$, solid solution strengthening $\Delta \sigma_{SS}$, hardening due to pearlite formation $\sigma_p$, strain hardening $\sigma_d$, dispersion hardening $\sigma_{ph}$ 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>
<thead>
<tr>
<th>№</th>
<th>Hardening factors</th>
<th>Calculation formula of hardening</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>Lattice friction stress $\alpha$-Fe</td>
<td>$\sigma_o = 2 \cdot G \cdot 10^{-4}$</td>
</tr>
<tr>
<td>2</td>
<td>Solid solution alloying</td>
<td>$\Delta \sigma_{SS} = \sum_{i=1}^{n} k_i \cdot c_i$</td>
</tr>
<tr>
<td>3</td>
<td>Hardening due to the formation of perlite</td>
<td>$\Delta \sigma_p = 2,4% \cdot \Pi$</td>
</tr>
<tr>
<td>4</td>
<td>Deformation hardening</td>
<td>$\Delta \sigma_d = 0,5 \cdot G \cdot b \cdot p^{1/2}$</td>
</tr>
<tr>
<td>5</td>
<td>Dispersion hardening</td>
<td>$\Delta \sigma_{ph} = (9,8 \cdot 10)/\lambda \cdot \ln(2\lambda)$</td>
</tr>
<tr>
<td>6</td>
<td>Intergranular (substructural) hardening</td>
<td>$\Delta \sigma_{gb} = k_h \cdot d^{-1/2}$</td>
</tr>
</tbody>
</table>

$$\Delta \sigma = k_s \cdot I^{-m}$$
Детерминация параметров микроархитектуры (содержание перлита, диаметр зерен феррита, размер и объемное содержание дисперсных частиц, др.) для количественной оценки предела текучести компонентов структуры методами описанными на Нейфоте 21 исследовательском микроскопе и электронном микроскопе UEMV-100V. Средняя длина прямого сечения, проходящего через зерно феррита (d) и дисперсной частицы (D) в низколегированной стали 16G2AF были определены при помощи электронной микроскопии, и дистанция между частицами (λ) определена по известному уравнению:

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

Доля перлита в структуре определена по методу Росивал. Согласно которому, для оценки эффективности различных механизмов закалки, результаты рассчитаны в форме круговой и столбчатой диаграмм, приведены в табл. 1 [3], [4]. Для удобства сравнения и анализа эффективности различных механизмов закалки, результаты расчетов приведены в виде круговых и столбчатых диаграмм, представлены в виде гистограмм (рис. 1-5). В стальных ствах St3sp, St5sp (закалка в жаропрочном состоянии), основные компоненты закалки являются жесткими и граничные механизмы закалки, затрагивающие 54% и 29%, соответственно (рис. 1). Они равны в абсолютном значении: 140.5 МПа и 89.9 МПа.

Таблица 2 - Изначальные данные для количественной оценки предела текучести исследованных сталей

<table>
<thead>
<tr>
<th>№</th>
<th>Характеристики стали</th>
<th>Сталь</th>
<th>St3sp</th>
<th>St3sp (закалка)</th>
<th>St5sp (термообработка)</th>
<th>16G2AF</th>
<th>10KHNDP</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>Содержание легирующих элементов в α-Fe, %</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td></td>
<td>Mn</td>
<td>0,52</td>
<td>0,65</td>
<td>0,65</td>
<td>1,5</td>
<td>0,45</td>
<td></td>
</tr>
<tr>
<td></td>
<td>Si</td>
<td>0,21</td>
<td>0,11</td>
<td>0,11</td>
<td>0,45</td>
<td>0,27</td>
<td></td>
</tr>
<tr>
<td></td>
<td>P</td>
<td>0,04</td>
<td>0,04</td>
<td>0,04</td>
<td>0,035</td>
<td>0,095</td>
<td></td>
</tr>
<tr>
<td></td>
<td>V</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>0,11</td>
<td>-</td>
<td></td>
</tr>
<tr>
<td></td>
<td>Ni</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>0,45</td>
<td></td>
</tr>
<tr>
<td></td>
<td>Cr</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>0,65</td>
<td></td>
</tr>
<tr>
<td></td>
<td>Cu</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>0,40</td>
<td></td>
</tr>
<tr>
<td></td>
<td>(C+N)</td>
<td>0,015</td>
<td>0,015</td>
<td>0,015</td>
<td>0,015</td>
<td>0,015</td>
<td></td>
</tr>
<tr>
<td>2</td>
<td>твердость</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>V(CN)</td>
<td>-</td>
<td></td>
</tr>
<tr>
<td>3</td>
<td>Пористость, %</td>
<td>22</td>
<td>35</td>
<td>26</td>
<td>17</td>
<td>14,3</td>
<td></td>
</tr>
<tr>
<td>4</td>
<td>Размер зерна, мм</td>
<td>0,056</td>
<td>0,051</td>
<td>0,033</td>
<td>0,014</td>
<td>0,028</td>
<td></td>
</tr>
<tr>
<td>5</td>
<td>Объемное содержание дисперсных частиц (f), %</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>0,096</td>
<td>-</td>
<td></td>
</tr>
<tr>
<td>6</td>
<td>Размер дисперсных частиц (D), нм</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>30</td>
<td>-</td>
<td></td>
</tr>
<tr>
<td>7</td>
<td>Расстояние между частицами (X), нм</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>765</td>
<td>-</td>
<td></td>
</tr>
<tr>
<td>8</td>
<td>Дислокационная плотность (ρ), 10³ см⁻²</td>
<td>10³</td>
<td>10⁸</td>
<td>10¹⁰</td>
<td>10⁸</td>
<td>10³</td>
<td></td>
</tr>
</tbody>
</table>

Примечание: Количественное определение плотности дислокаций и объемного содержания фазовых компонентов является трудной задачей, поэтому данные были взяты из надежных литературных источников.
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
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:

\[ K_{N_i} = 30; K_{Cu} = 40; K_{P} = 690; K_{Cr} = 30. \]

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), \[ \Delta \sigma_{p.h} = 94.0 \text{ MPa}. \] 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 the 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 \[ \Delta \sigma_p = 75 \text{ MPa} \] 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-
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.
О применимости механизмов упрочнения к малоуглеродистым и низколегированным сталям

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АННОТАЦИЯ
На основе экспериментальных исследований оценен орентировочный вклад различных механизмов упрочнения в предел текучести малоуглеродистых и низколегированных сталей. Установлено, что для горячекатаных сталей (Ст.3 с и Ст5с) наиболее значимый вклад в предел текучести дает твердо-растворное и зерно-граничное упрочнение (54,0% и 29,0, %). Преобладающим механизмом упрочнения низколегированной стали 10ХНД является твердо-растворный, 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.5с 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 твердо-растворное упрочнение с легированием дисперсными легирующими элементами (Мn, 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.

References


