THERMOMECHANICAL STRENGTHENING OF MIDDLE CARBON STRUCTURAL STEEL USING COLD DEFORMATION

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Abstract

The effect of Thermomechanical Processing, using cold deformation with the combination of post deformation annealing and induction hardening based on the of dislocation structure inheritance effect on the strength, ductility, and torsion static strength has been studied. The metal with an initial structure of lamellar pearlite has shown to be more advantageous with respect to the strengthening intensity during cold rolling and the combination of mechanical properties after the TMP. It can be seen that in the studied strain range (up to $\varepsilon_\Sigma \cong$ 60 %) the hardness of the high-strength middle carbon steel with the structure of lamellar pearlite is 20 - 40 % higher than that of the steel with the structure of granular pearlite. Steel with the initial structure of lamellar pearlite is strengthened more intensely and its final heat treatment with the use of induction hardening to martensite and low temperature tempering show the inheritance of fragmented dislocation substructure result in the highest static torsion strength and the best combination of other mechanical properties.

Keywords: Thermomechanical strengthening, cold deformation, dislocation substructure, middle carbon low alloyed steels

1. INTRODUCTION

By now experimentally proved the states of the complex impact, consisting of the operation of heating, cooling and deformation carried out in different sequence, resulting in the formation of the final structure and properties of metallic alloy occurs in the conditions of high density and corresponding distribution of imperfections in the structure created by plastic deformation, and is the essence of thermomechanical processing (TMP) [1-5]. The difference between the schemes TMP is determined essentially predominate type resultant substructure [1, 3, 4]. This can be either a substructure strengthening, when the dislocations are distributed through the body of grain uniformly or non-uniformly, concentrating in bulk low-angle boundaries and forming a cellular or polygonal substructure when planar dislocation sub-grains break the grain at a relatively defect-free volume-sub-grains or fine-grained recrystallized structure. It should be noted that the effect of TMP can only be eliminated by a collective recrystallization (grain coarsening). If the recrystallization process has not reached that stage, the final structure is formed by high density of imperfections in the structure, which certainly include grain and sub-grain boundaries, we have retained the essence of TMP [3]. A special place among the schemes of TMP are the so-called hereditary methods of thermomechanical strengthening. The meaning of inheritance is as follows. By plastic deformation and subsequent short time softening in the material creates a high density of imperfections (mostly-dislocations)in their specific configuration - the fragmented substructure. If during the subsequent heat treatment, the density of imperfections (dislocations) will not be noticeable to decrease, and fragmented substructure will not disappear, will be preserved and a high level of mechanical properties [4, 5]. The success of this treatment primarily depends on how stable dislocation configurations were created during plastic deformation. Therefore, the effect of inheritance is evident after TMP with determined temperature-strain-strain rate parameters under combination of deformation and heat treatment. It can be expected from the phase hardening, for example, as a result of repeated quenching with a high heating rate, which excludes the annihilation of defects. However, this treatment does not provide the optimal distribution of defects (dislocations), therefore at such hardening is not achieved the level of plasticity, as a result TMP. Currying out of cold plastic deformation before quenching with intermediate heating to the special stable substructure also allows to improve the structural strength of steel [4,6]. In the literature there is no unambiguous explanation for the observed effect. Some authors attribute the resulting strengthening with the manifestation of the inheritance of cold work hardening, when the subsequent processing is carried out under conditions that exclude recrystallization [6]. The paper emphasizes the importance of holding after deformation polygonizing tempering for a more complete manifestation of the effect of inheritance.

The objective of this work was to study the effect of Thermomechanical Strengthening of middle carbon structural steel using cold deformation in the specific TMP scheme. It was studied the structure and mechanical properties of a middle carbon low alloyed steels formed as a result of cold plastic deformation and heating carried out for different modes, to assess the possibility of the application of cold rolling in the scheme of TMP. The special attention was devoted to preparation of the initial structure of the steel before deformation. It is known that the initial structure affects the deformability and hardenability of carbon steels and the formation of their final properties. So, at the first stage there were studied the effect of the initial structure of medium-carbon (0.4 - 0.5 % C) low alloyed Cr-Ni (1-2 %) steel.

2. EXPERIMENTAL

There were studied initially hot-rolled products with a preliminary heat treatment to the lamellar pearlite (mode 1) represented by alternating lamellas of ferrite and cementite (about 70 %), free ferrite (about 20 %), and globular pearlite (about 10%) (**Fig. 1a**) and granular pearlite (mode 2), (**Fig. 1b**). The treated to the mentioned above structure billets were cold rolled in rolling mill with a total reduction $\varepsilon_\Sigma \cong 48$ % (~5 % reduction per pass) to the cylindrical bars. After deformation, the bars were annealed at 300-800°C to find the optimal temperature for thermomechanical strengthening in the proper treatment scheme. The final heat treatment consisted of quenching with a rapid induction heating to austenitization temperature with smallest austenite grain size (900-950 °C), followed by accelerate cooling for martensite and low temperature tempering (200 °C-for steel with 0.4 %C). The initial structure was studied by TEM at accelerating voltage of 175 kV. Mechanical properties were compared for steel treated by the following variants: 1) oil quenching after heating in the furnace with heating rate at 2-5 K/s to 870 °C, and tempering at 200 °C (without preliminary cold deformation), 2) oil quenching after induction heating with heating rate 1-2 °C/s and tempering at 200 °C (without preliminary cold deformation), and 3) TMP including cold deformation with a total reduction $\varepsilon_\Sigma \cong 48$ %, annealing at 500 °C, oil quenching after induction heating and tempering at 200 °C. The resistance of the metal to brittle fracture after TMP was evaluated in terms of the threshold of the ductile-brittle transition determined from the results of uniaxial static tensile tests at various temperatures. The tensile mechanical tests were carried out in the temperature interval from - 196 °C to + 20 °C. The fracture surface of the specimens was studied under Tesla BS-300 SEM at accelerating voltage of 35 kV.

3. RESULTS AND DISCUSSION

After annealing by mode 1 the structure of the steel consists of lamellar pearlite represented by alternating lamellas of ferrite and cementite (70-80 %), free ferrite (about 20-30 %), and globular pearlite (less than 10 %) (**Fig. 1a**). As a rule, the cementite contained in pearlite has the form of thin lamellas. In some cases the cementite lamellas have a “sawtooth” shape typical for alloy steels. In some colonies carbides segregate in the form of short and thinner plates. In addition to lamellar cementite the structure may contain segregations of globular cementite concentrated both over the boundaries and inside ferrite grains. The size of the segregations fluctuates from 0.2 to 0.6 μm. The dislocation density in the ferrite component is not high and does not exceed 5 x 10^8 cm^{-2}.

After annealing by mode 2 the structure is represented by granular pearlite. The carbides formed due to such annealing have a predominantly spherical shape (see **Fig. 1b**) in contrast to the carbides formed as a result of annealing by mode 1.

Cold rolling distorts the lamellar structure of the pearlite, which is manifested in bending of cementite lamellas, their crushing into parts, and rotation of individual lamellas. The spherical carbide particles undergo no visible change. The matrix becomes fragmented, i.e., crystals of the α-phase break into regions less than 1 μm size. A cellular structure forms in ferrite lamellas and pearlite colonies. Dislocation walls primarily form perpendicularly to cementite lamellas and their width is often commensurable with the size of the cells. Cementite lamellas are accompanied by accumulations of bent dislocations. A cellular structure forms in the regions of excess (free) ferrite.

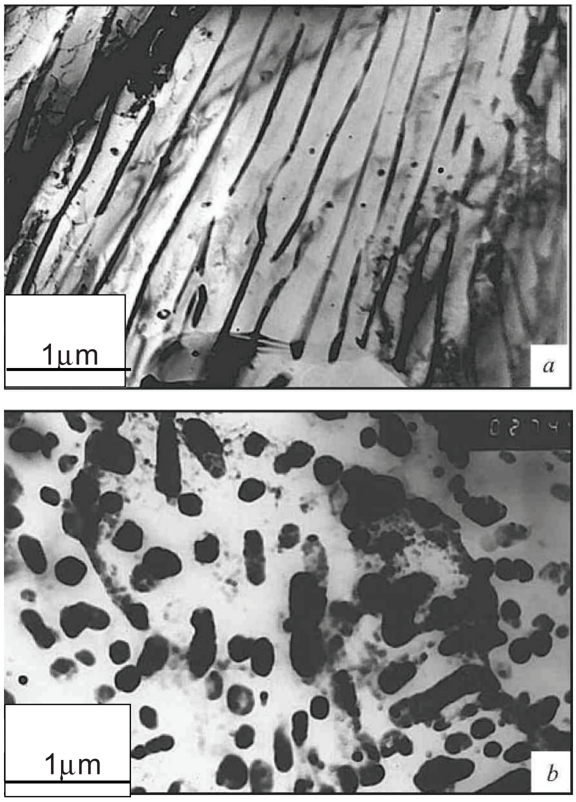

Fig. 1 Structure of steel 0.4CCrNi after annealing for lamellar (*a*) and granular (*b*) pearlite

Annealing of cold-deformed steel at a temperature of up to 400 °C causes regrouping and partial annihilation of dislocations and the beginning of formation of low-angle boundaries. The random distribution of dislocations in the ferrite is partially removed. The dislocations form subboundaries in the ferrite layers of lamellar pearlite. An increase in the post deformation annealing temperature to 500 °C increases the volume fraction of polygonized ferrite and decreases the fraction of cellular structure, which promotes spheroidization of the carbide phase. The largest fraction of polygonization mode and the beginning of recrystallization of ferrite were observed result in annealing at 600 °C. Recrystallization nuclei free of dislocations and separated from the surrounding matrix by high-angle boundaries formed all over the volume. Individual recrystallized grains have a size of about 5 μm. Cementite undergoes coagulation; after such heating, lamellar particles are not observed. Annealing at 700 °C causes the development of recrystallization. New ferrite grains virtually free of dislocations are formed in the whole volume of the specimen. Their size does not exceed 10 μm. Spherical carbide particles are located on the boundaries and inside the grains of the α-phase. The impact strength depends on the mode of annealing after cold rolling and on the preheating rate for quenching. The highest values of the impact strength were obtained in the case of induction heating for quenching; after rolling, annealing at 500 °C, and such quenching with low-temperature tempering(variant *3*) the value of *KCV* = 0.49 MJ/m^2(for variant *2: KCV* = 0.40 MJ/m^2). A similar dependence was observed in tensile tests. After the treatment involving cold rolling + annealing at 500 °C + quenching from induction heating + low-temperature tempering the ductility was the highest (RA = 55 %) in compare with bars treated without preliminary cold rolling (RA = 48 %). The structure of the steel after treatment by different variants is presented in **Fig. 2**.

After quenching with furnace heating and low-temperature tempering of non-deformed steel, the latter acquires the structure of tempered martensite (**Fig. 2*a***). We observe lath martensite with differently oriented packets. The grains bear large crystals of lamellar martensite. Plate carbide segregations with several crystallographic orientations are observed both inside large lamellas and inside the laths. Rare carbides are at most 0.06 μm in size. After quenching from induction heating and low-temperature tempering (**Fig. 2*b***), the structure of the non-deformed steel includes a great amount of carbides up to 0.1 μm in size. The dispersity of martensite crystals increases. In the structure provided by cold deformation, quenching from induction heating, and low-temperature tempering (**Fig. 2*c***), the degree of martensite crystals dispersity is lower. Annealing of cold-deformed steel at 500 °C for 3 h promotes the formation of a polygonal structure. In this case, the final heat treatment yields fine carbides (the size of the observed carbides does not exceed 0.08 μm) and more dispersed martensite crystals; the dispersity of the carbides formed in the low-temperature tempering increases as well (**Fig. 2*d***). When the annealing temperature is increased to 600 °C, the dispersity of the structure decreases (**Fig. 2*e***).

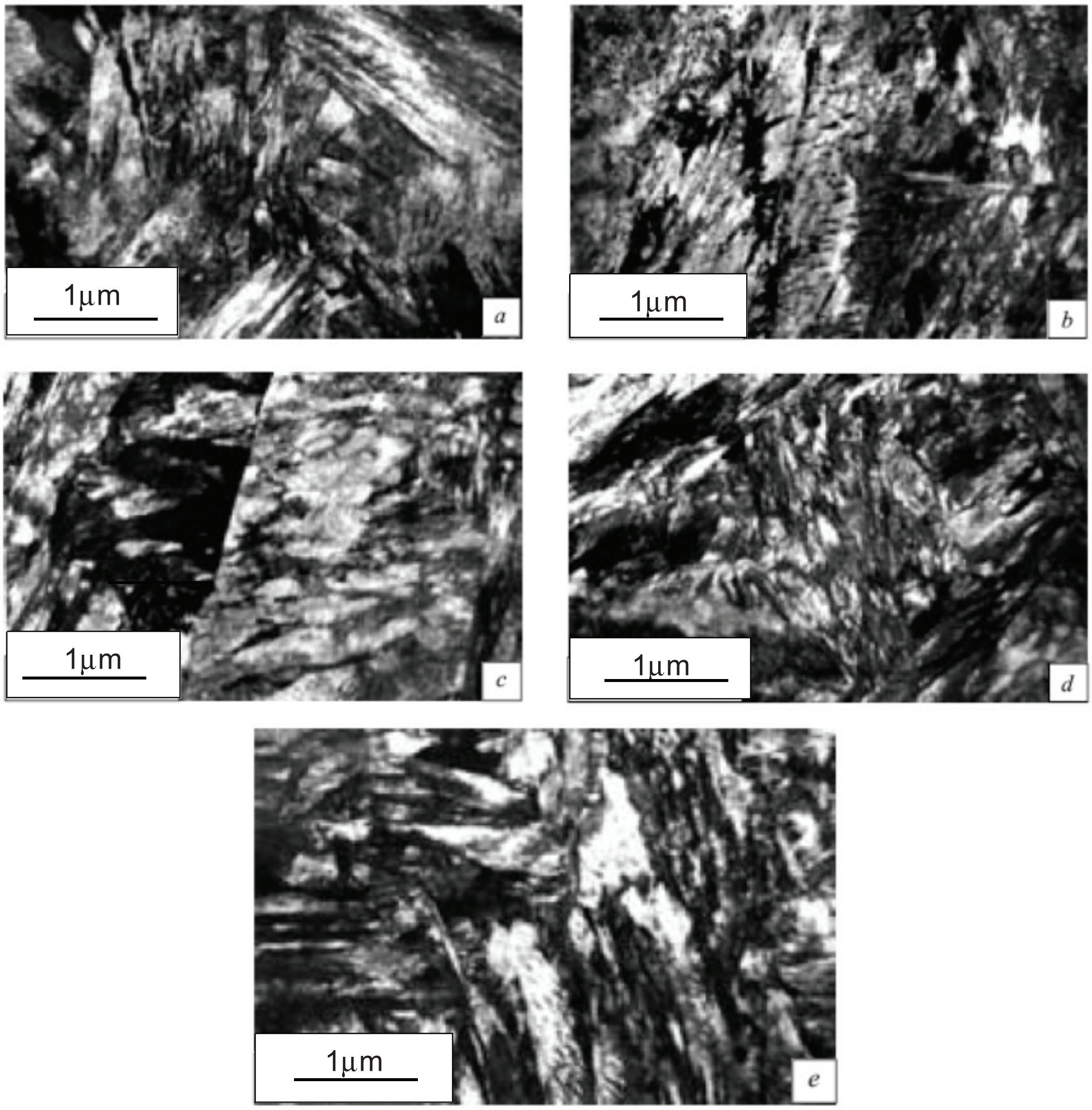

Fig. 2 Structure of steel 0.4CCrNi after: quenching of non-deformed billets from furnace and induction heating respectively (*a*, *b*); cold rolling before quenching from induction heating (*c*); cold rolling and annealing respectively at 500 and 600 °C before quenching and induction heating (*d*, *e)*

The temperature dependences of the mechanical properties are presented in **Fig. 3**. With decrease in the test temperature the values of tensile strength (TS) and yield strength (YS) tend to increase, which is typical for high-strength steels. The reduction in area (RA) somewhat decreases. At the temperature of the ductile-brittle transition the value of R decreases abruptly to several percent (**Fig. 3a, b**). It should be noted that the temperature of the ductile-brittle transition is - 170 and - 196 °C after treatment by variants *1* and *3*, respectively. Analysis of the fracture surfaces of steel 0.4CCrNi after TMP has shown (**Fig. 4**) that a circular asperity enclosing the detachment surface is visible in the entire fracture surface.

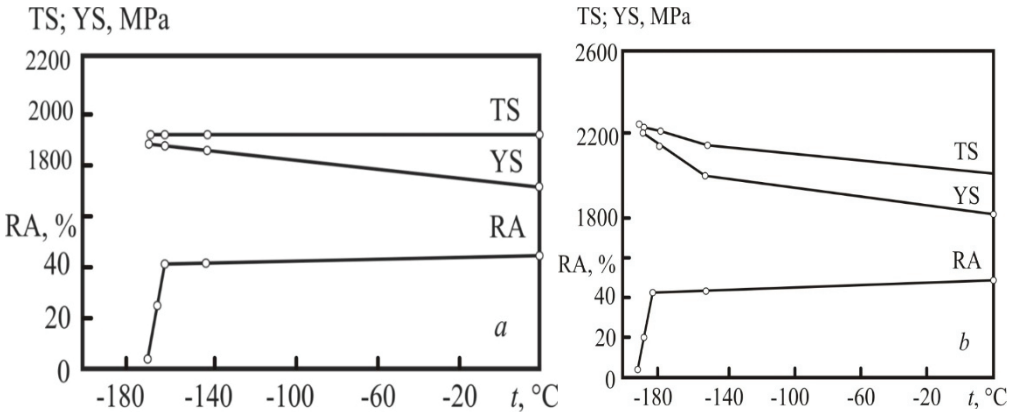

Fig. 3 Temperature dependences of mechanical properties of steel 0.4CCrNi: treatment by variant *1*(*a*); treatment by variant *3* (TMP) (*b*)

The cross section of the asperity has a shape resembling an equilateral triangle (**Fig. 4*a*** and ***b***). All the detachment surfaces on this circular asperity differ markedly from the internal surface of the fracture in smoothness. Under high magnifications (**Fig. 4*c***) the structure of these surfaces has a ductile nature and consists of a great number of dimples from 0.3 to 2 μm in size.

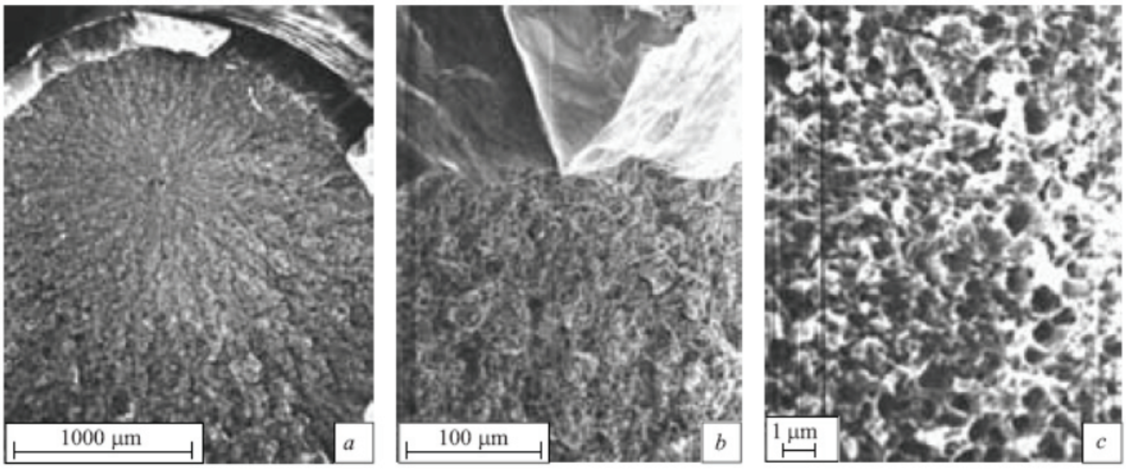

Fig. 4 Fracture surfaces (t_{test} = - 196 °C) of steel 0.4CCrNi after TMP (variant *3*)

After TMP with post-deformation annealing at 500 °C, the values of characteristics of static torsion tests: proportional limit (τ_{pr}) , yield strength at torsion ($\tau_{0.3}$) , and maximum residual shear (Υ_{max}) were higher than after the treatment by variant *1* (Υ_{max} was higher by 40 %). This effect allows to assume that the developed ferrite structure created by cold rolling is preserved after the double $\alpha \rightarrow \gamma \rightarrow \alpha$ transformation result in inheritance of dislocation structure realization provides a homogeneous and stable dislocation structure and thus enhances the effect of the TMP.

4. CONCLUSIONS

1) The cold rolling in the TMP scheme is an efficient method to increase structural strength of middle carbon low alloyed structural steel production.
2) Steel with the initial structure of lamellar pearlite is strengthened more intensely at the regimes of TMP with cold deformation followed by heating at 500 °C, induction hardening and low temperature tempering.

3) TMP with cold deformation followed by heating at 500 °C, induction hardening and low temperature tempering raises the resistance of the steel 0.4CCrNi to brittle fracture.

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