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EFFECT OF CONTROLLED ROLLING ON THE STRUCTURAL AND PHASE TRANSFORMATIONS

The development of ferrous metallurgy is mainly due to the requirements of the leading metal-consuming industries to improve the performance properties of structural steels to increase the permissible loads, to reduce metal consumption and to improve the reliability of machines, structures, main gas pipelines, *etc*. With significant volumes of rolled metal production, reducing energy consumption and consuming raw materials during its production also come to the fore. In this regard, important and relevant works are the works aimed at creating the fundamentals of metals science and the development of new technologies that allow manufacturing the products with the required combination of properties directly in the rolling mill stream (excluding subsequent heat treatment) by purposefully controlling the processes of structure formation, as well as expanding the areas of the practical application of such technological schemes (in terms of brand and size assortment, rental destination, *etc*.). Such an approach makes it possible to increase the competitiveness of metal products determined by the achieved combination of metal characteristics, while reducing the cost of its production.

Keywords: plastic deformation, controlled rolling, structure, phase, transformations.

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1. Introduction

Thermomechanical treatment (TMT) combines plastic deformation and thermal action to form the required structure of the processed metal. The combination of plastic deformation and heat treatment operations, their maximum convergence, and the creation of a single thermomechanical processing process provide a noticeable increase in mechanical characteristics (strength, viscosity, *etc.*), which allows to save up to 15–40% of metal or more, or increase the durability of products.

With TMT, both processes, — plastic deformation and heat treatment, — can be combined in one technological operation or carried out with a time gap. Nevertheless, simultaneously, the passage of phase transformations under conditions of increased density of lattice defects arising due to plastic deformation of the metal is a prerequisite.

Thermomechanical processing of steel is carried out mainly according to three schemes: high-temperature (HTMT), low-temperature (LTMT), and preliminary thermomechanical treatment (PTMT). Thermomechanical processing also includes technologies of controlled rolling and accelerated cooling.

We consider a controlled rolling as the most popular technology in recent years. Controlled rolling is a kind of HTMT and is an effective way to increase low-alloy steels' strength, ductility, and a viscosity [1–5].

Controlled rolling can be called optimized heating and deformation by rolling, which ensure the production of refined ferritic grains because of two mechanisms: through refined recrystallized austenitic grains formed during hot rolling in the medium temperature range, and through the deformation of austenite below the recrystallization temperature, which enhances the nucleation of ferritic grains. When implementing controlled rolling, the variable parameters are considered to be the heating temperature, the deformation temperature, in particular, the end of deformation, the total and single (in 1 pass) degree of compression, the rolling speed, the number of passes and the duration of pauses between them, the modes after deformation cooling.

The main goal of controlled rolling is to obtain an ultrafine-grained (12–13 points) final ferrite–perlite or ferrite–bainite structure in hot-rolled products. We can achieve this goal when, as a result of rough rolling and cooling to the temperature of the beginning of finishing rolling, a fine grain of austenite is formed in the workpiece (10–11 points). Since it is crushed due to the austenite recrystallization, this process kinetics is of decisive importance and has been studied by many researchers [6–8]. Almost all of them concluded that in steels containing minor additives of carbonitride-forming solid elements, the austenite recrystallization after rolling at 950–1000 °C proceeds very slowly.

One of the most important, in practical terms, is the question of ferrite nucleation in deformed austenite since the nucleation process largely determines the final structure dispersion. Therefore, it has been studied by many domestic and foreign scientists [9–11].

According to Ref. [12], there are 8 stages of the technology of controlled sheet rolling that follow below.

Austenitization at temperatures that ensure a sufficiently homogeneous metal structure before rolling. At the same time, the heating temperature for rolling for most steels microalloyed niobium, vanadium, and titanium is about 1150–1200 °C.

High-temperature deformation of stable austenite in the region of fast-flowing recrystallization processes when the deformation temperature is higher than the recrystallization temperature. High-temperature deformation aims to obtain as fine a grain of austenite as possible by alternating multiple compressions and recrystallization. For low-alloy steel with niobium, the degree of deformation for the development of dynamic recrystallization at 1100–1150 °C is 40–60%. However, the implementation of such regimes in industrial conditions is complex. Rough rolling on existing thick-sheet mills is carried out at temperatures not lower than 980–1000 °C with degrees of compression per pass of 15–20%.

Medium-temperature deformation in the austenitic region's lower part is carried out to increase the density of defects in the crystal structure of the metal and their ordered distribution (substructure), which leads to multiple formations of ferritic volumes during polymorphic $\gamma \to \alpha$ transformation.

Deformation of austenite in the region of polymorphic $\gamma \to \alpha$ transformation.

Deformation in the two-phase $\gamma + \alpha$ region. A decrease in the deformation temperature in the $\gamma + \alpha$ region contributes to the hardening of steel since, at the same time, the proportion of ferrite grains hardened by deformation increases.

Deformation in the three-phase region is advisable if priority is given to obtaining a very high strength in the complex mechanical properties.

Deformation below the Ar_1 point is possible in the presence of powerful rolling equipment and low requirements for the plastic characteristics of rolled products.

After the deformation is completed, the steel is cooled in the air at a cooling rate of 0.5–1 °C/s or in laminar cooling units of rolled products at a speed of about 15 °C/s [12].

In foreign and domestic practice, controlled sheet rolling is carried out in two or three stages in industrial conditions [13–16]. The first stage begins with the heating temperatures for rolling up to 950–1000 °C, when rapid recrystallization of austenite occurs, ending within the intraformational pause.

At the second stage of the process, in the temperature range from 950 °C to 3, recrystallization in conventional steels is complex, and in steels with niobium, it is practically suppressed. In this area, the temperature and degree of deformation significantly affect the kinetics of recrystallization. The grain decreases due to static recrystallization if the temperature and deformation conditions are chosen correctly. Deformation leads to the formation of slip bands and the separation of dispersed phases. With an increase in the degree of deformation, the number of sliding strips increases, and the uniformity of their arrangement increases that contributes to obtaining a fine, uniform ferrite grain after transformation. As a result, the cold resistance of steel is increased.

In the third stage, at temperatures below the phase transformation temperature (A_{r3}), the processes of dispersion hardening with grain grinding are accompanied by the development of subgrain texture. The last two factors are crucial in improving the properties of steel.

The main difference between conventional and controlled rolling is that deformation strips divide austenitic grains into several blocks during controlled rolling. The boundary of each block is the source of the nucleation of ferrite grains. As a result, smaller ferritic grains are formed from austenitic grains of the same size during controlled rolling than during conventional hot rolling, when the nucleation of ferritic grains is carried out at the boundaries of austenitic. In addition, an increase in the number of active nucleation centres of ferrite accelerates the α-transformation process. As a result, the probability of separation of the bainite structure, which gives low viscosity to steel, decreases.

In conventional hot-rolled and normalized steels, ferrite originates exclusively at the boundaries of austenitic grains, which limits grain grinding. In rolled according to controlled conditions or hardened steels, austenitic grain is divided into several blocks: in the first case by deformation strips, in the second by martensitic plates. In this case, the size of ferritic grains is determined not by the size of austenite grains but by the size of the blocks formed.

The authors of Ref. [17] indicate four factors that cause the grinding of ferritic grains during controlled rolling. Firstly, this is a decrease in the heating temperature for rolling, leading to a decrease in the initial size and recrystallized austenitic grain. The second factor is the deceleration of recrystallization during hot deformation, which can be achieved in various ways: by lowering the temperature and increasing the degree of deformation, increasing the recrystallization temperature due to doping of the solid solution, separation of dispersed particles of the second phase from the solid solution before recrystallization or during its process, inhibiting the migration of grain boundaries and blocks. The third factor is lowering the temperature of the

$\gamma \rightarrow \alpha$-transformation, which can be achieved both by appropriate alloying and by regulating the rate of post-deformation cooling.

Finally, an essential factor is to prevent the growth of ferritic grains in the upper part of the ferritic region, for example, when cooling steel in rolls. A fine ferritic grain can also be obtained from recrystallized or partially recrystallized austenitic grain with a high density of defects in the crystalline structure due to the nucleation of ferrite on defects inside the austenitic grain. These four mechanisms do not exhaust all the possibilities of grinding ferritic grains during controlled rolling.

The technology of controlled rolling consists of such a choice of rolling modes and cooling after it, which will ensure the production of fine and homogeneous grain in the finished rolling, which, in turn, will provide a higher level of mechanical properties. The most frequently controlled rolling is used in the production of sheets.

Thus, the state of the issue a few years ago was characterized by relatively limited use of controlled rolling, which made it challenging to develop and implement its new schemes, expand its practice areas of application, did not allow to formulate general approaches to the selection of the controlled rolling technological scheme and steel composition based on a wide range of structures and hardening mechanisms used.

Understanding the various factors influencing the formation of mechanical properties of metal in the manufacture of rolled products in modern industrial conditions is an urgent practical task. The study of the regularities of changes in mechanical properties in the process of controlled rolling is the fundamental basis for presenting reasonable requirements for rolling when creating regulatory and technical documentation, predicting the behaviour of metal during moulding, and achieving the required set of mechanical properties.

2. Existing Technologies of Controlled Rolling

In modern sheet-rolling production, the tendency to expand the output of steel products with unique combinations of service properties is dynamically increasing. For example, Fig. 1 shows one of the possible modern classifications of types of cold-rolled structural steel used by the overwhelming number of specialists involved in the aircraft industry [18].

With a more general approach, the types of steel indicated in Fig. 1 are considered steel for producing products using cold deformation. When separating these types of steel according to strength indicators, three classes are traditionally distinguished — a class with reduced strength (LSS, yield strength is limited to 210 MPa with a maximum tensile strength of 270 MPa), a class with high strength (HSS, 210–550/270–700 MPa) and a class of steels with exceptionally high strength

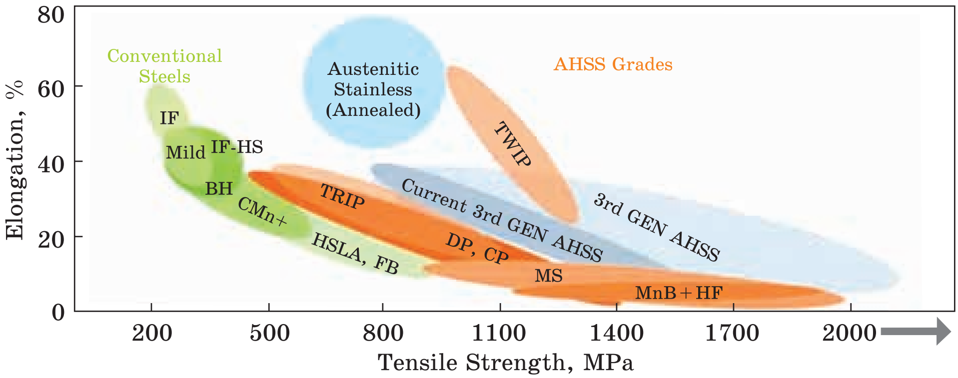


Fig. 1. Graphic representation of the classification of modern types of steel for cold forming [18]

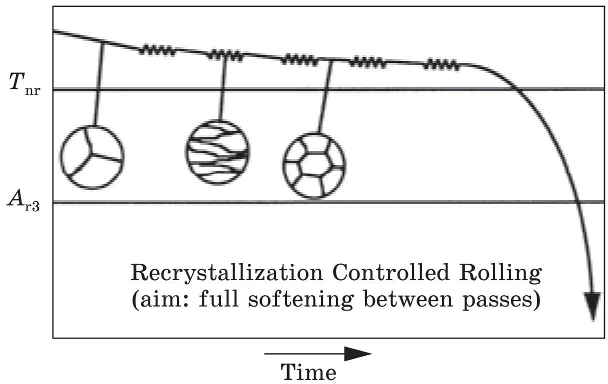


Fig. 2. Schematic time-temperature diagram of recrystallization controlled rolling showing the recrystallized microstructure between passes. T_{nr} (A_{r3}) — the no-recrystallization (phase transformation) temperature [20]

(UHSS, more 550/700 MPa). Along with the three classes described, there are traditional high-strength steels (Conventional HSS) and promising (Advanced HSS) high-strength steels. Promising high-strength steels have, as a rule, exceptionally high strength (designation U/AHSS) and, in their development, have passed through three stages (or generations), differing from each other by the types of structures used in their structure [18].

The purpose of controlled rolling is to obtain optimal ferrite grinding. Therefore, during processing, it is necessary to maximize the area of austenite grain boundaries per unit volume when phase transformation begins [19]. There are several ways to achieve this goal, the main ones being rolling with recrystallization control, conventional controlled rolling, and dynamic rolling with recrystallization control [20]. The type of controlled rolling is determined by the conditions at the last rolling passes, *i.e.*, in the finishing stands.

Periodic recrystallization of austenite provides gradual grinding of recrystallized grains. Rolling with recrystallization control means that comp-

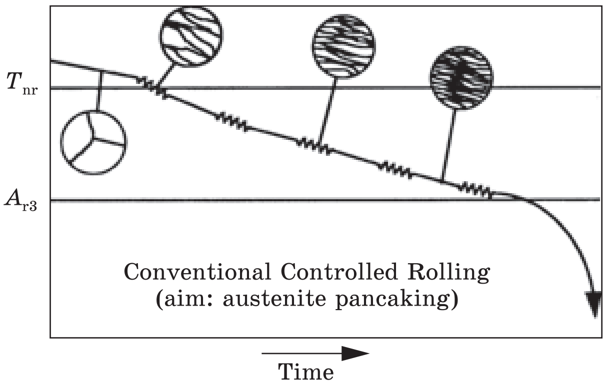


Fig. 3. Schematic time-temperature diagram of conventional controlled rolling, showing the deformed microstructure between passes

lete recrystallization occurs between passes. A schematic time-temperature diagram of rolling with recrystallization control is shown in Fig. 2.

Conventional controlled rolling is shown in Fig. 3. It is carried out above the phase transformation temperature, A_{r3}, and below T_{nr}, so that deformation accumulates between the passages and austenite grains are strongly deformed, *i.e.*, austenite pancake formation occurs. Strongly deformed austenite grains provide numerous nucleation sites during subsequent transformation.

Rolling with dynamic recrystallization control, shown in Fig. 4, means that dynamic recrystallization, DRX, is triggered in one or more passes during rolling, which in turn causes rapid metadynamic recrystallization, MDRX. DRX is caused either by the application of single large deformations or by the accumulation of deformation from passage to passage.

Dynamic recrystallization is caused either by applying large deformation forces or by accumulating deformation from passage to passage. Dynamic recrystallization control means that dynamic recrystallization is triggered in one or more passes during rolling, which in turn causes rapid metadynamic recrystallization. The authors of Ref. [20] claim that one of the advantages of this approach is the intensive grinding of grain caused by dynamic recrystallization.

In order to be able to use controlled rolling in practice and obtain the final mechanical properties of the product, it is necessary to know the evolution of the austenite microstructure during rolling. Therefore, models describing the microstructure evolution are needed to develop optimized controlled rolling schedules corresponding to each case.

In foreign practice, thicker slabs are used to obtain a low temperature at the end of rolling so that, spending more time on rolling, the temperature can be lowered to the set temperature. The low temperature of heating and the beginning of rolling also provide a reduced temperature at the end of rolling. However, the dispersion hardening effect decreases at a low heating temperature since the amount of dissolved niobium carbide decreases.

The Japanese company Kawasaki Steel Corporation has mastered the controlled rolling of sheets with a thickness of 19.5 and 20 mm of man-

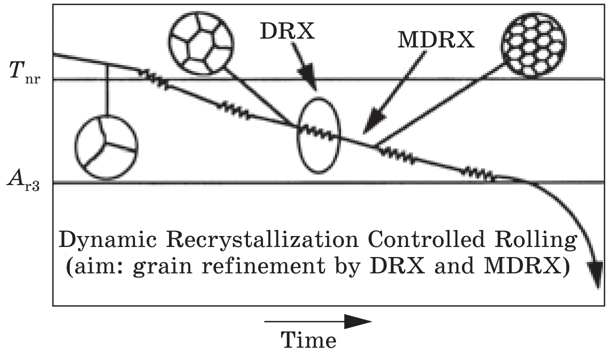


Fig. 4. Schematic demonstrating the very fine recrystallized microstructure after dynamic (DRX) and metadynamic recrystallization (MDRX)

ganese steels for main pipelines (0.08% C) and shipbuilding (0.15% C). Steel for pipes additionally contains 0.04% Nb and 0.07% V; the heating temperature of the slabs before rolling is 1150 °C. A fine-grained ferrite–pearlite structure characterizes the sheets. The steel structure for pipes contains significantly less perlite, and its yield strength is 535 N/mm. The yield strength of shipbuilding steel is 390 N/mm. Both steel plates have high ductility and viscosity characteristics and are designed for use in harsh climatic conditions. It should be noted that Russia has mastered the technology of deformation–thermal hardening, which provides a yield strength of 400–450 N/mm in thick-sheet steel of type 09G2S.

Republic Steel in the U.S.A. produces carbonitride-hardened steel in large volumes, which is supplied after controlled rolling. Its strength is not inferior to more alloyed steels subjected to double heat treatment — quenching and tempering. Steel contains up to 0.22% C, 1.4% Mn, 0.25% Si, 0.14% V, 0.015% Nb, and 0.015% N. It is additionally deoxidized with aluminium based on its residual content of 0,03%, and measures are taken to reduce the sulphur content to 0.012% and phosphorus (up to 0.010%). The temperature of the end of rolling sheets up to 20 mm thick is 760 °C. The yield strength of 620 N/mm characterizes the steel, a temporary resistance of 790 N/mm, a relative elongation of more than 22%, and an impact strength at negative temperatures of more than 100 and 65 J/cm on samples with a sharp incision along and across the rolling direction, respectively. Sheet metal is used in the manufacture of lifting and transport, mining equipment, and other areas.

The development of the theoretical foundations of controlled rolling and the results of studies of thick sheets of industrial steel produced in this way allowed us to establish that carbon content is not a determining factor for increasing the strength of sheets of such steel. The results of these studies were the basis for developing a new class of steels — the so-called low-perlite (up to 0.12% C) and non-perlite (up to 0.05% C) steels. In order to maintain the strength characteristics at a sufficiently high level, more stringent par ameters of controlled rolling are used in the production of rolled products from such steels. In several countries (U.S.A., Japan, Germany, *etc*.), up to 0.5% Mo is introduced into low-perlite and non-perlite steels. Molybdenum delays phase transformations, shifting them to the right towards longer exposures and increasing the amount of bainite or needle ferrite in the structure.

In Japan, the influence of the temperature at the beginning of rolling (820–900 °C), the degree of deformation during controlled rolling, and the cooling rate on the strength and viscosity characteristics of low-perlite steel with different niobium and molybdenum contents, the degree of deformation during controlled rolling and the cooling rate; the heating temperature of the workpieces for rolling was 1200 °C. After compression by 50%, some of the samples were cooled in the air (0.7 °C/s) to a temperature of 820–900 °C. After finishing rolling, the other part of the workpieces was cooled in water at a speed of 60 °C/s. Further, the rolling of sheets up to 15 mm thick was carried out with compression of 29, 50, and 73%. After rolling, some of the sheets were cooled in water to a temperature of 400–600 °C, then in the air.

The studies showed that changes in the carbon and niobium content practically did not affect the strength and viscosity characteristics. An increase in the molybdenum content increased strength and reduced the critical temperature of brittleness. Low-perlite and non-perlite steels have a low-carbon equivalent (CEV) and a low tendency to crack formation and are characterized by high mechanical properties and excellent weldability.

It was shown in Ref. [21] that controlled rolling of steel 09G2 with termination in the two-phase region at 750–700 °C allows, along with an increase in strength properties, to increase the impact strength by 300–400 kJ/m^2 at low temperatures and also to lower the critical temperature of brittleness by about 40 °C compared with the end of rolling in the lower part of the areas. In the case of the end of rolling at 700 °C, the following values of mechanical properties determined during the tensile test were obtained: yield strength 450 MPa, temporary tear resistance 555 MPa, and elongation of about 27%. A further decrease in the deformation temperature increases strength characteristics with a noticeable decrease in ductility and toughness at room temperature. Increasing the mechanical characteristics of the author of the work is associated with the grinding of ferritic grains and creating a stable dislocation substructure in ferrite.

Thus, the available information shows the possibility of increasing the mechanical properties of rolled products made of low-alloy steel. As a result, it becomes possible to replace complex alloy steels. Thus, in Ref. [22], it was shown that steel 09G2, in the case of its rolling at low temperatures, acquires a set of properties similar to those obtained for microalloyed steel 08G2MFB. However, reducing the rolling temperature to intercritical or subcritical values creates difficulties in implementing the process on existing rolling equipment.

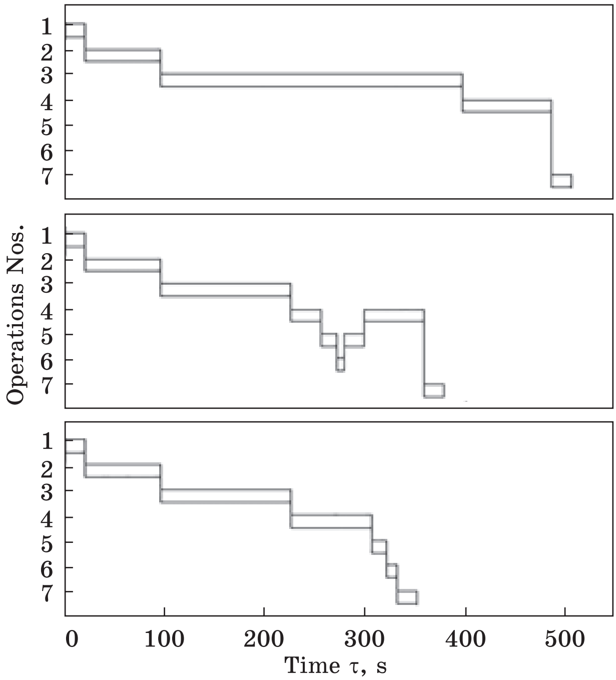


Fig. 5. Graphic representation of the classification of modern types of steel for cold forming [25]: *1* — slab transportation from the methodical furnaces to the roughing stand; *2* — rolling in the roughing stand; *3* — cooling of rolls in the air before rolling in the finishing stand; *4* — rolling in the finishing stand; *5* — transportation of slabs to the HRSG or back to the finishing stand; *6* — accelerated cooling of rolls in the HRSG; *7* — air cooling of the rolled products after the control process (CP)

With the existing low-temperature technology of controlled rolling sheets of low-perlite steels for main pipelines, the rolls are cooled to a set temperature, as a rule, in air, and the number of passes is optimized based on the time of pauses (exposures) according to the developed temperature models.

Such a process has been implemented on modern high-power thick-sheet mills 5000 by Mannesman-rohren (Germany), 4200 by Voest-Alpine (Austria), 4500 by Posko (South Korea), and at the mill 3000 Illich Steel & Iron Works (Mariupol, Ukraine), includes alternating processes of metal deformation and cooling to a predetermined temperature, *i.e.*, controlled rolling with one exposure of peals for air cooling. The controlled rolling process is possible with two cooling exposures or repeated recrystallization heating of the lining [23, 24].

The existing technology of controlled rolling on the reverse mill 3600 of PJSC 'Azovstal Iron & Steel Works' (currently destroyed by russian invasion in Mariupol, Ukraine) includes heating of slabs up to 1150 °C, the temperature of the end of rolling in the finishing stand at 710–740 °C (depending on the carbon equivalent) with a thickness of 50 mm after the roughing stand and compression in the finishing stand of 15–18% in each pass and the last 3 with 10% to ensure flatness.

On the one hand, such a rolling mode causes increased loads on the crate, and the drive of the rolls increases energy consumption and reduces the mill's productivity. On the other hand, its implementation is limited by the ductility of the metal, the strength of the rolling mill parts, and the drive power.

Let us consider possible controlled rolling schemes, including alternative ones, with accelerated cooling, in the flow of the mill 3600 of PJSC 'Azovstal Iron & Steel Works' (Fig. 5).

In the conditions of mill 3600, the standard, controlled rolling process (Fig. 5, *a*) includes holding the rolls in the air after rough rolling with reverse rocking on an intermediate roller. With the proven technology of low-temperature controlled rolling of sheets with a thickness of 16–20 mm, the duration of cooling (holding) of the roll with a thickness of about 50 mm until the set rolling temperature is reached in the area between the stands is about 300 s, which significantly exceeds the usual rate of hot rolling [25].

The duration of exposure of the lining in the air mainly depends on its thickness, and the decrease in productivity with the established low-temperature technology reaches 25–33% [23–25]. The increase in productivity by reducing the rolls' exposure when rolled rolling is cooled between the stands is directly related to the use of high-temperature CP and accelerated cooling of the rolls (Fig. 5, *b*) or finished sheets after rolling (Fig. 5, *c*).

3. Phase and Structural Transitions Occurring during Controlled Rolling

The primary condition under which the controlled rolling effect manifests itself as a method of grinding austenite grains by recrystallization is the need to complete the deformation in the finishing stand of the mill at a temperature above A_{r3}, and part of the total deformation should occur in the lower austenite region and the intercritical temperature range.

However, with compressions in the intercritical temperature range, although strength characteristics increase, ductility and viscosity indicators decrease, which can be partially compensated by low carbon and sulphur content (up to 0.005%), vacuuming, and other measures to improve the quality of continuously cast metal. Compliance with the required low-temperature controlled rolling mode to obtain a given level of properties causes specific difficulties in production practice, primarily associated with ensuring given mill productivity and trouble-free operation.

Difficulties arise because the slab can be rolled at high speed in the roughing crate and then be in the queue, cooling down to the initial temperature of the beginning of rolling in the finishing crate (750–700 °C), which must work continuously. With the existing low-temperature technology of controlled rolling sheets of low-perlite steels for main pipelines, the rolls are cooled to a given initial temperature, as a rule, in air, and the number of passes is optimized based on the time of pauses (exposures) according to the developed temperature model.

The fine-grained structure allows the use of high-strength structural steel under harsh operating conditions. Grain grinding can be obtained by controlling the following parameters during the rolling process: time, temperature, and degree of deformation. The most effective way to grind grain at an acceptable cost is the thermo-mechanical control process (TMCP), which combines thermomechanical rolling, *i.e.*, deformation without recrystallization, and accelerated cooling. The properties obtained by these processes cannot be achieved or repeated by heat treatment alone.

Thus, the analysis of structure formation processes during hot deformation allows us to establish conditions for the practical effect of deformation on the structure of steel with the reproducible achievement of a given structural state of austenite or its decay products. In other words, the technological scheme of controlled rolling assumes not only the control of technological parameters (the set temperature of the end of rolling, the degree of deformation, *etc.*) but also the control of the processes of structure formation by the correct choice of deformation parameters, cooling and chemical composition of steel [26].

The main stages of the controlled rolling technological process: (1) heating of the metal for rolling (austenitization); (2) rough rolling stage (maximum grinding of austenitic grain due to the recrystallization process after each pass); (3) cooling (ensuring the required temperature of the beginning of finishing rolling); (4) finishing rolling stage (bending austenitic/ferritic grain); (5) post-deformation cooling of rolled products (achieving hardening of steel without additional alloying).

All over the world, most small metallurgical enterprises still operate based on traditional hot rolling. As already noted, it is impossible to control microstructural changes during rolling during this process, which leads to the deterioration of mechanical properties [27, 28]. In the most advanced control rolling technology, the temperature during rolling is continuously monitored so that the grain size of austenite is refined, and rolling is interrupted with exposure to allow the sheet to reach a lower temperature during the rolling process. In addition, phase transformations and their consequences, such as precipitation behaviour, are optimized during slab cooling [29]. Refinement of the austenite grain size increases the proportion of the boundary area, which, in turn, increases the number of nucleation sites of ferrite grains during transformation. Consequently, the mechanical behaviour is improved due to the beneficial effect of smaller ferritic grains [30, 31]. A typical schematic representation of the various processes occurring during the controlled rolling of C–Mn–Nb steels is shown in Fig. 6.

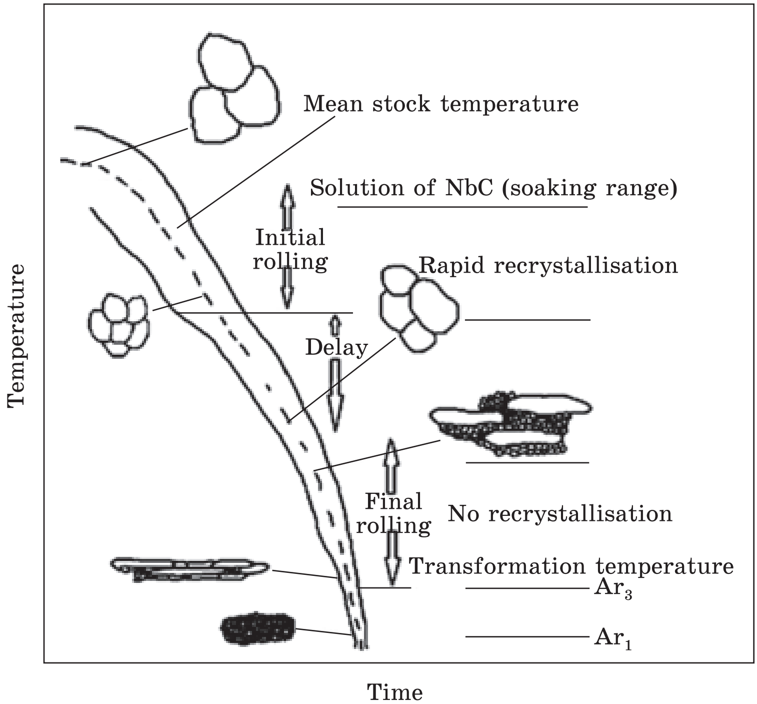


Fig. 6. Schematic representation of various processes occurring during control rolling of C–Mn–Nb steels [34]

Control rolling ends at a temperature slightly higher than A_{r3}, — the temperature, at which austenite begins to turn into ferrite during cooling, — or higher than A_{r1}, — the temperature, at which the transformation of austenite into ferrite or ferrite and cementite is completed during cooling, — but below, the austenite temperature is the recrystallization temperature. The degree of grain grinding depends on the number of dislocations and slip bands (localized sliding of dislocations in a single grain) in non-crystallized austenite [32, 33]. A higher dislocation density and a more significant number of slip bands increase the number of nucleation sites during the austenite transformation into ferrite, leading to a more refined ferrite granular structure [32–34]. More advanced control rolling methods have been developed to control the recrystallization process better.

This method is known as dynamic rolling with recrystallization control. In this process, the overall deformation is higher than in finishing rolling, while the deformation from passage to passage accumulates. As a result, the critical deformation required to cause dynamic recrystallization is exceeded, and consequently, a tiny grain is obtained. The cooling rate during the control rolling process usually depends on the desired microstructure. Air-cooling is usually used after rolling to produce ferrite–pearlite structures. However, accelerated rolling is used to refine the ferrite grain size and enrich the rest of the matrix with non-equilibrium phases of martensite and/or bainite [32–34].

In Ref. [35], connections were established between the structure parameters and properties on a wide range of steels: carbon (0,05-0.23% C), low-alloyed and alloyed (0.03–0.11% C) of the G2, HG2H, HG2NM, H2G2H types, including microalloyed Nb, V, Ti, as well as model materials with ferritic, bainitic and martensitic structures. It was found that in the optimal state of austenite before $\gamma \to \alpha$ transformation (grain size ≈20 microns, subsequent deformation $\varepsilon = 70\%$), steel with all the studied types of matrix structure: ferrite (≈15% perlite), ferrite with a perfect subgrain structure, needle ferrite, upper bainite; lower bainite and tempered batch martensite are described (within the range of the spread) by a shared dependency. Thus, if all other things being equal, the resistance to viscous fracture in the first approximation depends only on the strength level of the steel. In the case of conventional hot rolling ($T_{cr} = 950$ °C), the structural states of resistance to brittle fracture are sharply different.

Thus, the possible type of structures is expanded (from ferrite-pearlite to intermediate-type structures and martensitic), providing increased strength after controlled rolling in combination with the necessary level of fracture resistance. Conditional upper limits on the yield strength under the conditions under consideration are: for ferrite (≈20% perlite) — 550 N/mm^2; ferrite with a subgrain structure — 650 N/mm^2; needle ferrite — 570 N/mm^2; bainite — 750–850 N/mm^2; and batch martensite — 900 N/mm^2.

The deformation of austenite at temperatures below the track makes it possible to influence the final structure of the steel to a greater extent than in the case of the $\gamma \to \alpha$ transformation from recrystallized grain. The elongated grain shape in deformed austenite (increased specific surface area of the boundaries), deformation bands forming inside them, twin boundaries, cellular dislocation structure, and other defects may be the sites of the origin of a new phase increase the specific effective surface of austenite.

After austenite deformation below the track, martensite packages are oriented at an angle of 90 or 30° relative to the rolling plane (the boundaries of elongated grains). The boundaries of austenitic grains hinder the growth of martensitic crystals. Therefore, a decrease in the latter size in the direction of the thickness of the rolled products, respectively, reduces the size of the packages and the length of the martensite rails (by 2–5 times at $\varepsilon = 70\%$). In this case, martensitic crystals inherit the substructure of deformed austenitic grains, *i.e.*, the crystals are divided by sub-boundaries.

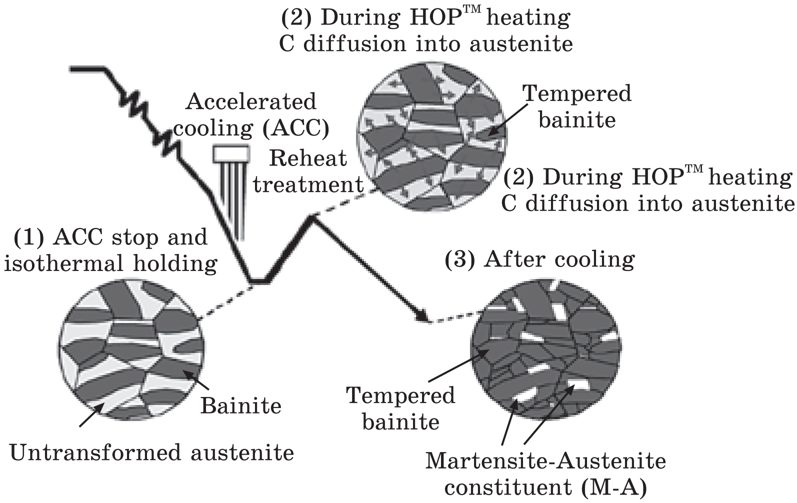


Fig. 7. Microstructure control by induction heating technology [36] (HOP indicates the heat treatment on-line process)

The small-angle boundaries between the upper bainite rails do not prevent the propagation of cracks (they are delayed only by the boundaries of the bainite package or the initial austenitic grain). Therefore, the viscosity of the bainite steel can be increased only by affecting the structure of the initial austenite. The limits of grinding of austenitic grain, achieved by heat treatment or deformation above the track, do not effectively affect the structure of bainite and the complex properties of steel. The mechanism of influence of austenite deformation on the structure of bainite is similar to that described above for martensite: a decrease in the length of the bainite rail and the size of the package with an increase in the degree of compression below the track (a decrease in the size of austenite grains), as well as inheritance of the subgrain boundaries of deformed austenite.

In contrast to ferrite-pearlite steels, steels with martensitic and bainite structures have a high density of cleavages in the fracture and after rolling in the γ region, which may be due to the inheritance of the texture of deformed austenite (the shear nature of the transformation and the orientation of martensite and bainite crystals) [35].

Thus, in Ref. [36], JFE Steel company produced YP460 steel with a low yield strength coefficient of 780 MPa. Continuation with the tempering process was also achieved by JFE Steel's HOP (heat treatment on-line process), realizing online production of high strength plates. Excellent mechanical properties of the base metal and excellent weldability and viscosity of the weld are achieved due to the formation of a multiphase microstructure, which consists of an essential bainite phase

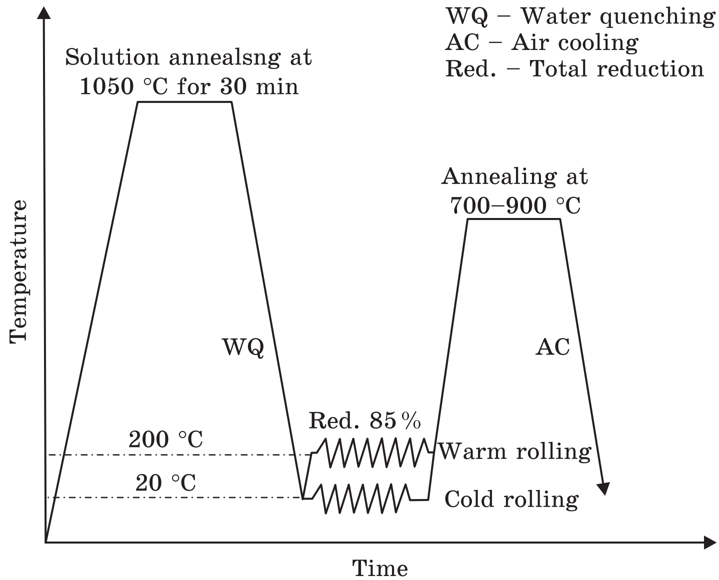


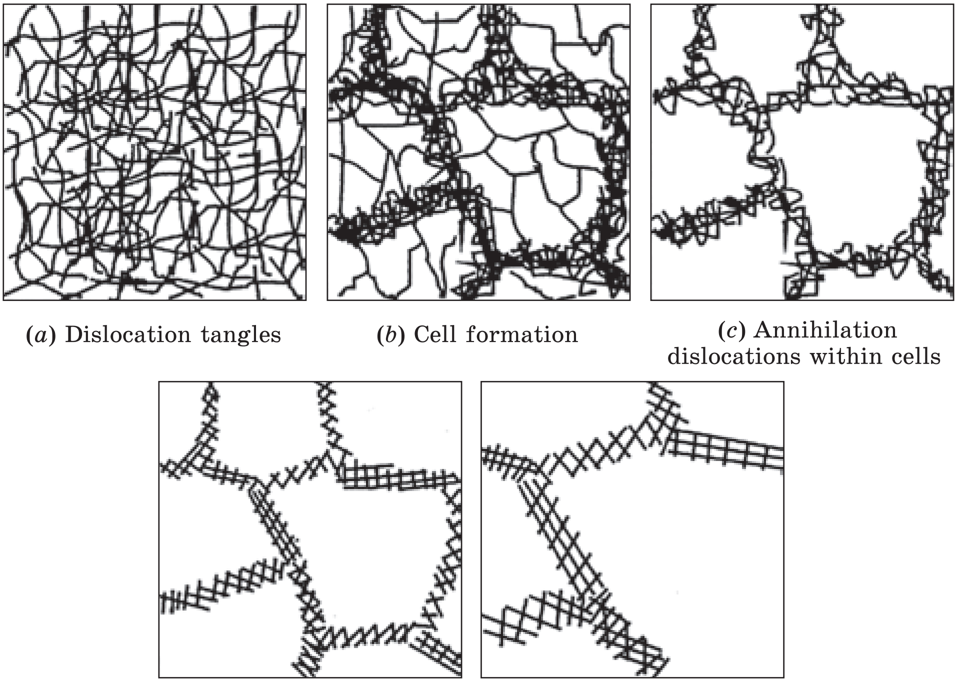
Fig. 8. Scheme of advanced thermomechanical processing [37]

and a thin $M–A$ (martensite–austenite phase), by applying the processing mode shown in Fig. 7.

In Ref. [37], 304-grade steel was rolled at various temperatures — room temperature (cold rolling) and 200 °C (warm rolling) to elucidate various deformation mechanisms according to the scheme shown in Fig. 8. Then, the influence of deformed microstructures on the subsequent microstructural evolution and texture development during annealing was studied. Microstructural studies have shown that the microstructure obtained after rolling at room temperature has a two-phase structure consisting of ≈60% α'-martensite and residual austenite, while the structure after warm rolling is completely deformed austenite.

Most of the applied energy during controlled rolling is converted into heat, and only a small part remains in the form of energy (≈1%). The increase in such energy occurs mainly due to an increase in the density of dislocations and an increase in the area of grain boundaries. The energy accumulated from the material is the driving force for the processes of return and recrystallization.

The return is made when the metal is heated below the recrystallization temperature by removing (the ‘rest’ stage) and subsequent redistribution in order to reduce the concentration (the ‘polygonization’ stage) of defects in their crystal structure, primarily the so-called dislocations. During rest, the concentration of point defects decreases; after this one, they annihilate and move to the boundaries of dislocations; the latter are redistributed by sliding in their planes without forming new bounda-



(*a*) Dislocation tangles (*b*) Cell formation (*c*) Annihilation dislocations within cells

(*d*) Subgrain formation (*e*) Subgrain growth

Fig. 9. The stages of recovery with the first cells formed due to the rearrangement of dislocations; then, subgrains are formed and grow due to annihilation of dislocations [38]

ries. During polygonization, dislocations are redistributed by diffusion and sliding, accompanied by partial annihilation. As a result of this stage, so-called ‘polygons’ are formed — areas inside the crystallites that are separated by small-angle dislocation boundaries and therefore do not contain dislocations. Polygonization may be the initial recrystallization stage, if we discuss heating after large deformations. In this case, it is essential to reduce the density of dislocations in the material to remove the effect of cold deformation altogether.

During the return, the dislocations are rearranged into lower energy configurations. The return is a series of events: the formation of cells, the annihilation of dislocations inside cells, the formation of small-angle subgrains, and the growth of subgrains [38] (see Fig. 9).

In the area of high-temperature deformations, the austenitic grain is crushed due to dynamic recrystallization co-occurring with plastic deformation. If complete softening of the metal does not occur during intraformational pauses and cooling, metadynamic and static recrystallization may occur. Metadynamic recrystallization occurs immediately after deformation without an incubation period — the growth of embryos formed due to dynamic recrystallization. Static recrystallization is the generation of new grains at the boundaries of deformed former grains after the completion of plastic deformation (during intraformational pauses) in an incubation period.

Static recrystallization can be primary, collective, and secondary. During primary static recrystallization, the nucleation and growth of new grains occur. During the collective, 'normal' growth occurs. During secondary recrystallization, the uneven growth of primary recrystallized grains occurs due to neighbouring grains, which decreases the metal structure uniformity. The optimal use of static recrystallization to obtain a homogeneous fine-grained austenite structure assumes the full completion of primary recrystallization in each pause but excludes excessively long exposures leading to collective recrystallization. By varying the heating temperature, the beginning, and end of rolling, the holding time, the degree and speed of deformation, it is possible to obtain a set of different structures — from dynamically polygonised and dynamically recrystallized to statically recrystallized. Austenitic grains remain deformed and uncrystallised with a decrease in the temperature of hot deformation. With subsequent accelerated cooling in the process of phase transformation, inheritance in one form or another of the structure of deformed austenite is observed by the resulting final structural components (martensite, bainite, *etc.*), which contributes to an increase in the dispersion of the structure with a high level of strength and viscoplastic properties [12, 39].

The finishing stage of rolling is carried out to increase the density of defects in the crystal structure of the metal and their ordered distribution (substructure), which leads to an increase in the number of nucleation centres of a new phase. In addition to the boundaries of deformed austenitic grains and subgrains, the places of formation of the α-phase are deformation bands and twin boundaries, the number of which increases with increasing degree of deformation in the temperature region where there is no recrystallization. The essential condition for rolling in the medium-temperature region is the absence of recrystallization, which could eliminate a significant part of the defects in the crystal structure induced by plastic deformation. Compressions carried out directly near the A_{r3} point, when recrystallization is wholly inhibited, are essential to obtain a fine-grained structure. Cooling steel after deformation completion during thermomechanical processing is accelerated with the use of controlled cooling systems.

The peculiarity of the $\gamma \rightarrow \alpha$ transformation during continuous cooling in low-carbon low-alloyed steels is the low stability of austenite, which significantly depends on minor fluctuations in the chemical composition [40], and hot plastic deformation during thermomechanical processing initiates transformations proceeding by a diffusion mechanism. The

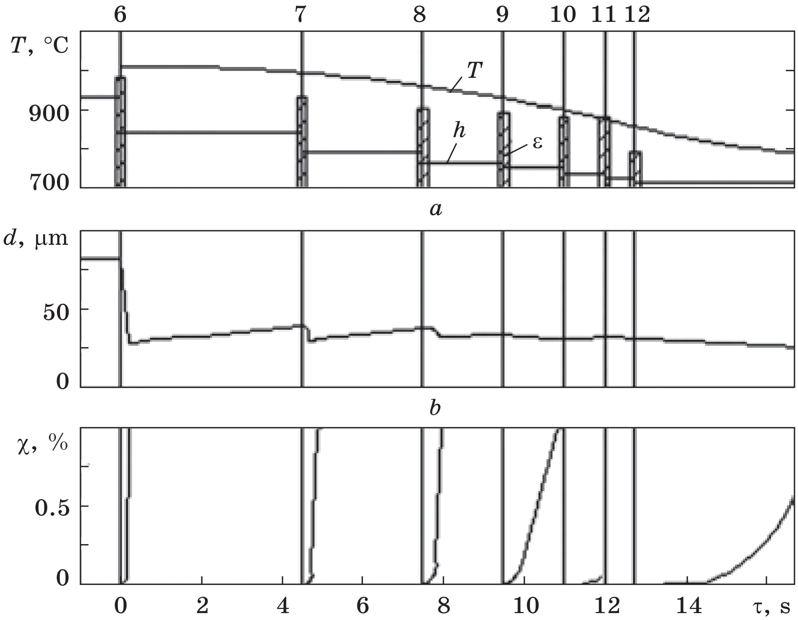


Fig. 10. Temperature and deformation mode of rolling (*a*), change in (*b*) and kinetics of austenite recrystallization in the finishing group of stands (*c*); *T* — rolling temperature; *h* — rolling thickness; ε — relative compression [42]

structure formation of steel during hot plastic processing depends on such processing parameters as the degree, speed, temperature of deformation, and composition, which determine the ratio between hardening and softening during deformation. The condition of austenite before transformation depends on the conditions of roughing and finishing rolling. In the case of rolling at the rough stage at temperatures above the beginning of a dynamic and subsequent static recrystallization, a recrystallized austenite structure is formed. At the finishing stage of rolling at low temperatures, an oriented austenite structure with a relatively low degree of grain misalignment is formed. In the case of rolling at low temperatures, both at the roughing and finishing stages, austenite acquires a crystallographic and morphological texture of deformation, which leads to forming a final banded structure [41].

The deformation of the strip in the crates of the finishing group is accompanied by a decrease in the temperature of the metal, a reduction in the duration of intraformational pauses, as well as a decrease in the relative compression in a separate crate (Fig. 10, *a*). Under these conditions, after deformation in the first four crates (crates 6–9), the primary recrystallization of austenite has time to complete before the strip enters the next crate, after deformation in crate 10, recrystallization takes place partially, and after deformation in crate 11 does not develop

(Fig. 10, *c*). As a result, the accumulated total deformation at the exit from cage 12 is 44%. Subsequent recrystallization causes the retention of a residual hardening equal to 14% by the time the supercooled austenite begins to decay. After grinding the austenite grain to ≈25 μm in the first two cycles of deformation–recrystallization, its size remains almost unchanged (Fig. 10, *b*) [42].

After ferrite formation is completed at 680 °C, the pearlite transformation begins, which ends during the winding of the strip into a roll.

In modern rolling mills, a homogeneous fine-grained structure over the entire thickness of the rolled product can be achieved only in the area near the surface of the sheet. When moving closer to the middle of the rolled sheet, uniformity deteriorates significantly. Thus, in pipe steels in the middle of rolled products, there is usually a significant orientation of structural components and texture due to a high degree of deformation. The formation of such a structure in strip steels is allowed; in shipbuilding steels, this is impossible due to the magnitude of the relative narrowing in the thickness direction and the requirements for the type of fracture. Thus, achieving high strength and plastic characteristics in pipe steels is ensured by forming fine-grained austenite due to dynamic, metadynamic and static recrystallization. In shipbuilding steels, the content of chemical elements responsible for the position of critical points is higher, which eventually leads to a decrease or absence of the proportion of austenite subject to complete dynamic recrystallization. Only initial recrystallization processes, such as polygonization and return, are possible, leading to partial softening of the metal due to transverse sliding dislocations.

The choice of the cooling start temperature and the cooling trajectory plays a vital role in the formation of the microstructure and properties of steel. With rapid cooling of steel and subsequent exposure to air (under conditions close to isothermal), the restructuring of the crystal lattice of f.c.c.→b.c.c. (shear transformation) and diffusion redistribution of carbon are realized, accompanied by the decay of austenite with the release of ferrite and cementite [43–45]. To manage effectively the structure formation processes, it is necessary to understand the mechanisms that determine the development of phase transformations in iron and steel.

The model approaches proposed to date correctly cover the main features of polymorphic transformations in steel [44]. However, they leave questions concerning the conditions for implementing transformations and their driving forces in the shade. In Refs. [46, 47], a microscopic approach was proposed to construct the theory of polymorphic transformations in iron and steel based on the results of first-principle calculations, taking into account the contribution of lattice and magnetic degrees of freedom (as well as defects [48–50]), as well as the diffusion redistribution of carbon and alloying elements in steel. The concepts developed in

Ref. [44, 47] can serve as a reliable basis for developing mathematical models of the formation of the phase and structural state in steel.

In accelerated cooling, the nucleation and subsequent growth of the α-phase occur at the grain boundaries, resulting in a bainite (at a high cooling rate) or bainite–ferritic (at a moderate cooling rate) structure being formed. In this case, the microalloying of Mo and B, the grain size, and the deformation accumulated at the finishing rolling stage can play an essential role, especially if the rolling ends at a temperature lower than the beginning of the ferritic transformation of A_{r3}. In the latter case, the ferritic transformation accelerates [51, 52], which can lead to the grinding of the forming structure. The kinetics of transformations under conditions close to isothermal has been well studied. The curves of ferritic, perlite, and bainite transformations on the isothermal decay diagram of supercooled austenite have a C-shaped appearance and usually overlap in microalloyed steel [53].

According to popular opinion [54, 55], the alloying with Mo and, especially, B can significantly slow down the kinetics of ferrite–perlite transformation, shifting its beginning to the region of long times. At the same time, these elements practically do not affect the bainitic transformation, the development of which is mainly determined by the growth stage of ferritic plates in the grain volume. Therefore, the alloying with Mo and B is considered an effective way to control the kinetics of transformation and microstructure formation in high-strength low-alloy (HSLA) steel.

However, recent studies [56, 57] on the kinetics of austenite decay in low-carbon steel have led to an unexpected conclusion. As was shown in Ref. [56], the formation of ferrite accelerates at elevated temperatures (650–700 °C) in steel containing B and Mo, while at lower temperatures (550–600 °C), a slowdown in ferritic transformation is observed. The results of [56] indicate that in the latter case, an increase in the barrier of formation of α-phase nuclei at the boundaries of grains containing B or Mo segregation plays a decisive role. At the same time, these elements increase the driving force of the growth of the α-phase (which correlates with the change in temperature T_0 during doping [58]), accelerating the transformation in the region of elevated temperatures.

4. Conclusions

Structural–phase transformations have always been closely related to deformations and solid strength theory. The transformation of materials' structure and phase composition during heat treatment automatically assumes various atomic displacements associated with changes in the type of symmetry of crystal lattices, the generation of defects, internal stress fields, *etc*. There are cases when macroscopic plastic deformation of products and even their destruction were observed during structural–phase transformations. On the other hand, external mechanical influences are also capable of causing numerous phase transitions, leading to changes in the phase and chemical composition of a substance at a fixed temperature.

The modern technology of steel production is a complex, knowledge-intensive process that should provide control of structure formation at all stages of TMT, combining controlled rolling and accelerated cooling. The strength and viscoplastic characteristics necessary for thick-rolled products are achieved due to forming a ferrite-bainite structure with a small grain size and increased dislocation density.

Currently, the main processes controlling the structure formation at various stages of TMT have been established. They include the grinding of austenitic and ferritic grains through controlled rolling, the formation of carbonitride phase emissions during hot rolling, and the decomposition of a significant part of austenite in the temperature range of bainite transformation. However, several tasks remain, the solution of which is essential for further improving the efficiency of the TMT process and achieving the required structural state. To solve these problems, further experimental and theoretical studies are required with the extensive use of computer modelling methods of phase and structural transformations.

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ВПЛИВ КОНТРОЛЬОВАНОГО ВАЛЬЦЮВАННЯ НА СТРУКТУРНО-ФАЗОВІ ПЕРЕТВОРЕННЯ

Розвиток чорної металурґії значною мірою зумовлений вимогами провідних металоспоживчих галузей промисловости щодо підвищення експлуатаційних властивостей конструкційних криць з метою збільшення допустимих навантажень, пониження металомісткости та підвищення надійности машин, конструкцій, маґістральних газопроводів тощо. За значних обсягів виробництва металопрокату на передній план висувається також завдання пониження енергоспоживання та витрат сировинних ресурсів під час його виробництва. Важливими й актуальними у зв'язку з цим є роботи, спрямовані на створення металознавчих основ і розроблення нових технологій, що дають змогу виготовляти продукцію з потрібним поєднанням властивостей безпосередньо в потоці вальцівного стану (за виключенням подальшого термічного оброблення) шляхом цілеспрямованого керування процесами структуроутворення, а також розширення сфер ефективного застосування таких технологічних схем (щодо марочного та розмірного асортименту, амортизаційного призначення тощо). Такий підхід дає змогу підвищити конкурентоспроможність металопродукції, яка визначається досяжним поєднанням характеристик металу із пониженням витрат на його виробництво.

Ключові слова: пластична деформація, контрольоване вальцювання, структура, фаза, перетворення.