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PROBLEMS ASSOCIATED WITH CONTROLLING THE TEXTURE OF METALLIC MATERIALS

PROBLEMI DOLOČEVANJA TEKSTURE KOVINSKIH MATERIALOV

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In this paper we consider some of the problems associated with the texture control of metallic materials during such processes as recovery, primary, collecting (normal grain growth) and secondary recrystallization. The study focuses on the most extensively studied electrical steels and takes account of well-known and assumed influencing mechanisms involving chemical elements and aluminium nitride particles dissolved in solid solution during the formation of the steel's microstructure and texture. We concluded that the two main, influential mechanisms are as follows: first, inhibiting of the recovery and recrystallization processes; second, the energy changes associated with the grains of different orientation. It is assumed that the energy fields being created by the nitride particles are significant when it comes to slowing down growth and selecting the direction of the grain-boundary mobility. Two hypotheses were advanced relating to the asymmetry of the grain boundaries' properties and the availability of grains with different energies, which make it possible to explain the secondary-recrystallization process.

Key words: recrystallization, structure, texture, grain, nitride

Preučili smo nekatere probleme določevanja teksture kovinskih materialov pri procesih poprave, primarne rekristalizacije, rasti zrn in sekundarne rekristalizacije. Raziskava je bila vodena na najbolj raziskanih orientiranih elektropločevinah. Študirali smo dobro poznana mehanizma vpliva kemijskih elementov in aluminijevih nitridov, raztopljenih v trdni raztopini, na oblikovanje strukture jekla in teksture. Ugotovljena sta bila dva glavna mehanizma vpliva, in sicer: prvi v zadržani popravi in rekristalizaciji in drugi v spremembi energije različno orientiranih zrn. Predpostavljamo, da nastanejo energijska polja zaradi nitridnih delcev, ki zavirajo rast zrn in vplivajo na mobilnost mej. Predloženi sta dve hipotezi asimetrija mej zrn in razpoložljivost zrn z različno energijo, s katerima lahko razložimo proces sekundarne rekristalizacije.

Ključne besede: rekristalizacija, struktura, tekstura, zrna, nitridi

1 INTRODUCTION

The significance of a metallic material's texture is becoming increasingly important. The role of texture was first discovered in electrical steels, this was then followed by the effect of texture in deep-drawing steels, other structural steels, thin films, etc. Texture control is a mostly well-mastered technique in electrical grain-oriented steel (transformer steel) and deep-drawing steels. Transformer steel is made using a secondary-recrystallization process and deep-drawing steel, using a primary-recrystallization process. In addition to these two processes there are also the processes of recovery and normal grain growth.

During heating, two basic schemes for these processes are possible in different metallic materials after cold deformation:

A. Recovery → primary recrystallization → normal grain growth

B. Recovery → primary recrystallization → normal grain growth → secondary recrystallization.

Depending on requirements, it is possible to carry out only recovery, or, recovery and primary recrystallization. The control of the steel's texture and the microstructure formation during heat treatment is made by altering the conditions relating to temperature and time, and the composition of the atmosphere.

In production, the final microstructure of the metallic material and the texture formation begin with melting and end with the final heat treatment. Take, for example, the mechanism of "microstructure inheritance". This involves an initial material with a fine grain structure that has a finer grain microstructure after the primary recrystallization during the final heat treatment with identical treatment conditions. This mechanism may refer to texture. For example, the most modern process (relatively low-cost steel with good properties) for the production of electrical non-oriented electrical steels (dynamo steel) is a technology that involves melting and continuous casting, hot rolling, normalizing in a pusher-type furnace, cold rolling and final annealing in a pusher-type furnace. At this technology the texture of the finished steel is determined by texture of hot-rolled strips approximately on 80 %.

The texture of metallic materials is formed during continuous casting and solidification, hot and cold rolling, and heat treatment. The general rules and mechanisms of texture formation are well known.

However, this knowledge does not always provide the required result, as there are mechanisms effecting texture formation which are not yet completely understood. The most important, in this respect, are the effects of chemical elements and particles of secondary phases dissolved in solid solution. And these problems are best examined on familiar electrical steels.

2 THE INFLUENCE ON RECOVERY AND PRIMARY RECRYSTALLIZATION

Transformer-steel technology involves melting and continuous casting, hot rolling, initial cold rolling, intermediate decarburizing annealing, secondary cold rolling to achieve the final thickness and the final high-temperature annealing. During the process of heating, in the case of high-temperature annealing, secondary recrystallization with (110)[001] texture formation takes place. Here, aluminium nitrides are used as inhibitors. To get a sharp (110)[001] texture, certain primary-recrystallization microstructure parameters are required, together with the texture, the aluminium nitride particles and certain kinetics of their precipitation, coalescence and dissolution, until the moment of secondary recrystallization in steel. With any type of transformer-steel technology, there occurs an increase (precipitation) and then a reduction (coalescence and dissolution) in the number of aluminium nitride particles in the process of heating during final high-temperature annealing. This behaviour of the aluminium nitrides is very important. The formation of microstructure, texture and aluminium nitrides begins with melting. The modes of every subsequent operation then have an influence on its parameters. In addition, the chemical composition of the steel also has an effect. The component {111}<uvw> is the most sensitive in terms of texture. Studies of the rules of formation for the component {100}<uvw> are interesting from the point of view of its increase in the finished dynamo steel and several structural steels.

Figure 1 illustrates the pole component density {100}<uvw> and the {111}<uvw> change in the process of transformer-steel production, which was described above. It should be noted that high-temperature annealing is usually conducted at a temperature of 1150 °C. **Figure 1** presents data on the texture of primary recrystallization (abscissa point of high-temperature annealing) which occurs during the process of heating up to 700 °C in the case of high-temperature annealing (the samples were taken out of the furnace at this tempe-

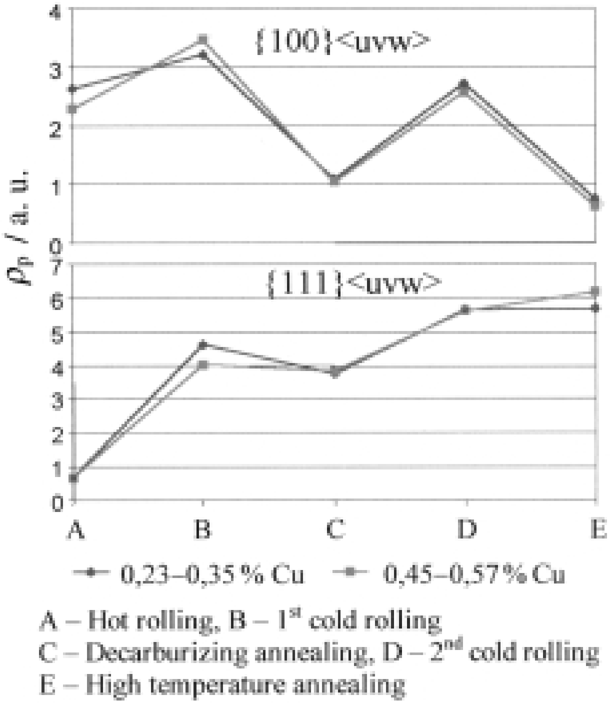

Figure 1: Change of polar density ρ_p of the components {100}<uvw> and {111}<uvw> at manufacture of steel with the different contents of copper

Slika 1: Sprememba gostote polov ρ_p {100}<uvw> in {111}<uvw> pri izdelavi jekla z različno vsebnostjo bakra

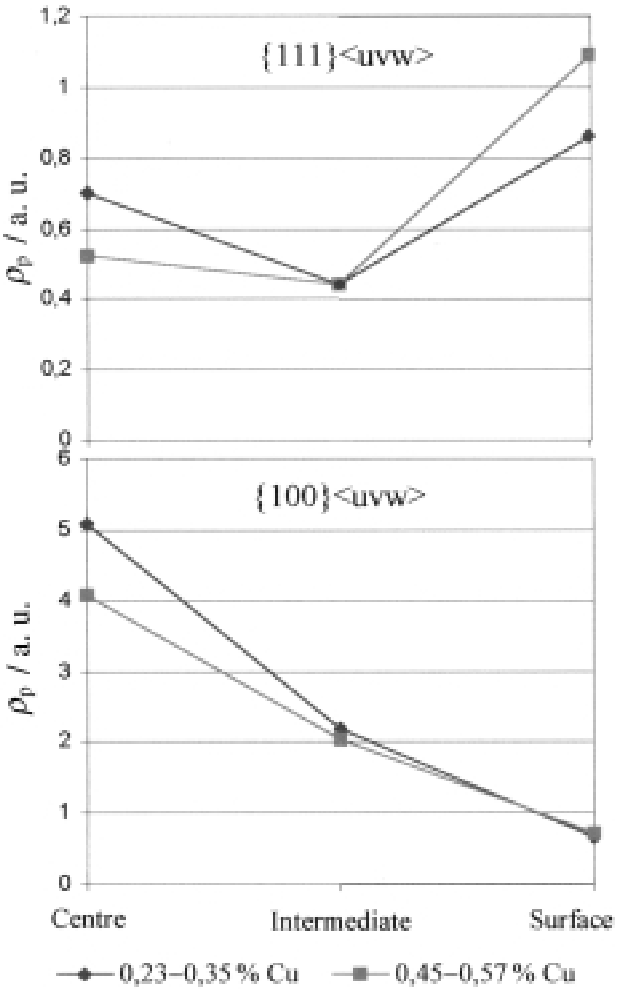

Figure 2: Distribution of polar density ρ_p of the component {111}<uvw> and {100}<uvw> on depth of flat hot-rolled bars

Slika 2: Razporeditev gostote polov ρ_p komponent {111}<uvw> in {100}<uvw> v globini vroče valjanih slabov rature). **Figure 2** gives the components {100}<uvw> and {111}<uvw> distribution along the thickness of the hot-rolled strips. The conditions of the steels' treatment were identical. The chemical composition of one group (12 heats, 100 t each) differed from another steel group (10 heats per 100 t) in terms of copper content: 0.23–0.35 and 0.45–0.57 %, respectively.

Figure 1 proves that the number of {100}<uvw> components in the process of steel production is decreasing, and the number of {111}<uvw> components goes up. The first and second cold-rolling procedures result in the growth of these components number, that is natural for metals with a BCC lattice. The conditions of decarburizing and high-temperature annealing used in industry result in component {100}<avow> reduction. The search for mechanisms of {100}<uvw> component preservation during annealing can become a serious reserve in the influence on texture formation. During decarburizing annealing the {111}<uvw> component quantity is reduced, and during high-temperature annealing it goes up in steel with a higher copper content, but it does not change in steel with a lower copper content.

The described results illustrate the influence of copper on the texture formation in steel. The texture-component distribution in **Figure 2** is characteristic of hot-rolled metal strips with a BCC lattice. This is due to the influence of differently directed energies across the metal strip cross-section during hot rolling. However, the quantity of this or another orientation in different layers of hot-rolled strips may be influenced by hot-rolling modes, by admixtures and by amounts of alloying elements. In this case (**Figure 2**) the results prove that a reduction in the copper content with the applied regime of hot-rolling modes makes it possible to reduce the quantity of {100}<uvw> orientations in the central layer of the hot-rolled steel strips. The mechanism of such an influence has not, however, been discovered yet. Nevertheless, we can assume that the recovery process takes place when the copper content in the steel is modified.

This assumption proves one of the studied mechanisms of the influence of copper. **Figure 1** shows that the {111}<uvw> component quantity increases in steel with a higher copper content (0.45–0.57 %) during primary recrystallization in the process of heating during high-temperature annealing. Meanwhile, the grain size after primary recrystallization is 2–3 micrmeters bigger than steel with a lower copper content. It is well known that the deformation texture can be preserved in the recrystallization texture during the inhibition of recovery processes and primary recrystallization by chemical elements dissolved in solid solution or simultaneous solid solution break down (second-phase particle precipitation). Copper has this particular effect; it inhibits the recovery and primary recrystallization. This inhibiting results in a well-developed steel substructure. Primary recrystallization is developed out of this substructure, but not from the deformed matrix.

In summary, texture and structure control can be achieved with chemical elements and second-phase particles dissolved in solid solution during recovery and primary recrystallization.

3 THE INFLUENCE ON NORMAL GRAIN GROWTH

As knowledge and methods of control improve, the process of grain growth will become more widely used. In transformer-steel production, this process may take place during hot rolling, decarburizing and high-temperature annealing. In dynamo-steel production it may take place during hot rolling, normalization and final annealing. In transformer-steel production the collecting-recrystallization process is very important during high-temperature annealing. If additional precipitation of phase-inhibitor particles takes place, then the polar density of the {111}<uvw> orientation is strengthened in the process of grain growth, and this is a condition to get a sharp {110}<001> texture during secondary recrystallization.

It is considered that the forces that inhibit the mobility of grain boundaries in polycrystalline metallic materials restrain grain growth. Dissolved admixtures and the presence of secondary-phase particles in metals [1,2] condition these forces. Admixture adsorption on grain boundaries reduces their energy and, consequently, their mobility. The separation of the second-phase particles or the contouring of the second-phase particles results in an increase in the boundary area, i.e., its energy. Boundary energy growth is disadvantageous; it causes the appearance of braking forces. Based on data in [2] the grain size (*d*) of a polycrystalline metallic material is dependent upon the size (*r*) and the volumetric content (*f*) of the second-phase particles:

$$d = g \cdot r/f$$

where g is a factor dependent upon particle morphology. The grain size should reduce with an increase of the volumetric content and the dispersion of the particles. It is considered [3,4] that the secondary-recrystallization process is carried out in metals and alloys under the condition of collecting-recrystallization inhibition, with admixtures or second-phase particles. These phases have been called inhibiting phases.

Research into thin structures of transformer and dynamo steels using transmission electron microscopy showed that the precipitation of second-phase particles occurs not on the boundaries but inside the grains [5-9], **Figure 3**. Similar data were obtained by Yanovskoy [10]. In the process of heating during final high-temperature annealing in transformer steels the following regularities are observed, which result in secondary recrystallization and a sharp {110}<001> texture:

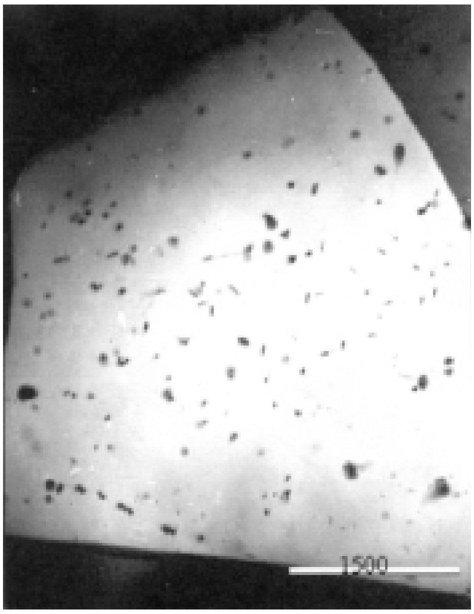

Figure 3: Precipitation of aluminium nitrides in the volume of grains in the process of heating at the high-temperature annealing at 700 °C. State 2.

Slika 3: Izločki aluminijevih nitridov v volumnu zrna med procesom visokotemperaturnega žarjenja pri 700 °C, stanje 2

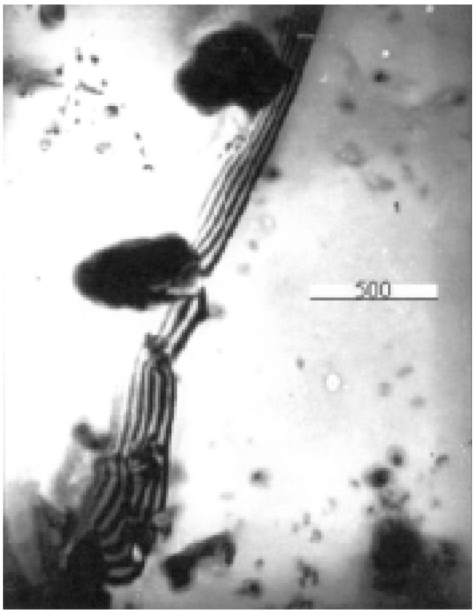

Figure 4: Accumulation by grain boundaries of nitrides in the process of heating at the high-temperature annealing to 900 °C. State 2.

Slika 4: Akumulacija mej zrn nitridov med procesom visokotemperaturnega žarjenja pri 900 °C, stanje 2

- Prior to the start of secondary recrystallization, normal grain growth takes place. In this period the grain size grows by 2–20 micrometers. The initial mean grain size makes up 12–25 micrometers. The growth of the {111}<uvw> grain is dominant;
- The beginning of grain growth during collecting recrystallization coincides with the beginning of nitride precipitation;
- Depending on previous operation regimes, the high-temperature annealing mode and the chemical composition of steel nitride precipitation is characterized by certain kinetics;
- Aluminium nitrides are not precipitated on boundaries but inside the grains. In grains with orientation {111}<uvw> quantity of nitrides is higher than in grains of other orientations, and separate grains {110}<uvw> have the smallest quantities. For example, in grains with {111}<211> orientation nitrides the density can make up 10^{14} cm^{-3}, and in {110}<001> grains it is $5\cdot10^{12}$ cm^{-3}. Grains {110}<001> based on nitrides density are divided into two groups: the first, $(3–6)\cdot10^{12}$; the second, $(1–5)\cdot10^{13}$ cm^{-3}. The receipt of various nitrides quantity in grains of different orientations is connected with the realization of two mechanisms of grain formation during primary recrystallization: 1, the mechanism of grain-boundary migration; 2, the mechanism of subgrain coalescence. The action of these two mechanisms is more significant when primary recrystallization develops from a well-generated substructure. Grains from the first mechanism have fewer crystalline-structure defects than grains of the second mechanism. In steel compositions that are used the solid solution of silicon ferrite contains small quantities of aluminium and nitrogen. The driving force behind nitride formation is small. In these conditions the defects of the crystalline structure become the sites of nitride precipitation. Therefore, grains of the first mechanism precipitate fewer nitrides than the grains of the second mechanism;
- The individual particles of nitrides and their congestion occur on mobile boundaries at the end of normal grain growth before the secondary recrystallization begins, **figure 4**. Their curvature testifies to the boundaries' mobility.

The comparison of the known and given experimental data shows the discrepancy. First, how do nitrides slow down grain growth if they are not at the boundaries, but inside the grains? Second, why do grains grow when the nitride quantity increases?

Explaining the first discrepancy, it is possible to assume that the energy (the stresses) fields created by nitrides inhibit grain-boundary motion. With an identical quantity of nitrides in adjacent grains similar stress fields from both grains influence the boundary between them. Probably, up to certain temperatures, the boundary cannot move to the area with higher energy, i.e., in the direction of the stress source. With the quantity of nitrides the growth fields' energy goes up, thus making the inhibition of boundary movement more effective. Up to certain temperatures, grain growth is impossible, as the driving force of grain growth is less than the braking force of the energy fields.

The second discrepancy is explained in [11]. In this study, two alloys of Fe + 3 % Si contained mass fraction of aluminium 0.018 % and 0.010 (alloy 1) and 0.001 % (alloy 2) of nitrogen. After hot rolling the alloys were subjected to cold rolling with an 88 % deformation (2.50 → 0.30 mm), decarburizing annealing (800 °C, 7 min, mass fraction of carbon content decreased from 0.030 % to 0.004 %) and then high-temperature annealing (1050 °C, heating rate from 400 °C/h – 30 °C/h). Two variants of high-temperature annealing were used. In one variant the annealing was conducted in a neutral atmosphere of argon, in the second an atmosphere of 5 % H_2 + 95 % N_2 + 0.001 % NH_3 was used. The second type of atmosphere provided saturation of alloys with mass fraction of nitrogen of up to 0.010 % during heating. During heating the samples were taken out of a furnace for investigation. Their condition was fixed by fast cooling in water.

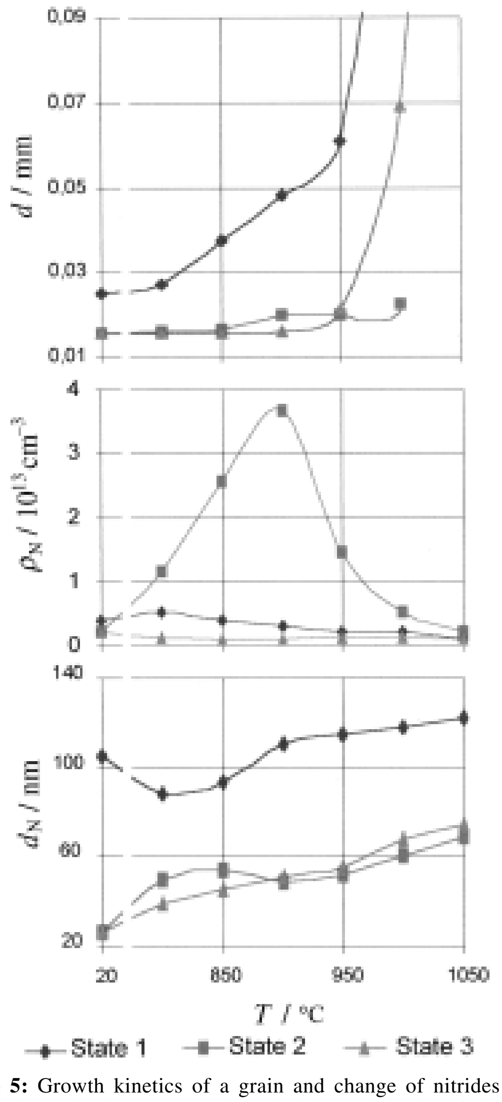

Figure 5: Growth kinetics of a grain and change of nitrides parameters (d – grain size, ρ_N – density of nitrides, d_N – size of nitrides)
Slika 5: Kinetika rasti zrn in sprememba parametrov nitridov (d – velikost zrn, ρ_N – gostota nitridov, d_N – velikost nitridov)

The initial microstructure and the texture state of the alloy 2 samples before high-temperature annealing were identical. Alloy 1 is distinguished by its large grain size and the coarse nitrides. The difference is formed during the high-temperature annealing because of the nitrogen content in the metal and the different atmosphere for the annealing. As a result, three different conditions were obtained:

Alloy 1. The aluminium is removed from the solid solution and is combined in large nitrides during decarburizing annealing. Irrespective of the atmosphere during the high-temperature annealing only insignificant additional nitride precipitation takes place in the metal. Such a condition results in a structure with a grain size of about 1 mm and the absence of texture after annealing.

Alloy 2. During high-temperature annealing in an atmosphere of 5 % H_2 + 95 % N_2 + 0.001 % NH_3, intensive additional precipitation takes place in the metal, i.e., aluminium is removed from the solid solution. As a result, the grain size is 15–50 mm and the orientation is {110}<001>.

Alloy 2. During high-temperature annealing in argon all the aluminium remains in solid solution. As a result the structure has a grain size of 0.3–1.0 mm, more than 60 % of which have the {111} plane in the sheet plane.

Secondary recrystallization occurs only in condition 2. **Figure 5** gives data on the grain (mean size) growth kinetics and the change in the aluminium nitride parameters during high-temperature annealing.

The most intensive grain growth is observed in condition 1, when all the aluminium is removed out of the solid solution as coarse nitrides, the quantity of which is small and does not exceed 10^{12} cm^{-3}. Grain growth is practically absent in condition 3 with the temperatures below 900–940 °C, when all the aluminium is in solid solution. In condition 2 grain growth starts between 800–850 °C and aluminium is removed out of the solid solution in the form of nitrides. However, at 900 °C and higher the grain growth is inhibited prior to the start of secondary recrystallization. In this period, plenty of nitrides are observed at the grain boundaries.

The conclusion is made that the aluminium located in the solid solution is an effective tool for slowing down grain growth at temperatures of 900–940 °C. Aluminium segregation is not observed. It is assumed that the inhibition is connected with a reduction in the diffusive mobility of the metal atoms in the presence of aluminium. Nitrides start to inhibit grain growth when grain boundaries collect a certain quantity of nitride in their motion.

But what is the reason for grain growth being mainly with the {111}<uvw> orientation during nitride quantity growth? Research has shown that additional nitride precipitation in different degrees occurs in grains of different orientations. As was mentioned above, the maximum nitride quantity is precipitated in grains with orientation {111}<uvw>, and the minimum for {110}<uvw>. In this case grain growth is well explained by the effects of energy fields created by nitrides. The driving force for grain growth is also the energy of the grain boundaries. The inhibiting action of the energy fields is unequal, with inhomogeneous nitride distribution in the grains of various orientations from two adjacent grains, and it results in grain growth. The heterogeneity of the nitride distribution determines a boundary motion direction. It appears from the results that the motion of the boundaries occurs in the direction of grains with a smaller nitride density, i.e., in the direction of fields with a lower energy. **Figure 6** indicates the direction of the boundary movement with an arrow. This may explain the second discrepancy.

Thus, with homogeneous nitride distribution along all the grains, their quantity being less than 10^{13} cm^{-3}, and aluminium being removed out of the solid solution, grain growth is already possible at relatively low temperatures. With an increase in the quantity of nitride the temperature of the initial growth will increase. There is no preference for grain growth in any particular orientation in this case. It can be illustrated by the results obtained for condition 1 (alloy 1). However, the nitride density was low in this case. With an inhomogeneous nitride distribution in the grains of various orientations, grain growth starts with an additional inhomogeneous nitride precipitation, i.e., aluminium removal out of the solid solution. The preference for growth in a certain grain orientation is observed in this case. In the present case these are the grains with the {111}<uvw> orientation.

The process of grain growth is of great importance for dynamo-steel making, based on the technology described above. Now the level of this steel's magnetic properties is basically determined, and it is directly proportional to the grain size. The ratio of a components quantity in the texture may be insignificantly changed with existing possibilities. The control of the steel microstructure and texture is carried out through varying the temperature-time conditions and the values of the rolling reduction and the heat treatment. Attempts to use phase-inhibitors have not proved to be successful, since the anisotropy of the steel properties increased significantly. For these steels the anisotropy should be minimised. Experiments on the introduction of surface-active chemical elements, such as phosphorus, etc., into the steel, are now under way [12]. The experience of steelmaking with a 2.8–3.2 % content of Si has shown that phosphorus introduction makes it possible to improve the magnetic properties of the steel. However, the mechanism for the positive influence of phosphorus has not yet been discovered. In spite of this, there are prospects in this direction.

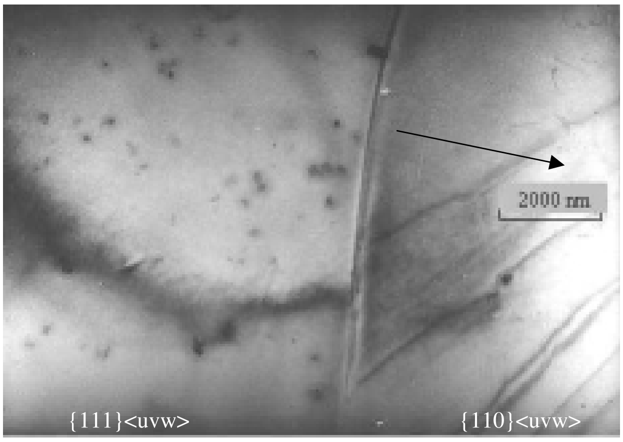

Figure 6: Nonhomogeneous distribution of nitrides in the grains {111}<uvw> and {110}<uvw>

Slika 6: Nehomogena porazdelitev nitridov v zrnih {111}<uvw> in {110}<uvw>

4 ABNORMAL GRAIN GROWTH OR SECONDARY RECRYSTALLIZATION

A lot of research has been dedicated to this process. However, the true mechanism of secondary recrystallization is not yet been clarified. Reference [13] reviews all the proposed mechanisms of secondary recrystallisation, and the author came to the important conclusion that there was no unified mechanism of secondary recrystallization. This can be explained by the variety of the chemical compositions of the alloys and the technologies of their manufacture. Secondary recrystallization or abnormal grain growth involves the growth of single grains due to the surrounding matrix up to sizes exceeding the initial size by a factor of a thousand. The driving force of secondary recrystallization is the energy of the grain boundaries. Now it is believed that the secondary recrystallization occurs as a result of the growth of single grains that have a dimensional advantage. The complexity of the problem involves the definition of the formation mechanism of such coarse grains – nuclei or centres of secondary recrystallization in a polycrystalline matrix. The following points of view are true: 1. Coarse grains are formed during primary recrystallization. 2. Coarse grains are formed due to the coalescence of grain groups with a close orientation in a matrix after primary recrystallization. 3. Coarse grains are formed due to the existence of special, highly mobile boundaries with adjacent grains of a primary recrystallization matrix. 4. Coarse grains are formed out of grains that have a smaller energy in comparison with other grains.

In transformer-steel production with a sharp {110}<001> texture based on the described technology, the mechanism of separate grains' energy advantage is the most likely. Further analysis has resulted in the conclusion that the mechanism realization of highly mobile boundaries is very likely too. The latter is possible if the boundaries' properties are asymmetrical.

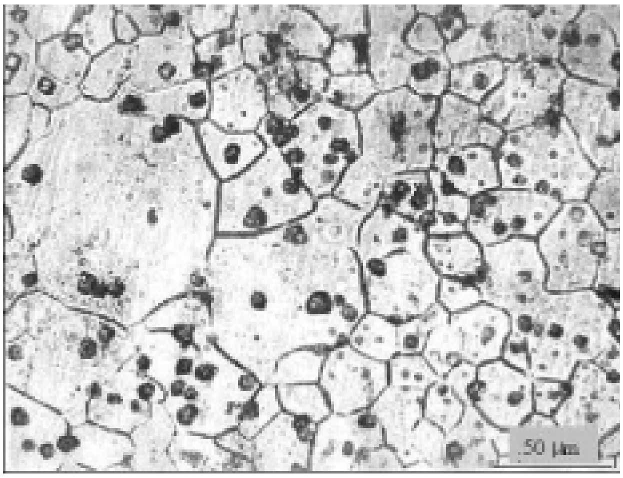

Figure 7: Grains {111}<uvw> in the matrix of the primary recrystallization at temperature 900 °C
Slika 7: Zrna {111}<uvw> v matriksu po primarni rekristalizaciji na temperaturi 900 °C

The selection of a separate grain-energy advantage is based on the authors' researches. During high-temperature annealing the following regularities are observed:

- After primary recrystallization grains with a {110}<001> orientation have no dimensional advantage over other grain orientations;
- Secondary recrystallization precedes the grain growth;
- With grain growth there are objective changes of texture – grain growth of the {111}<uvw> orientation owing to other grain orientations. This growth is particularly intensive before the beginning of the secondary recrystallization;
- The centres of secondary recrystallization with a {110}<001> orientation emerge in a very short period of time;
- Grains with a {111}<uvw> orientation have a higher quantity of aluminium nitrides compared to the grains of other orientations, but it is the least in some grains {110}<001>;
- Secondary recrystallization starts when the maximum is reached during the further reduction of nitride density;
- Secondary recrystallization starts with nitride density, which varies within wide ranges. For example, starting from $3.1\cdot10^{13}$ cm^{-3} up to $9.3\cdot10^{13}$ cm^{-3}, with grain sizes from 0.018 mm to 0.020 mm.
- Before the beginning secondary recrystallization the dimensional advantage belongs to some grains {111}<uvw>, instead of {110}<001>, **Figure 7**.

All this testifies to the fact that secondary recrystallization is caused, not by the fact that a certain degree of matrix stabilization with nitrides is reached, but by changes of texture and microstructure caused by the additional inhomogeneous precipitation of aluminium nitrides taking place during grain growth. The assumed mechanism considers the energy advantage of single grains and results from the experimental supervision set forth above.

The assumed mechanism of secondary recrystallization, sharp {110} <001> texture formation, consists of the following. In the process of grain growth with a {111}<211> orientation at the expense of a neighbouring grains, it inevitably will sprout up to a grain {110}<001> with the smallest quantity of aluminium nitride. Thus, the return growth of the grain {110}<001> due to {111}<211> and its transformation into coarse grain – centre of secondary recrystallization – is started. Some conditions are necessary for this mechanism to occur:

- The grains with the orientation {110}<001> have less energy than the inter-growing grains {111} <112>, owing to a smaller (by an order of magnitude) nitride density. From the above-mentioned example, it follows that the nitride density in the grains {110}<001> is up to $5\cdot10^{12}$ cm^{-3}, and it is equal to 10^{14} cm^{-3} in the {111}<211> grains. The size of the nitride grains of various orientations is practically identical. The calculation demonstrates that the area of the particles' surface, i.e., the energy of the particle-matrix interface and the elastic energy created by the nitrides in the {110}<001> grains is 20 times less. That grain energy became identical, and the {111}<112> grain growth is required owing to neighbouring grains and an increase in their volume by 20 times. And, in attached grain sections {111}<211> there should be no nitrides. Some grains' {111}<uvw> volume growth by only 8 times was experimentally observed. Thus, the energy of {110}<001> grains will be less than the energy of coarser grown grain {111}<211> by not less than 10 times.
- Grains segments {111}<211> inter-growing to grains {110}<001> have a small nitride quantity as with boundary motion they are moving with a boundary, accelerated by coalescence and dissolution in a boundary.
- Favourable for the growth orientation of these grain lattices owing to each other.
- The reduction of the general nitrides density in this period.

By virtue of this condition, grains with a {110}<001> orientation absorb grain sections of {111}<211> grains first of all, and for a short period of time they are attached to them in the process of collecting recrystallisation, and they are free of most nitrides. This allows them to gain a dimensional advantage and to turn into centres of secondary recrystallization. Before the start of secondary recrystallization such {111}<211> grain sections amount to 2–8 grain areas of average size. The {111}<211> grains absorb {110}<001> grains, neighbouring with {111}<211> grains, after the completion of primary recrystallization, during grain growth. This is confirmed by the pole-density reduction of the {110}<uvw> orientation at the beginning of secondary recrystallization. They may not grow due to the {111}<211> grain, as boundaries will meet a lot of nitrides on the way, and a stronger field effect onto the boundaries by the {111}<211> grains. In contrast, {111}<211> grains, which have unified {110}<001> grains, will reduce its energy and matrix due to a reduction of the boundaries' area.

Such a mechanism of secondary-recrystallization centres formation should be less sensitive to the nitride density at the moment of the start of secondary recrystallisation, and this is confirmed experimentally. The main thing is to create the above-stated texture and structure composition.

Many researchers have examined the mechanism of highly mobile boundaries [14–17]. They consider that the centres of secondary recrystallization with a {110}<001> orientation result from the availability of highly mobile boundaries between {110}<001> and {111}<211> grains. However, research has shown that the presence of a highly mobile boundary was required, but an insufficient condition, for secondary recrystallisation, and moreover, for obtaining a sharp texture during secondary recrystallization {110}<001>. The additional condition is necessary. In [15–16] the increased size of the {110}<001> grains is considered to be such a condition. However, experiments did not prove it. In the above-stated assumed mechanism a difference of grain energies of various orientations is considered as such a condition.

An analysis of the known data has suggested the idea that the realization of the mechanism of highly mobile boundaries is possible if the boundary properties are asymmetrical. Boundary motion is the transition of an atom or atom groups from one grain to another. If the speed of transition from a {111}<211> grain to a {110}<001> grain will be higher than the reverse, the realization of the mechanism of highly mobile boundaries is possible. Ideally, grains of these orientations contact with each other with the planes {211} and {001}, respectively. From the point of view of geometry, the plane texture {001} is simpler, and the distance between the atoms in this plane is less, compared to the {112} plane. It can cause asymmetry of the boundary properties of the grains with these orientations. The value of the boundary property asymmetry is likely to be changed by the atoms of admixtures and alloying elements. Their adsorption on the {211} and {001} planes will occur in a different way. This refers to the chemical elements included into a structure of phase-inhibitors. Researches showed that the speed of coalescence and dissolution of phase-inhibitors is greater on grain boundaries than inside grains [18]. Atoms of these elements will be the first to be located at boundaries and change the value of their asymmetry.

5 CONCLUSIONS

Both the known and the assumed influence mechanisms of chemical elements and aluminium nitride particles dissolved in a solid solution have been discussed. The two main influence mechanisms consist of the following: the first in recovery and recrystallization processes inhibition; the second in the change of grain energy with different orientation. It was assumed that tension fields are of major importance for growth inhibition and the selection of grain-boundary motion direction. Two hypotheses have been put forward on the asymmetrical properties of grain boundaries and the availability of grains with different energy value, which help explain the secondary recrystallization.

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