Effect of work hardening and recrystallization annealing on structure and properties of low-carbon steel wire

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Abstract: *In this work it was studied the influence of low-carbon steel wire annealing at different temperatures from 500 to 750 °C that was obtained by cold drawing on its structure and properties. It was sown that bigger work-hardening during the drawing process leads to the decreasing of collective recrystallization starting temperature, which could resulting in obtaining less plastic properties than are appear after lower annealing temperatures with additional loss of strength. The found effects give valuable information for cases when it is desirable to obtain the minimal strength or the maximum plasticity of the wire.*
Keywords: WIRE, COLD DEFORMATION, ANNEALING, LOW-CARBON STEEL, *RECRYSTALLIZATION*.

1. Introduction

Low strength properties are rather often needed from low-carbon steel wire. For instance, wire for fastening nails in pneumatic tools should have tensile strength on the level less than 350…360 MPa with not very high plasticity, because it ought to be able easily tear. Sometimes wire is obtained in several stages of cold drawing with intermediate annealing. For this case the wire condemned to the following high cold deformation must be primary as more plastic as it possible. High plasticity is also desired for many cases of ready product, for example, it is crucial for knitting wire and sometimes for welding wire. Low strength at this situation is also may be needed but not as much as for the first example. Moreover, very low strength is not desired for welding wire to prevent its bending and twisting when feeding from a welding machine. So the optimal tensile strength here is on the level about 400 MPa or even some more.

Industrial experience meets a problem that properties of low-carbon wire after recrystallization annealing are not always simply predictable. The desirable level of properties is obtained in some cases but in others is not in spite of the same annealing time and temperature. Also it was noticed that properties sometimes could be highly nonlinearly and spasmodically depend on the annealing temperature. Thus it was deiced that the problem needs a detailed investigation.

It is known, that the main cause of strength decreasing of cold-worked metal during annealing are such structural phenomena as recovery and recrystallization processes [1]. There could occur primary and collective recrystallization [2]. The first one is proceeding at deformed material and resulting in obtaining equiaxial grains instead of deformed ones. These grains could be even smaller than the parent ones [3]. Another process is collective recrystallization. It could appear even in not deformed or annealed materials and may be described like coagulation and Ostwald ripening of the grains [4]. Recrystallization is accompanied with recovering process which is decreasing of dislocation density that leads to the loss of material strength [2]. Recovering could appear even without noticeable recrystallization, especially at lower temperatures [5]. It is distinguished dynamic and static recrystallization [2]. The first one proceeds simultaneously with deformation and is mostly has to do with hot deformation processes which are not under the consideration in this study. So here we have static recrystallization that occurs during annealing after deformation.

The aim of this study is to determine dependence between previous work hardening and mechanical properties of low-carbon steel wire after annealing at different temperatures and also to specify influence of structure changes on them.

2. Materials and experimental details

The first supposed reason of different behavior of the material under the same annealing regime that ought to be considered is its chemical composition. However, the statistical analysis of the industrial protocols did not reveal a significant influence of any element of the composition in its present variation range on strength properties after annealing. It was found a slight dependence of tensile strength in the deformed state on Cr, N, and As impurities amount, but even them are not very valuable and low statistical reliability. It only could be noticed that Cr and N might lead to some increase of deformed metal strength and As may slightly decrease it. Thus the most reasons in this case are not in chemical composition, moreover it could be considered as averagely same for all the studied specimens. The averaged chemical composition of the studied samples material is given in table 1.

***Table 1. Chemical** composition of the studied material (mass %)*

Elements, %							
C	Mn	Si	S	P	Cr	Ni	Cu
0,04 – 0,06	0,31 – 0,37	0,026 – 0,080	0,004 – 0,026	0,008 – 0,017	0,02 – 0,07	0,02 – 0,13	0,03 – 0,05
As	N	Ti	B	V	Mo	W	Sn
0,002 – 0,005	0,005 – 0,011	0,001 – 0,005	< 0,0004	0,001 – 0,005	< 0,01	< 0,027	0,003 – 0,005

It is a low-carbon steel that contains only some amount of Mn as alloying element. Other elements are just on the typical level of impurities.

For the main process where the final properties of the material are forming is annealing, an experimental study was carried out which consist in annealing of the wire specimens at different temperatures during different keeping times. The initial samples were taken directly after cold deformation (drawing) which is wire production process. They have different initial level of work hardening and different diameter. Here are the diameters and the average values of ultimate tensile strength: 3.0 mm, 849±17 MPa; 2.0 mm, 882±8 MPa; 1.2 mm, 1040±18 MPa. Initial tensile strength before the cold deformation was similar to all the samples, about 360…380 MPa. The studied annealing temperatures were following: 500, 600, 650, 700, and 750 ºC with temperature hysteresis about ±15 ºC. The annealing time varied from 2 to 5,5 hours. The annealed samples were used to study tension mechanical properties and microstructure.

Also additional experiments were carried out, where the samples of wire were undergo heating at 930 ºC (to austenite area) during about 5 – 7 minutes and then cooled in the air or salt water (~10 % of NaCl solution). After cooling some of the specimens were tempered at 650 and 700 ºC during about an hour.

3. The results of the experiments and their discussion

The obtained results showed that considered varying of annealing time does not make a valuable impact on the mechanical properties. But according to the industrial data lesser times could significantly affect them, especially first 30 minutes. Rather more important factors are the annealing temperature and work hardening made through drawing deformation. The effect of ultimate tensile strength (a) and yield strength (b) on the annealing temperature are shown on figure 1. A plot are given for every of the specimens type.

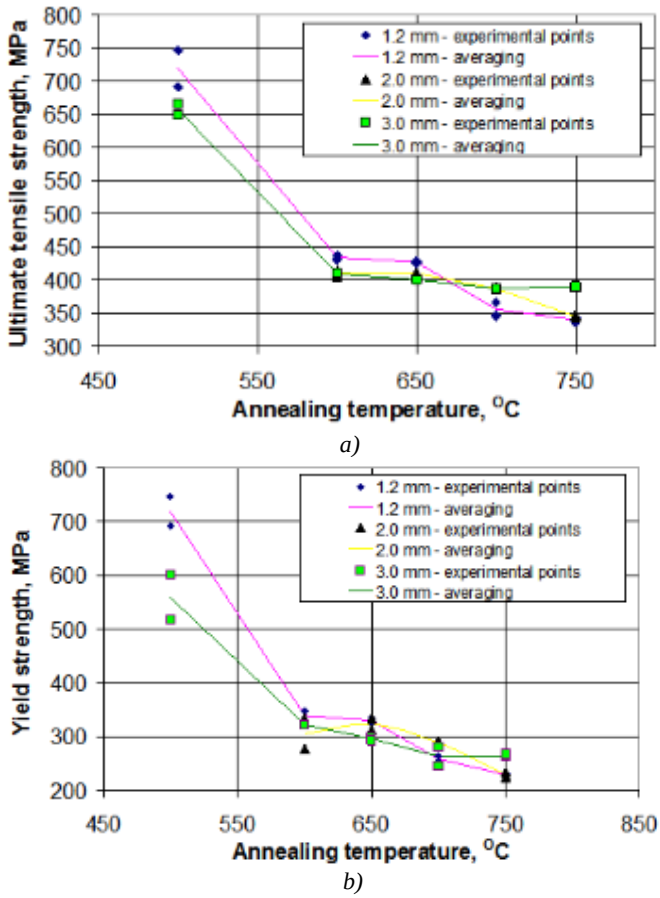


Fig. 1. *Effect of annealing temperature on strength properties of the low-carbon steel wire*

From the obtained data it is clearly perceptible some difference in regularity of changing strength with annealing temperature alteration for different types of the samples, which are distinguished only with their work hardening, but not the composition.

After 500 ºC annealing the obtained strength is already significantly less than in the corresponding deformed state but yet rather high. It is notable that for this case the higher work hardening is resulting in somewhat bigger strength after annealing. More distinctly this effect is observed for the yield strength property, there the gap between average values for the samples with initial 849±17 MPa and 1040±18 MPa correspondently after annealing becomes about ~ 160 MPa (~ 550 and 710 MPa correcpondently).

The same tendency remains at annealing temperatures 600 and 650 ºC – more work hardened metal has higher strength also after annealing. It is noteworthy that at this temperature range there is no statistically significant difference in tensile strength obtained after annealing at 600 and 650 ºC for the corresponding specimens types. Only for the less work hardened samples there might be a slightly decrease in yield strength.

Higher annealing temperatures give the most notable results. Fore the least work hardened samples ultimate tensile strength remains almost at the same level as it has been obtaining after 600 and 650 ºC annealing. There may be only just insignificant decline. The decrease in yield strength here is some more significant, however, it is almost linear without any leaps. In contrast to that for the most work hardened samples we can see an abrupt additional decline of tensile strength after annealing at 700 ºC comparing with one obtained after 600 and 650 ºC annealing. Almost the same strength level remains also after annealing at 750 ºC, may be just slightly less. Wire samples of Ø 2 mm which are some more work hardened than Ø 3 mm ones shows the same behavior, but only at 750 ºC annealing. At lower annealing temperatures only insignificant tensile strength decrees could be observed and the values are close to ones for Ø 3 mm specimens.

It is remarkable that after annealing at 750 ºC strength values (both ultimate tensile strength and yield strength) for the Ø 1.2 and Ø 2 mm samples appear very similar: ~ 340 MPa vs. ~ 346 MPa and ~ 229 MPa vs. ~ 230 MPa correspondently. Taking into account that ultimate tensile strength values for the most work hardened wire (Ø 1.2 mm) are almost not change in the range of annealing temperatures 700 – 750 ºC, there is a reason to suppose that these values are ultimate least that could be obtained after recrystallization low-temperature annealing for this material.

Not the less interesting looks the dependence of relative extension, which is a measure of material plasticity, on the annealing temperature. The corresponding plot is given on figure 2.

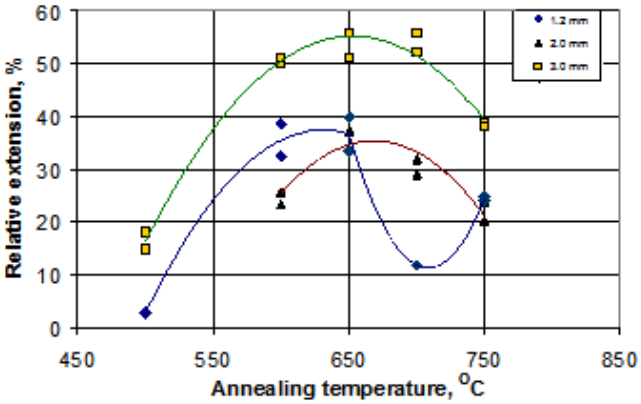


Fig. 2. *Effect of annealing temperature on relative extension of the low-carbon steel wire*

The first notable feature here is that the least work hardened wire (Ø 3 mm) shows the highest relative extension values for each corresponding annealing temperature. However, just more hardened wire (Ø 2 mm) demonstrates the least of its values fore the most of the considered annealing temperatures. The most worth notable at this case is that for the most hardened wire (Ø 1.2 mm) is observed a collapse of plasticity after annealing at 700 ºC, but 750 ºC annealing leads to its higher values. The common thing for all of the considered samples types is that the maximum values of relative extension could be obtained after annealing at about 650 ºC. Then, at higher temperatures, it anyway tends to decline. Annealing temperature does not the same significantly affect the values of constriction ratio (another plastic property).

The observed behavior of plasticity, especially for the samples type of the most work harden, were the plasticity collapse is present, is to be compared with mentioned above dependence of strength on the annealing temperature. As we can see, the plasticity decline after annealing appears at 700 ºC, at the same temperature in these specimens also occurs the additional strength decrease, which is more clearly seen for the yield strength for Ø 1.2 and 2 mm samples and also for ultimate tensile strength for Ø 1.2 samples. So we have simultaneous decrease of both strength and plasticity at rising annealing temperature to 700 ºC and higher comparing with properties that could be obtained at lower annealing temperatures.

As it was mentioned above and is seen from the plots, the dependence of strength and plasticity on annealing temperature for different specimen types is not the same. And the only difference between the types is the level of work hardening. To make this dependence more clear, plots are given on the figure 3, which demonstrate an influence of initial ultimate tensile strength in cold-worked state on one obtained after annealing at different temperatures. Data for 500 ºC annealing are not given on this plot because of comparably higher values, which could make the plot system less clear.

For the temperatures 600 and 650 ºC we observe almost linear dependences which show that the more work hardened the material was, the higher its strength appear after annealing. And here the difference in the dependences is almost absent for both of these temperatures. For the temperatures 700 and 750 ºC the

dependencies appear quite different, and the most specific fact in these cases is that they change their directions – higher strength in a work hardened state leads to less one after the annealing.

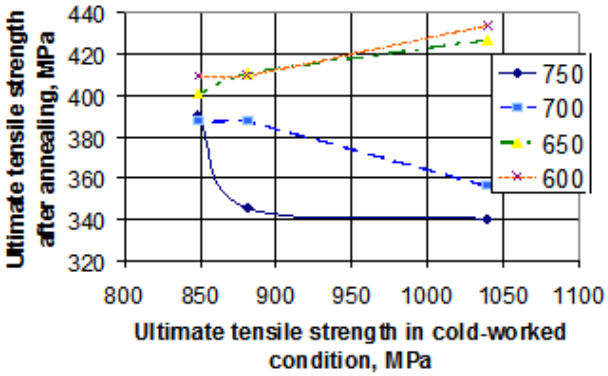
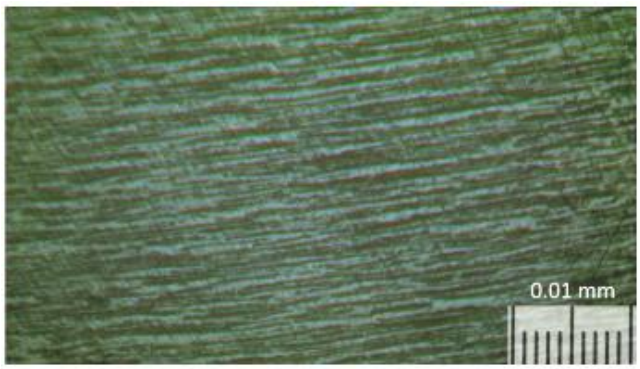


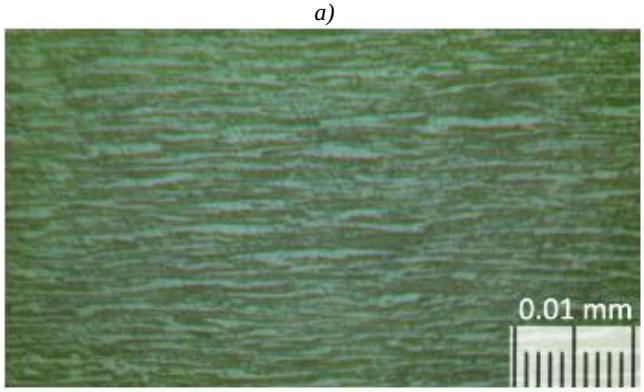
Fig. 3. *Effect of initial strength after drawing deformation on the strength obtained after annealing*

For the annealing temperature 700 ºC increase of the initial strength in cold-worked state from 849 to 882 MPa almost does not affect the resulting one obtained after the annealing. However, hardening up to 1040 MPa does – the resulting tensile strength appears significantly less. For 750 ºC annealing even work hardening to 882 MPa is enough to obtain significantly lesser strength after annealing than it could be received for hardened to 849 MPa specimens. Additional work hardening leads to just slight decrees of the strength obtained after annealing at this temperature.

It seemed that the causes of the observed behavior are in recrystallization that has a place during the annealing processes. Therefore it was carried out an exploration of the annealed metal microstructure. Figure 4 shows the microstructure of deformed metal for each type of the samples.



a)



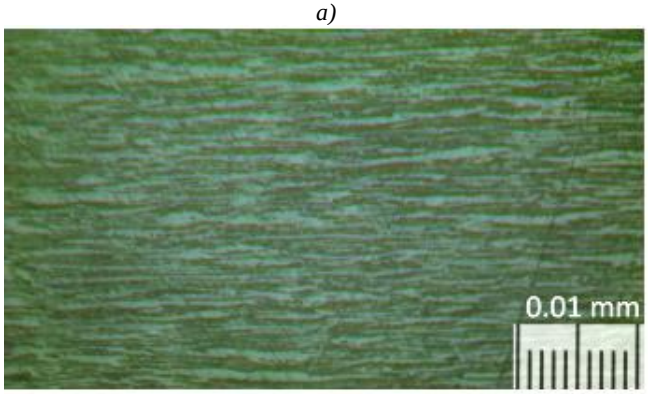
b)

c)

Fig. 4. *Microstructure of the of low-carbon steel wire in work hardened state: a) 1.2 mm, 1040±18 MPa; b) 2.0 mm, 882±8 MPa; c) 3.0 mm, 849±17 MPa*

The structure in the deformed state is quite ordinary. It consists of elongated ferrite grains, as longer and thinner as more significant drawing deformation the metal was affected. It is noteworthy that the structure after 500 ºC annealing remains rather qualitatively similar. As we can see from figure 5, there also observed the same elongated grains.

a)



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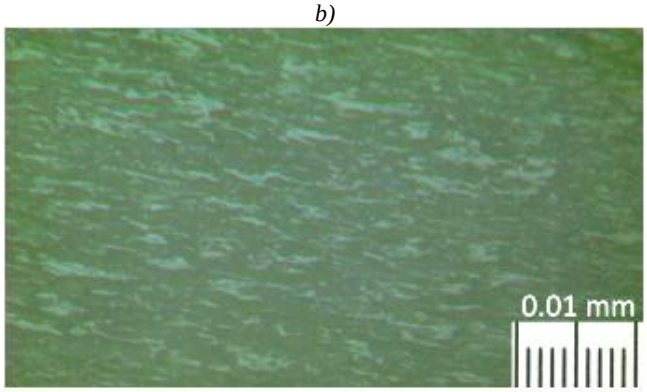
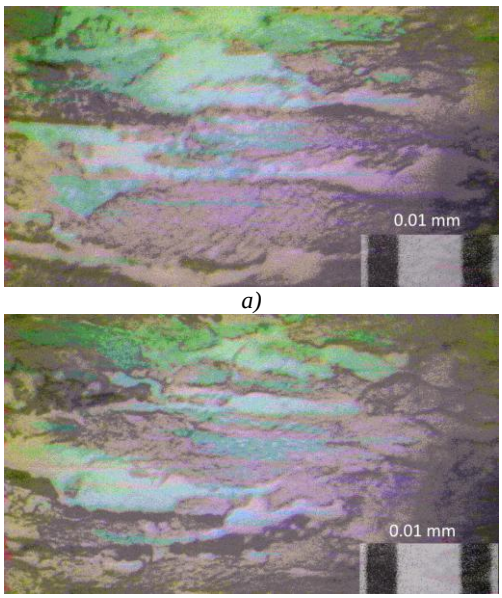


c)

Fig. 5. *Microstructure of the of low-carbon steel wire after annealing at 500 °C: a) 1.2 mm, 1040±18 MPa; b) 2.0 mm, 882±8 MPa; c) 3.0 mm, 849±17 MPa*

Photos taken at bigger magnification (×1000) and shown at figure 6 suggest supposing that some initial recrystallization might occurred near the grain boundaries even during annealing at this temperature. However, it seems like the main cause of the observed some strength loss in this case is merely recovering, but not recrystallization.



a)

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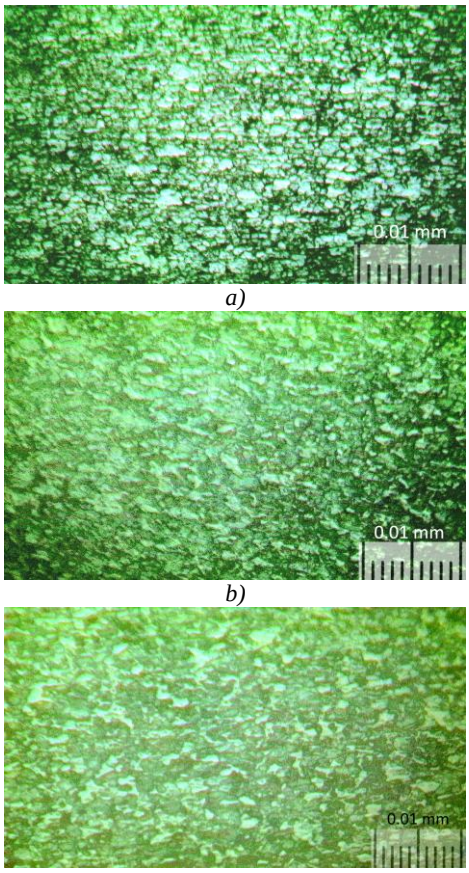
Fig. 6. *Microstructure of the of low-carbon steel wire after annealing at 500 °C, ×1000*

Annealing at 600 ºC leads to an active primary recrystallization. As it seen from figure 7 the structures is formed by almost equiaxial small grains. The grains are the smaller the higher was the work hardening.



a)

b)

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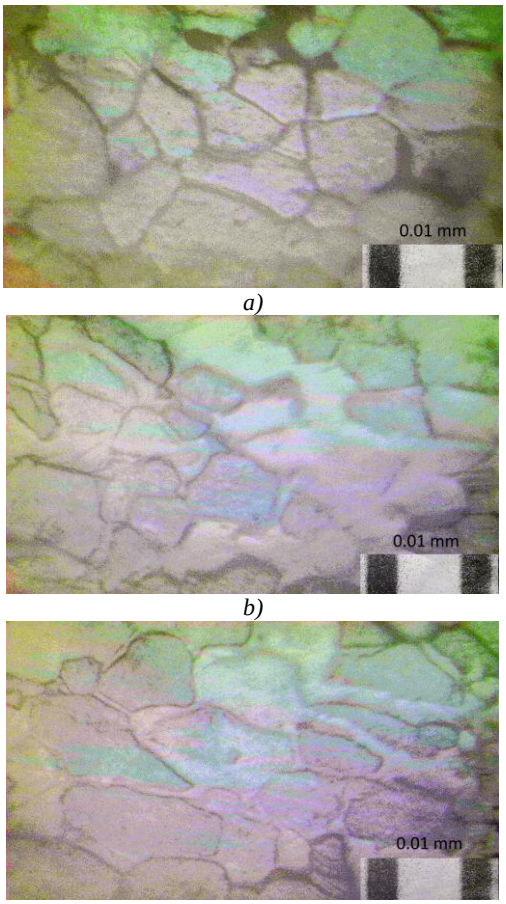
Fig. 7. *Microstructure of the of low-carbon steel wire after annealing at 600 °C: a) 1.2 mm; b) 2.0 mm; c) 3.0 mm*

From the structures just above and more clearly from the figure 8, where are shown structure photos of larger magnification, in seams like in Ø 1.2 samples the primary recrystallization is rather complete, but in the other types is not.



a)

b)

c)

Fig. 8. *Microstructure of the of low-carbon steel wire after annealing at 600 °C: a) 1.2 mm; b) 2.0 mm; c) 3.0 mm*

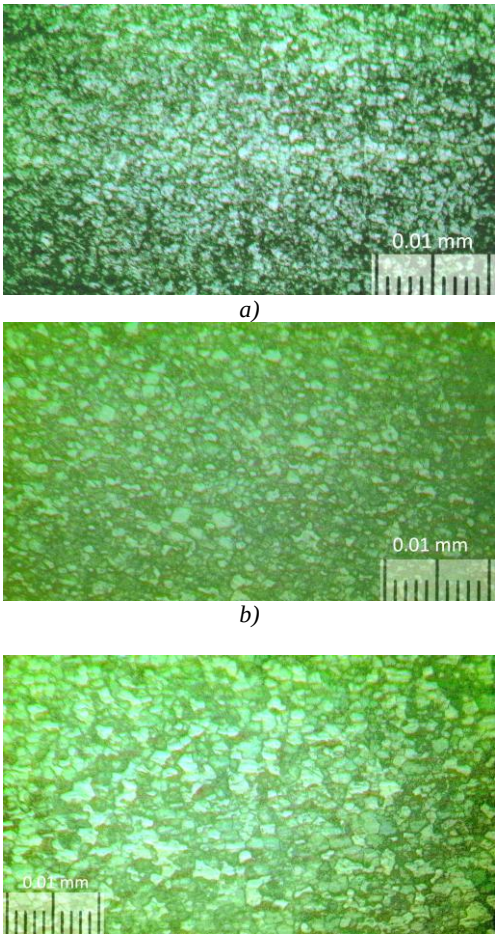
The difference in the structure parameters between the corresponding samples annealed at 600 and 650 °C is insignificant. We also have small equiaxial grains of similar size. However, it could be mentioned, that at this temperature has a place more complete primary recrystallization and grains appear more equiaxial than they were after 600 °C annealing, especially for less initially work hardened samples.

During annealing at 700 °C starts collective recrystallization – grains become lager. Its appearing highly depends on the level of work hardening of the material, so, as we can see from photos on figure 10, it could vary through the cross-section of the wire.

Figure 10 (a) shows the surface layer of the wire, which is more work hardened that its core. So collective recrystallization near the surface goes more actively, which is resulting in obtaining larger grains in that place. In the Ø 3 mm wire, which was least deformed, collective recrystallization process almost not started, but primary recrystallization is complete. Only in a thin surface layer, not more than 50 microns, which is likely was the most work hardened, some signs of collective recrystallization might be seen. Collective recrystallization occurs in Ø 2 mm wire at this temperature. However, its work hardening does not significantly vary through the cross-section, so the grain size in it is close to one in the core of Ø 1.2 mm wire. Thickness of the collective recrystallized layer in Ø 1.2 mm wire is up to 0.45…0.50 mm, which is a significant part of the cross-section. Hence we observe the mentioned abrupt strength and plasticity loss. In contrast, more evenly deformed Ø 2 mm wire shows more facile strength decrease by rising the annealing temperature.

Annealing at 750 °C initiates collective recrystallization in all of the specimens types. In Ø 1.2 mm samples in this case it touches

almost all the cross-section, as in could be seen from figure 11 (a), but inner layers still have somewhat smaller grains than the surface ones.



a)

b)

c)

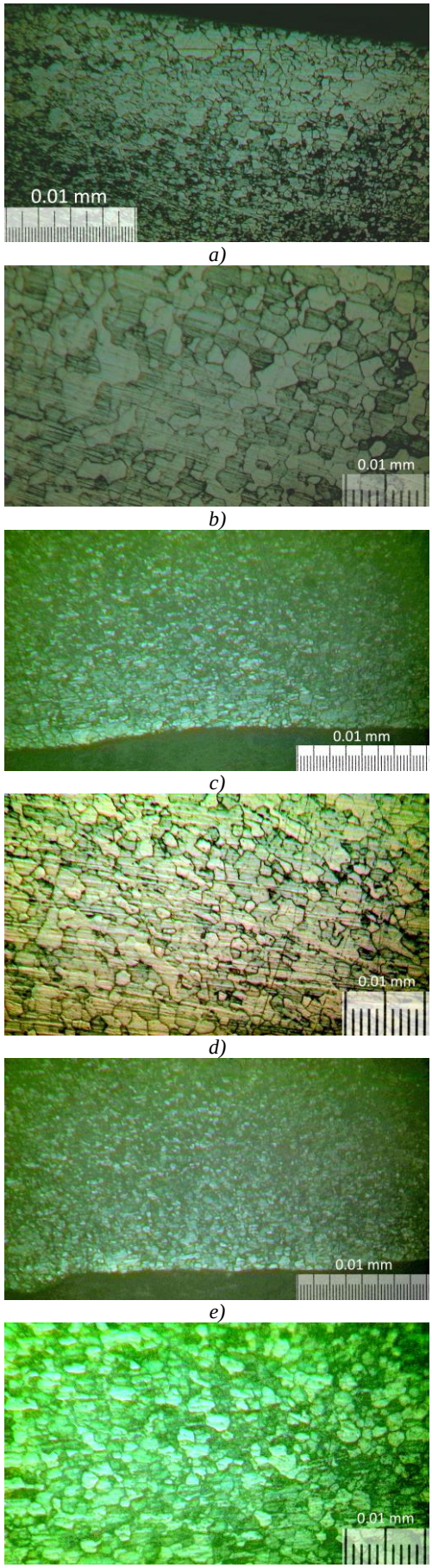
Fig. 9. *Microstructure of the of low-carbon steel wire after annealing at 650 °C: a) 1.2 mm; b) 2.0 mm; c) 3.0 mm*

Significant collective recrystallization in this case appear through all the cross-section of Ø 2 mm wire (figure 11 – c, d). Certainly that results in additional loss of strength to the level close to one that is been obtained for Ø 1.2 mm wire after annealing at 700 and 750 °C.

Collective recrystallization also has a place in Ø 3 mm specimens. However, it occurs only in the surface layer with thickness not more than 0.5 mm, which is likely was the most deformed and work hardened. In contrast to Ø 1.2 mm wire in this case it forms significantly lesser part of the cross-section. Hence, we do not observe such additional loss of strength as it was for Ø 1.2 mm at 700 °C, where the similar situation appears. Although a decrease of the plasticity is definitely seen.

The hypothesis that work hardening facilitates collective recrystallization is additionally proofed by micro hardness measurement, which change through the cross-section in a deformed state strongly correlates with structural layers after annealing. Figure 12 shows the micro hardness variation from the surface to the core of the wire in a work hardened sate.

In Ø 1.2 and Ø 3.0 mm wire surface layer appear significantly harder, Ø 2.0 mm wire does not have such difference, and its hardness only slightly decreases from the surface to the core. Thus in Ø 1.2 and Ø 3.0 mm wire after a certain annealing temperature we obtain a surface layer in which a collective recrystallization evolves rather more significantly, but in Ø 2.0 mm wire we do not observe such behavior.



a)

b)

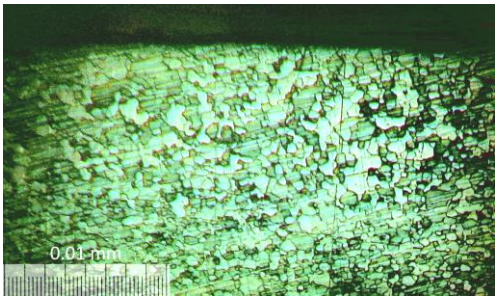
c)

d)

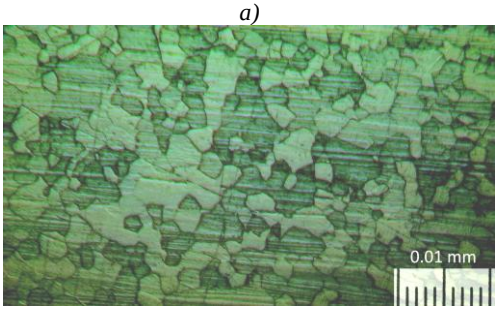
e)

f)

Fig. 10. *Microstructure of the of low-carbon steel wire after annealing at 700 °C:a) 1.2 mm, ×40; b) 1.2 mm, ×100; c) 2.0 mm, ×40; d) 2.0 mm, ×100; e) 3.0 mm, ×40; f) 3.0 mm, ×100*



a)



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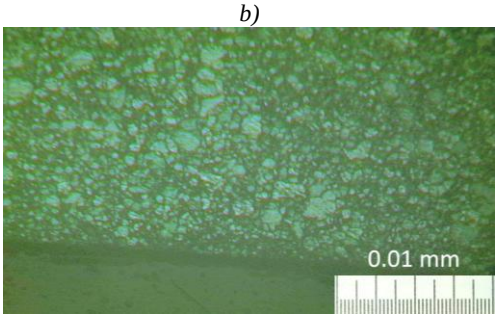

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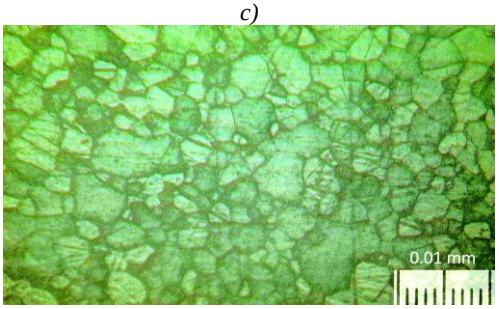

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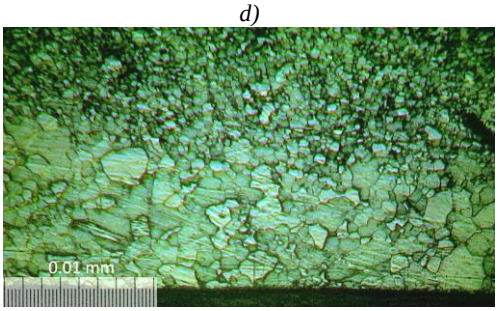

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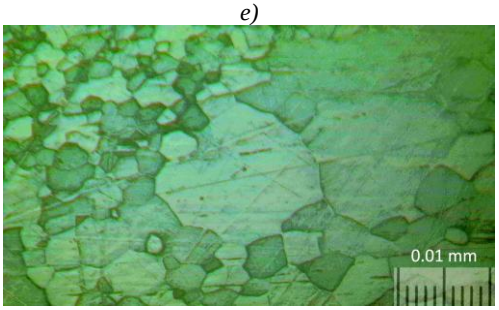
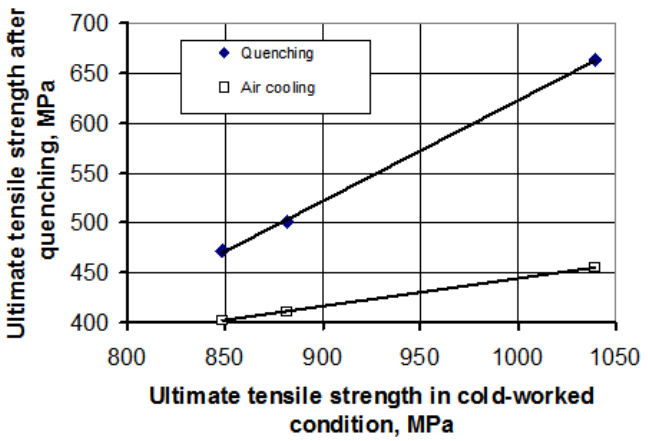


f)

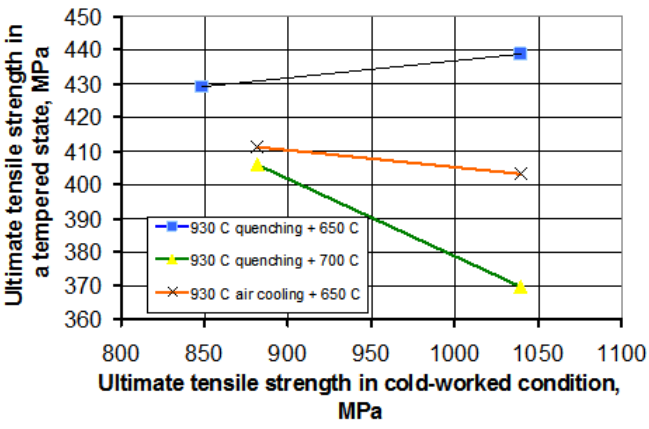
Fig. 11. *Microstructure of the of low-carbon steel wire after annealing at 750 °C: a) 1.2 mm, ×40; b) 1.2 mm, ×100; c) 2.0 mm, ×40; d) 2.0 mm, ×100; e) 3.0 mm, ×40; f) 3.0 mm, ×100*

Such low-carbon steels are not much susceptible to quenching, but extreme cooling from the austenite area could have some effect, which was studied for the case if this wire. Work hardened samples were quenched in a ~10 % NaCl water solution, which could provide a fast cooling of rate up to 1000 °C/s or even more. Measurement of the mechanical properties showed that this material after quenching has less strength than in a work hardened state, but it is significantly more than after annealing. Moreover, it was also found a dependence of tensile strength after quenching from one in deformed state. The values are following, the values for the work hardened state are given in parentheses: Ø 3.0 mm – 471 MPa (849 MPa), Ø 2.0 mm – 501 MPa (882 MPa) and Ø 2.0 mm – 663 MPa (1040 MPa). The dependence is almost linear, which could be seen from the figure 13 (a). So we can see that more work hardened material has bigger strength also after quenching. The qualitatively similar to quenching is the dependence for the air cooling but the values are less and rather close to those after annealing at 600 °C.

Tempering at 650 °C after quenching decline strength and the decrease is the more the higher strength was after quenching. This tempering almost neutralizes the effect of work hardening, and only a weak dependence might remain. Rising the tempering temperature to 700 °C leads to a significant strength decline for the quenched samples. The direction of the dependence changes as it was previously observed for annealing. The values are also close to those after the corresponding annealing temperature, may be somewhat higher. So we observe the same effect of greater strength deceasing after more significant previous work hardening.



a)



b)

Fig. 13. *Effect of initial strength after drawing deformation on the strength obtained after quenching and air cooling (a) and after subsequent tempering (b)*

Tempering after air cooling also neutralizes the effect of work hardening and even slightly demonstrates an effect of additional strength decreasing for more work hardened material. There is almost no effect on the specimens that had strength 882 MPa in deformed state. Their strength remains at about 410 MPa, which

could be obtained directly from air cooling. For the samples that had after work hardening strength about 1040 MPa and 455 MPa after air cooling from the austenite area tempering gives its noticeable additional decline down to 400 MPa. Thus, for air cooling even 650 °C is enough to evoke the effect of turning the dependence of obtained strength from one in work hardened state into opposite direction.

The driving force for recrystallization is energy stored within the metallic material, which arises during the lattice strains and accumulates on the crystalline imperfections generated during deformation [1]. According to [6] dislocations are the major contributor into the stored deformation energy. A cold working deformation process could increase the dislocation density in a metal to an estimated $10^{16}m^{-2}$ from about $10^{10}m^{-2}$ in annealed state. Each dislocation is a crystal defect that generates lattice disturbances in form of strains within its vicinity. The increased lattice strain is associated with the increase in strain energy in the metal [1, 7]. If deformation is performed at rather low temperature the defect accumulation to occur resulting in an increase in the stored energy [8]. The deformed state due to defect accumulation and work hardening is thermodynamically unstable, and there is a natural tendency to revert to the or annealed state to minimize its overall energy. However it often requires heating.

The recrystallization that eliminates almost all the deformation induced dislocations in worked metallic materials is known as the primary recrystallization, which is the process merely formation and growth in a deformed matrix of new grains which are distortion free and appreciably more perfect than the proper matrix [1, 9]. This process is propelled by the excess volume energy accumulated during the plastic deformation [10].

Collective recrystallization process isn't conjugated with neither new grains nucleation nor extra grain boundaries formation. It consists in grain growth when the boundaries are disappear (coagulation) and some larger grains consumes the smaller ones (Ostwald ripening). So this process does not need the stored in lattice defects energy that primary recrystallization does, but its driving force is surface energy of grain boundaries [11].

It is known that because more intensive deformation forms more lattice defects, especially dislocations, smaller grains in greater amount appear during primary recrystallization. Thus larger amount of stored in grain boundaries surface energy will be obtained, which is, as was mentioned, is the driving force of collective recrystallization. Hence, the observed effect becomes more clear: more developed grain boundaries obtained after primary recrystallization of more work hardened material also enforce collective recrystallization, which in this case actively starts at lower annealing temperature.

The effect of lowering recrystallization start temperature is described in [12], although for steels of different composition. According to that paper, the recrystallization temperature of this material can be considered to be ~670 °C despite the fact that it has a dependency of cold deformation degree. With increasing of the deformation degree, the recrystallization temperature declines. This data are in very qualitative and numerical accordance with our results. More work hardened metal has lesser temperature of the active collective recrystallization start. So the same annealing temperature could be enough for initially more deformed sample to start the collective recrystallization in it, but for other this temperature still not enough, so we do not observe so active collective recrystallization in it, or in may go partially – only in the surface layer, which work hardening is already enough to collective recrystallization be started in these conditions.

The critical points for the considered steel chemical composition were estimated by the method described in our previous work [13] that consists of primary estimation using empirical formulas from [14] with subsequent elaboration of them by a CALPHAD-like method. The obtained values are as follows: A_1 = 722 °C, A_3 = 918 °C. Hence, annealing at 750 °C actually was performed higher than the pearlitic transformation temperature, although there amount of pearlite is such kind of steel is very small. According to the transformation kinetics calculations, performed by the method from [3333], after slowly cooling (~0.01 °C/s) there may be about 4.6 % of pearlite. Thus, most of the structure even during such annealing actually was ferrite, so the results for the 750 °C annealing might be considered to some extent comparable to those for lower temperatures.

Heating up to 930 °C is heating to the purely austenite area for this steel, according to the given values of critical points. A calculated using the technique from [13] and the corresponding software [15] CCT-diagram for the given steel is given on figure 14.

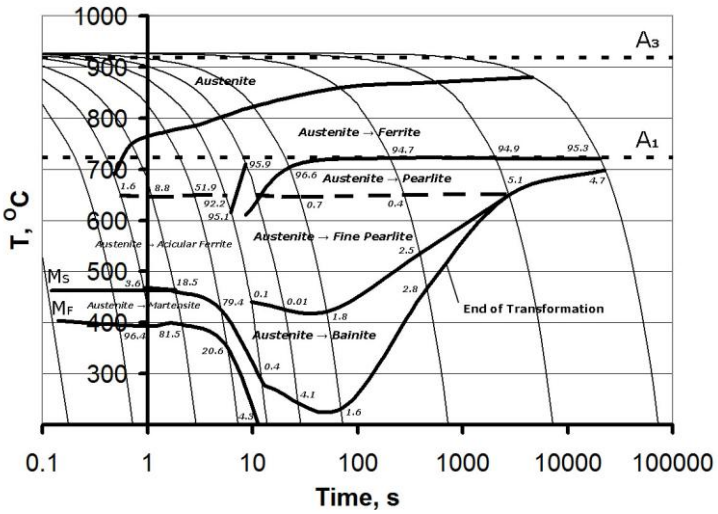


Fig. 14. *Calculated CCT-diagram for the steel of the given composition*

Even though the presented CCT-diagram is merely a simulated estimation it does not much contradict observed behavior of such steel and could be used fore appraisal judgments. Salt water cooling for such small diameter samples is able to give the cooling rate about 1000 °C/s or even more that visually looks like almost immediate. As we can see, such cooling is resulting in purely martensite structure. However, such low-carbon and not alloyed steels do not form hard martensite, here we have so called lath martensite, which is not very hard, although, according to the experimental results the quenched samples have significantly grater strength than annealed or air cooled ones, but their strength is even less than in a work hardened state.

Refrigeration after annealing of a raw bar or ready product in industrial conditions is a rather long term process, which is cooling of several tones of metal with in a quite large furnace (actually with the furnace). It lasts for many hours and the average cooling rate is less than 0.02 °C/s. Hence, it is close to the most right (the lowest) cooling rate of the CCT-diagram. So in such case is obtained a nearly equilibrium structure that consists mostly of globular ferrite with an insignificant amount of coarse pearlite, which could be presented in the structure even like structurally free carbides in ferrite matrix located mostly near ferrite grains boundaries. The part of this pearlite in the structure is less than 4.7 %. No martensite, bainite, acicular ferrite or fine pearlite are not able to form in such conditions.

Analysis of the transformation processes that occur during air cooling seems to be some more complicated. The first reason is presence of various possible transformation types for intermediate refrigeration rates range; the second problem is that for the considered case size (diameter) of the samples may significantly affect cooling process. Also it should be taken into the account that real cooling rate is not constant but slowing with decreasing of temperature difference between a sample surface and environment. Heat transfer coefficient also somewhat changes during the cooling process, actually, as was estimated for this case it varies from 6.6 to 41.6 $W/m^2 \cdot K$ depending on temperature difference and specimen diameter. It is rather difficult to directly measure a cooling curve for such thin samples, but it could be estimated, which is quite enough for a coarse valuation. A model used for the cooling curve

estimation has the following assumptions: during the air refrigeration process only convection heat exchange between the specimen and environment has a place; no forced air fluxes are present; temperature gradient inside the sample is neglected owing to low value of Biot criterion (from $3.3 \cdot 10^{-4}$ to $1.9 \cdot 10^{-3}$); the heat flux is proportional to temperature difference according to Newton's law of cooling; heat transfer coefficient is estimated from the Nusselt criterion, which value is calculated using Prandtl and Grashoff criterions. The calculated cooling curves are shown on figure 15.

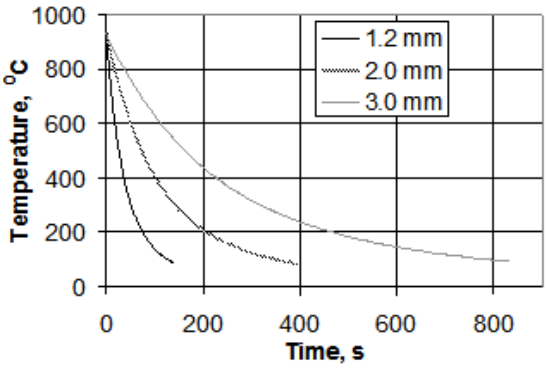


Fig. 14. *Estimated cooling curves for air cooling of the samples*

As we can see, the cooling curves for the wire of different diameter seem quite different. Austenite decomposition kinetics simulation was performed to estimate the influence of this difference of cooling curves on obtained structure. According to the calculations the fallowing final structure characteristics were predicted to be obtained after air cooling:

Ø 1.2 mm wire: 96.1 % ferrite; ~0.5 % fine pearlite; 1.8 % bainite; 1.6 % lath martensite

Ø 2.0 mm wire: 96.2 % ferrite; 1.6 % fine pearlite; 2.2 % bainite

Ø 3.0 mm wire: 95.0 % ferrite; 2.3 % fine pearlite; 2.7 % bainite

Thus, we can see that in spite of visually valuable difference between cooling curves, there is no cardinal difference in the samples structure. For all of the cases it is mostly ferritic with small amount of harder phases. However, there is some difference in the ratio of this phases. In the Ø 1.2 mm wire small amount of lath martensite might be formed, which is not predicted for the thicker specimens. Those may have some fine pearlite and bainite. Nevertheless, this small distinction seems not to make a significant contribution into mechanical properties difference that has been observed. So, there is a reason to believe, that exactly work hardening during deformation is the most valuable factor for the considered cases.

Conclusions

From the observed results, it can be concluded that increased work hardening promotes grain growth at higher annealing temperatures, when collective recrystallization occurs. This is evidenced by the presence of a coarse-grained surface layer, which was hardened to a greater hardness, and the fact that in specimens of Ø 1.24 mm, which had a higher strength in the initial state and underwent a greater degree of deformation than specimens of 3.0 mm, is observed a slightly greater average grain size after annealing. The fact that the mentioned coarse-grained layer in 3.0 mm samples appears at a higher annealing temperature suggests that a large degree of work hardening reduces the temperature of active collective recrystallization start. For this case, exactly occurrence of collective recrystallization at most of the wire material (over the cross section) makes it possible to achieve an additional softening, which is required in some cases from annealing. However, this reduces the resulting ductility and is therefore not desirable in intermediate stages (between drawings). This effect of work hardening on the properties obtained after annealing remains also after quenching and air cooling from austenite area.

For the minimal strength obtaining after annealing the material should to be previously work hardened up to 1000…1200 MPa and then annealed at temperature higher 700 °C during not less than 2 – 3 hours. The complex of mechanical properties in this case will be as fallows: σ_u = 325 – 350 MPa, σ_y = 250…270 MPa, δ = 20…30 %, ψ > 45 %.

To obtain maximal ductility, there are the following recommendations: initial tensile strength of the material condemned to the annealing should be not more than 780 – 850 MPa; the optimal annealing is to be performed at temperature about 640 – 670 °C during not less than 2 – 3 hours. The expected mechanical characteristics after annealing are as follows: $\sigma_u \approx 400$ MPa, σ_y = 300…350 MPa, $\delta \geq 50$ %, ψ = 70…80 %.

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