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Influence of plastic deformation on CCT-diagrams of new-developed microalloyed steel

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Properties

ABSTRACT

Purpose: The aim of the paper is to investigate the influence of plastic deformation and cooling conditions on a structure and a shape of CCT-diagrams of new-developed Nb-Ti-V microalloyed steel.
Design/methodology/approach: The diagrams of undeformed and plastically-deformed supercooled austenite transformations for Nb-Ti-V microalloyed steel were determined. A part of the specimens were austenitized at a temperature of 885°C and next cooled to ambient temperature with a various rate from 234°C/s to 1°C/min. To investigate the influence of plastic deformation on a shape of CCT (Continuous Cooling Transformations) diagrams, another part of the specimens were 50% deformed at 885°C or 1100°C and cooled to ambient temperature with a rate from 95°C/s to 1°C/min. The DIL 805A/D dilatometer, with a LVDT-type measuring head, was used to carry out dilatometric test.
Findings: Performed dilatometric research revealed that the steel is characterized with A_{c3}=843°C, A_{c1}=707°C and a relatively low M_s temperature equal 370°C. Plastic deformation of steel at the temperature of 885°C prior to the start of phase transformations results in distinct acceleration of pearlitic transformation and slight translation of bainitic transformation towards shorter times.
Research limitations/implications: Elaborated curves of supercooled austenite transformations of studied steel fully predispose it to production of forgings quenched directly from forging finish temperature and successively subjected to high temperature tempering.
Practical implications: The obtained CCT diagrams of supercooled plastically-deformed austenite transformations can be useful in determination of cooling condition of the thermo-mechanical processing for high strength forged machine parts obtained from microalloyed steels.
Originality/value: The diagrams of the plastically-deformed supercooled austenite for a new-developed microalloyed steel were obtained.
Keywords: Microalloyed steel; CCT-diagram; Supercooled austenite; Thermo-mechanical treatment; Forged elements

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1. Introduction

The condition necessary for formation of fine-grained microstructure of steel products is to perform metallurgical processing under conditions assuring fine-grained microstructure of austenite prior to transformation of this phase which occurs during cooling of products from the temperature of hot-working finish. In case of conventional constructional steels, fine-grained microstructure of austenite can be obtained through reduction of hot-working finish temperature, assuring the course of recrystallization of plastically deformed austenite, however preventing grain growth of this phase prior to the beginning of transformation occurring during cooling of products. Taking into consideration that the size of grains of recrystallized austenite is the function of temperature and strain rate, the same size of grains of that phase can be obtained only in case of not very thick plates, when plastic strain is uniformly distributed on their section during rolling. Whereas, in case of complex shape and diversified thickness forgings, plastic strain is not uniformly distributed therefore the grain size of recrystallized austenite is diversified in different areas. This is why forgings made of conventional steels are subjected to normalization in order to obtain grain refinement and unification of their properties and the ones made of alloy steels – subjected to toughening. Normalization is not required in case of forgings made of micro-alloyed steels, produced under properly selected conditions of plastic working, as microadditions introduced into the steel facilitate formation of homogeneous fine-grained microstructure in respect of grain size and prevent grain growth of recrystallized austenite. The presence of microadditions in toughening steels allows to produce forgings using the methods of thermo-mechanical treatment what has an important economical significance [1-8].

Economic considerations determine that the majority of forgings for automotive industry, mining, agricultural and other machines is currently produced of ferritic-pearlitic micro-alloyed steels. Steel designated as 49MnVS3 containing 0.44-0.54%C, up to 0.6%Si, 0.6-1.0%Mn, 0.045-0.065%S and 0.08-0.13%V, characterized with YS > 450 MPa, UTS from 750 to 900 MPa and DVM impact energy of specimens ranging from 15 to 30 J, was the first grade of micro-alloyed steel used for an engine crankshaft in Thyssen Edelstahlwerke [9]. Such high mechanical properties of forged parts can be achieved by appropriate selection of forging conditions, i.e. temperature of charge heating and plastic deformation, since the distribution of strains and strain rate during production of complex shape die forgings is difficult to be adjusted. The conditions of charge heating for forging should not lead to total dissolution of interstitial phases of microadditions introduced into steel in a solid solution for it causes disadvantageous grain growth. Deformation at high rate and short duration intervals for moving the produced part from one die impression to another do not create convenient conditions for the course of static recrystallization, allowing grain refinement of austenite grains. Indeed, $\gamma \rightarrow \alpha$ transformation of both thick- and fine-grained plastically deformed austenite, begins on grain boundaries, twin boundaries and deformation bands, in case of coarse-grained γ phase it doesn't assure sufficiently fine-grained microstructure and expected mechanical properties of forged parts. Forgings produced under such conditions, free-air cooled from the temperature of plastic working finish, admittedly obtain high strength as a result of strong precipitation hardening, but also low crack resistance. An effective way to increase ductility and strength of ferritic-pearlitic steel is to obtain a microstructure consisting of ultrafine excess ferrite and finest areas of pearlite limited with narrow angular boundaries, which are individual colonies or areas enclosing several neighbouring colonies. This can be realized through transformation of austenite with finest grains and decrease of ferritic and pearlitic transformation temperature. The studies on the increase in toughness of micro-alloyed ferritic-pearlitic steels have led to development of grades with decreased concentration of carbon. An example of such a grade is 27MnSiVS6 steel containing 0.25-0.30%C, 1.30-1.60%Mn, 0.5-0.8%Si, 0.030-0.050%S and 0.08-0.13%V. This steel is characterized with YS > 500 MPa, UTS from 800 to 950 MPa and DVM impact energy ranging from 40 to 60 J [10].

Higher mechanical properties, especially crack resistance, compared to forgings with ferritic-pearlitic microstructure, can be obtained for parts of low-alloy toughening steels microalloyed with Ti, Nb and V and N or B forged in dies with the method of thermo-mechanical processing [11-14]. This method consists in plastic deformation of steel under conditions of controlled forging with successive customary or isothermal quenching of forgings directly from the forging finish temperature. Nevertheless, hardening of forgings from the forging finish temperature directly after plastic deformation is done does not assure expected utilizing properties of products, especially those made of alloy steels containing Cr, Mo and V. It's connected with an impact of high density dislocations and precipitation of dispersive particles of carbides on these lattice defects on martensitic transformation in plastically deformed austenite during hardening of manufactured products. Then steel obtains high hardness and brittleness directly after quenching and martensite, which is depleted in carbon and alloying additions, is more susceptible to tempering. It causes a decrease of temperature of phase transformations of alloying carbides occurring during tempering as well as cuts and even decay of secondary hardness. Hence, plastically deformed austenite should be at least 50% recrystallized prior to hardening in order to avoid this disadvantageous impact of high density dislocations and precipitation – with their participation – of dispersive carbides of microadditions introduced into steel. This can be realized through holding of forgings in forging finish temperature for the $t_{0.5}$ time needed for formation of 50% fraction of recrystallized austenite, meanwhile performing trimming, as per example. Direct customary hardening of forgings from forging finish temperature or after the $t_{0.5}$ time limits heat treatment of forged products only to tempering, while isothermal holding of forgings eliminates completely the need of expensive toughening. For example, parts forged from 25GVN steel with microstructure of upper bainite produced with the method of thermo-mechanical treatment, applying the $t_{0.5}$ time and hardening close to isothermal, can possibly obtain $YS_{0.2}$ > 650 MPa, UTS > 900 MPa, impact energy $KV_{20°C}$ > 45 J and hardness values from 280 to 290 HB [15]. Thermo-mechanical treatment with the use of customary hardening of forgings from plastic working finish temperature and successive high-temperature tempering is easier when it's about realization. In this case, steels with microaddition of boron, which increases hardenability and microaddition of titanium which is a shield against its bonding in BN stable nitride, are particularly useful.

Knowledge of diagrams of supercooled austenite transformations is necessary for proper design of conditions of thermo-mechanical treatment and controlled cooling of forgings from the temperature of forging finish, in particular. However, classical CCT diagrams have limited suitability for elaboration of conditions of products cooling from hot-working finish temperature. Diagrams of transformations of supercooled plastically deformed austenite have significant technical suitability. For example, studies of influence of plastic deformation on the course of supercooled austenite transformation curves were performed in [16] on steel containing 0.17%C, 1.37%Mn, 0.26%Si, 0.24%Cr, 0.48%Mo and microadditions of Nb, V, Ti and B in the amount of 0.025%, 0.019%V, 0.004% and 0.002%, respectively. Conducted examinations indicated that plastic deformation of austenite prior to transformation causes considerable acceleration of diffusive transformations, i.e. ferritic and pearlitic transformation and leads to shorter duration of bainitic transformation as well as to slight decrease of M_s temperature for investigated steel.

It was also found in [17, 18] that plastic deformation of austenite prior to the beginning of phase transformations of steel consisting of 0.24%C, 1.55%Mn, 0.87%Si, 0.4%Al, 0.034%Nb and 0.023%Ti caused an increase of the ferritic bay and increase of $\gamma \rightarrow \alpha$ transformation temperature, independently from the cooling rate. Moreover, decrease of bainitic transformation start temperature and clear translation of ferritic transformation to shorter times was observed. Similar issues have been studied in [19-23].

2. Material and methodology

The research was carried out on newly elaborated steel containing 0.28%C, 1.41%Mn, 0.29%Si, 0.008%P, 0.004%S, 0.26%Cr, 0.11%Ni, 0.22%Mo, 0.20%Cu, 0.027%Nb, 0.028%Ti, 0.019%V, 0.003%B and 0.025%Al, assigned for production of forged machine parts with the method of thermo-mechanical treatment.

Steel melt, weighing 100 kg, was done in VSG-100S type laboratory vacuum induction furnace, produced by PVA TePla AG. Casting was performed in atmosphere of argon through heated intermediate ladle to quadratic section cast iron hot-topped ingot mould: top – 160/bottom – 140 mm x 640 mm. In order to obtain 32x160 mm flat bars, initial hot plastic working of ingots was performed, implementing the method of open die forging in high-speed hydraulic press, produced by Kawazoe, applying 300 MN of force. Heating of an ingot to forging was done in a gas forging furnace. The range of forging temperature was equal 1200-900°C, with interoperation reheating in order to prevent the temperature of the material to drop below 900°C.

Evaluation of the influence of hot plastic deformation on phase transformations of supercooled austenite of investigated steel applying continuous cooling of samples was done using dilatometric method. The experiment was performed in the Institute of Ferrous Metallurgy in Gliwice, implementing DIL 805A/D dilatometer, manufactured by Bäehr Thermoanalyse GmbH, equipped with LVDT type measuring head with theoretical resolution equal ±0.057 mm. Heating of specimens in dilatometer was realized with the induction method using a generator at frequency of 250 kHz. Both, heating and isothermal holding of samples at assigned temperature were carried out in $5 \cdot 10^{-4}$ bar vacuum, created by rotary and turbomolecular pump. Temporary temperature deviations from assigned value did not exceed ±1.0°C. Temperature measurement was done using S PtRh10-Pt type thermoelement with diameter of wires equal 0.1 mm. Both thermoelement ends were welded onto samples in the middle of their length.

Examinations and analysis of results were performed using the technique consisting in putting a tangent against a dilatation curve in the vicinity of the start and finish of phase transformation. In case of inseparable transformations (occurring one after another) numerical differentiation of dilatation curves was used for the analysis. In case of ferritic and pearlitic transformations, the method based on linear transformation of analyzed section of dilatation curve was applied in order to determine the phase transformation start and finish temperature.

Basing on performed examinations, critical points of steel (A_{c1}, A_{c3} and M_s) were determined as well as ranges of phase transformation of supercooled austenite in non-deformed state and after plastic deformation at 885°C and 1100°C as well. Investigation of phase transformations of non-deformed austenite was performed on ϕ4xϕ3x7 mm tubular specimens, while studies of phase transformations of plastically deformed austenite were carried out on ϕ4x7 mm solid cylindrical samples. Prior to the experiment, all samples were subjected to thermal stabilization, i.e. they were heated to the temperature of 650°C at the rate of 10°C/s, then held for 600s at this temperature and successively cooled to ambient temperature with the rate of 30°C/min.

In case of determination of phase transformations of supercooled non-deformed austenite, specimens were heated at the rate of 10°C/s up to the temperature of 885°C, being the initiation of controlled cooling. Samples were austenized at this temperature for 600 s and then cooled to ambient temperature at diversified rate, i.e. 234°C/s, 99°C/s, 50°C/s, 20°C/s, 10°C/s, 4°C/s, 2°C/s, 1°C/s, 30°C/min, 15°C/min, 6°C/min, 3°C/min, 1°C/min.

Two diagrams of phase transformations of supercooled plastically deformed austenite have been determined. In the first version, after the samples were heated up to the temperature of 885°C at the rate of 10°C/s, they were austenized for 600 s and plastically deformed in this temperature applying compression. In the second variant, after austenitizing at the temperature of 1200°C for 300s samples were cooled down to the temperature of 1100°C at which plastic deformation took place. In both variants, the value of true strain was equal 0.69 and the strain rate equal 1 s^{-1}. After plastic deformation, specimens were cooled down to ambient temperature at diversified rate, i.e. 95°C/s, 79°C/s, 50°C/s, 20°C/s, 10°C/s, 4°C/s, 2°C/s, 1°C/s, 30°C/min, 15°C/min, 6°C/min, 3°C/min, 1°C/min.

In order to identify microstructure of products of supercooled austenite transformations, after dilatometric studies, samples were subjected to metallographic analysis in NEOPHOT 2 light microscope with digital image recording, at magnification of 400x and 800x. Investigation of microstructure of specimens was carried out on transverse microsections – in case of non-deformed specimens and on longitudinal microsections – in case of plastically deformed samples. HV10 hardness of samples was studied using Vickers method applying the load of 98 N, implementing Swiss Max 300 universal testing machine. There were five measurements performed on each sample.

3. Results and discussion

The diagram of transformations of supercooled austenite of investigated steel and selected microstructures of samples cooled from the temperature of 885°C at the rate ranging from 234°C/s to 1°C/s are shown in Figs. 1-2, while detailed results of the analysis of dilatograms for the examined steel are set together in Table 1. Conducted experiment revealed that studied steel obtained the values of A_{c3} = 843°C, A_{c1} = 707°C and considerably low M_s temperature, equal 370°C. Cooling the samples at a wide range of cooling rates, i.e. from 234 to 50°C/s assures obtaining martensitic microstructure (Fig. 1), however hardness of samples cooled in this range is slightly decreased and is equal 527 HV for the cooling rate of 234°C/s, 512 HV for the cooling rate of 99°C/s and 506 HV - for the cooling rate of 50°C/s. Specimens cooled in analyzed range of cooling rates, i.e. from 234 to 50°C/s, demonstrate microstructure of fine lath martensite (Figs. 2a-c). Decrease of the cooling rate of samples to 20°C/s results in obtaining martensitic-bainitic microstructure (Fig. 2d) with slight portion of bainite (approx. 2%). Such small fraction of this phase in microstructure of steel cooled at the rate of 20°C/s is a result of very short time for realization of bainitic transformation, equal around 6 s. Further decrease of the cooling rate causes appearance of ferrite in steel microstructure. Multiphase microstructure of steel, which consists of martensite, bainite and ferrite, is present in a wide range of the cooling rate, i.e. from 10°C/s to 15°C/min. Estimate portion of individual phases in this range of the cooling rate, determined with dilatometric method, changes as follows: martensite – from 95% to 2%, bainite – from 4% to 95% and ferrite – from 1% to 3%. Hardness of specimens cooled in the analyzed range of cooling rate decreases from 488 to 256 HV. Particular attention should be brought by the fact of dominant fraction of martensite in microstructure, which is equal around 63% at the cooling rate of 2°C/s. Decrease of the cooling rate to 6°C/min results in formation of pearlite in microstructure (Figs. 2f-h). Participation of this phase in steel microstructure increases from 2% to 38% along with a decrease of the cooling rate from 6°C/min to 1°C/min. Steel cooled at the rate of 1°C/min demonstrates fine-grained ferritic-pearlitic microstructure (Fig. 2h) with the value of hardness equal approx. 144 HV.

Plastic deformation of investigated steel in austenitizing temperature (885°C) prior to the beginning of controlled cooling resulted in slight displacement of interfacial boundaries and temperature-time areas of individual phase transformations of supercooled austenite (Fig. 3) in respect to boundaries of phase transformations of supercooled non-deformed austenite towards the direction of short time. Plastic deformation of austenite prior to transformation causes distinct acceleration of pearlitic transformation and slight displacement of bainitic transformation to short time. Increased diffusion rate in plastically deformed austenite and high density of areas suitable for heterogeneous nucleation of products of diffusive transformations of the phase, i.e. deformation and shearing bands with high density dislocations and piling-up of dislocations in front of grain boundaries are the factors which decide about displacement of phase transformations boundaries of supercooled austenite. Whereas, no significant influence of plastic strain at the temperature of 885°C on ferritic transformation has been found. No displacement of the phase towards shorter time in respect to limits of this phase transformation of non-deformed supercooled austenite was noted. In addition, $\gamma \rightarrow \alpha$ transformation start temperature did not increase. Determined M_s temperature of plastically deformed austenite is equal 356°C and is lower than martensite start temperature of non-deformed γ phase. High density of dislocations, caused by plastic deformation of input phase prior to the start of transformation, makes the movement of phase boundaries and growth of martensite crystals difficult. These lattice defects are intrinsic obstacles for migration of phase boundaries and cause disordering of a proper crystal structure. Hence, high density of dislocations in plastically deformed austenite caused the decrease of $\gamma \rightarrow \alpha'$ martensite start temperature. Detailed results of analysis of dilatograms of supercooled plastically deformed austenite together with the results of hardness test for studied steel are presented in Table 2.

Obtaining fully martensitic microstructure (Fig. 4a) requires application of the cooling rate of 95°C/s. Hardness of sample cooled at that rate is equal 529 HV. Reducing the cooling rate to 50°C/s, 20°C/s and 10°C/s results in appearance of bainite in steel microstructure (Figs. 4b-d), yet the portion of this phase for indicated cooling rates is minor and does not exceed 2%. Apart from martensite and bainite, small quantities of ferrite can be found in the microstructure in a range of the cooling rate from 6°C/s to 30°C/min. Similarly as in case of the diagram of phase transformations of non-deformed supercooled austenite, also in case of deformation of γ phase prior to controlled cooling, martensite is a dominant phase in a wide range of cooling rates. Percentage fraction of this phase changes from 100% for the cooling rate of 95°C/s to 3% – for the cooling rate of 15°C/min. Along with reduction of the cooling rate, the portion of bainite in microstructure increases distinctly. Its maximum, i.e. 94%, is noted after cooling the steel at the rate of 15°C/min. Pearlite will appear in the microstructure once the steel is cooled at this cooling rate. Decrease of the cooling rate to 6°C/min and 3°C/min results in increase of pearlite fraction (Figs. 4f-g). Cooling the steel at the rate of 1°C/min assures formation of ferritic-pearlitic microstructure with small fraction of bainite (Fig. 4h). Plastic deformation of γ phase prior to controlled cooling resulted also in grain refinement of microstructure, what is confirmed by higher values of hardness for specimens cooled at the same cooling rates in respect to hardness obtained for non-deformed specimens.

The diagram of transformations of supercooled austenite plastically deformed at the temperature of 1100°C and microstructures of samples cooled from the temperature at the rate in a range from 79°C/s to 1°C/min is presented in Figs. 5-6, while detailed results of analysis of dilatograms prepared for investigated steel are set together in Table 3.

Determined martensite start temperature is equal 347°C and is slightly lower than M_s temperature of austenite plastically deformed at the temperature of 885°C and distinctly lower than M_s temperature noted for non-deformed austenite. Obtaining martensitic microstructure (Fig. 6a) requires application of the cooling rate equal 79°C/s. Hardness of sample cooled at this rate is equal 516 HV and is clearly lower than hardness of sample cooled at similar cooling rate after plastic deformation at the temperature of 885°C.

Cooling the samples in a wide range of cooling rates, i.e. from 50°C/s to 1°C/s after plastic deformation at the temperature of 1100°C guarantees obtaining martensitic-bainitic microstructure.

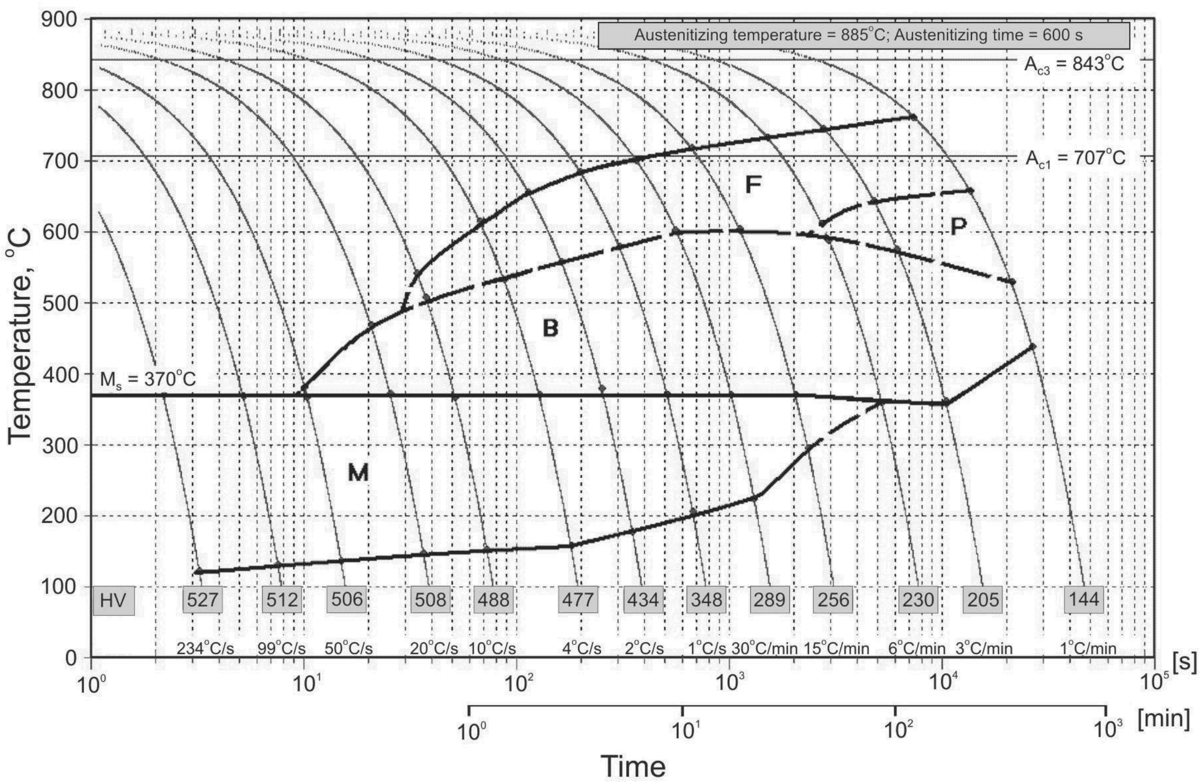

Fig. 1. Diagram of supercooled austenite transformations of investigated steel

Table 1.
Results of the analysis of dilatograms of supercooled austenite transformations and results of the hardness test for studied steel specimens

Cooling rate	Hardness HV10	Start and finish temperature of transformation, °C							
		M_s	M_f	B_s	B_f	P_s	P_f	F_s	F_f
234°C/s	527	370	122						
99°C/s	512	368	131						
50°C/s	506	366	136	380	366				
20°C/s	508	372	146	468	372				
10°C/s	488	366	152	507	366			541	507
4°C/s	477	369	158	533	369			614	533
2°C/s	434	378	177	557	378			655	557
1°C/s	348	369	205	578	369			684	578
30°C/min	289	369	223	602	369			700	602
15°C/min	256	370	295	604	370			717	604
6°C/min	230			590	359	610	590	733	610
3°C/min	205			575	360	642	575	745	642
1°C/min	144			528	437	657	528	761	657

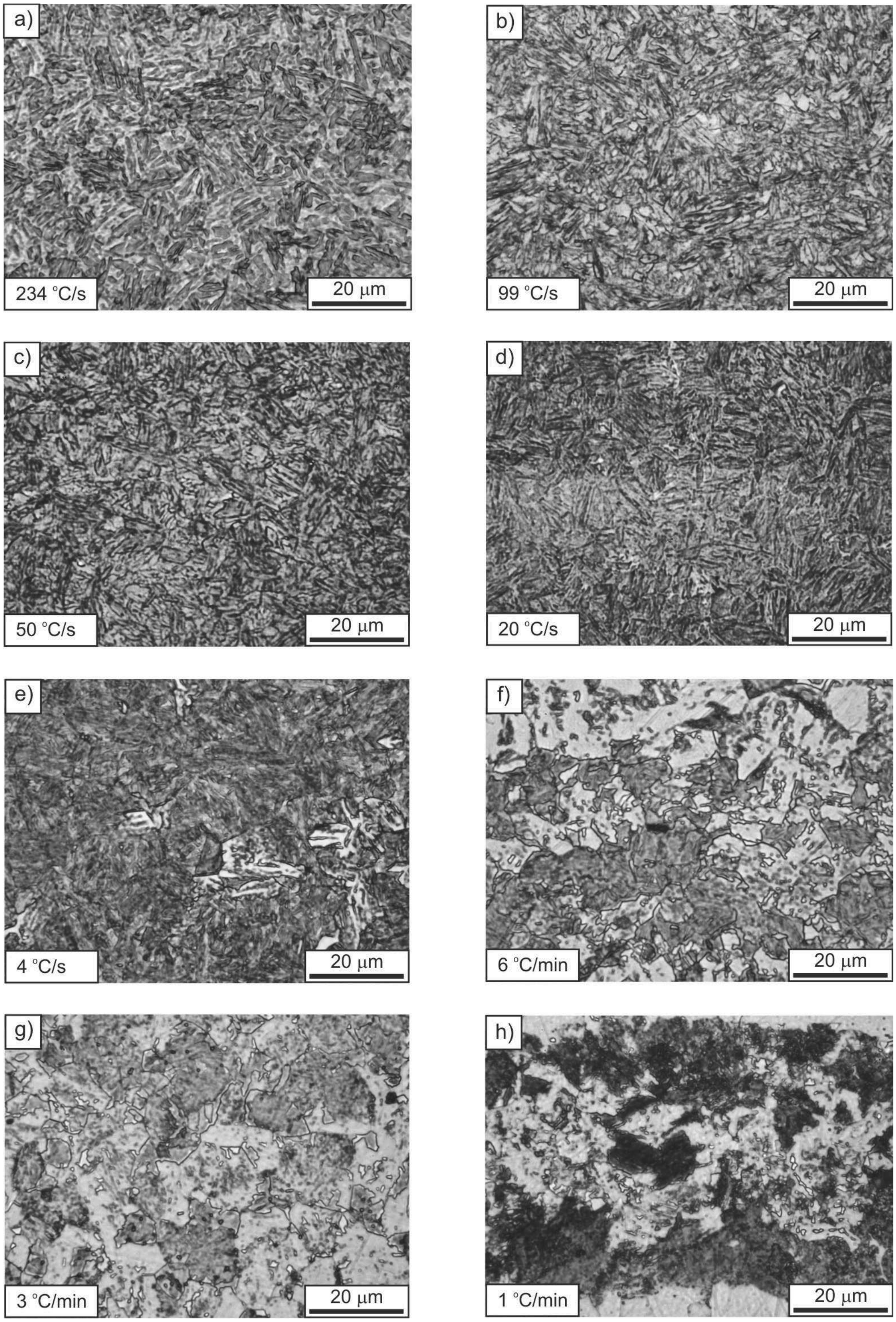

Fig. 2. Structures obtained after cooling the specimens from the austenitizing temperature of 885°C with a rate: a) 234°C/s, b) 99°C/s, c) 50°C/s, d) 20°C/s, e) 4°C/s, f) 6°C/min, g) 3°C/min, h) 1°C/min

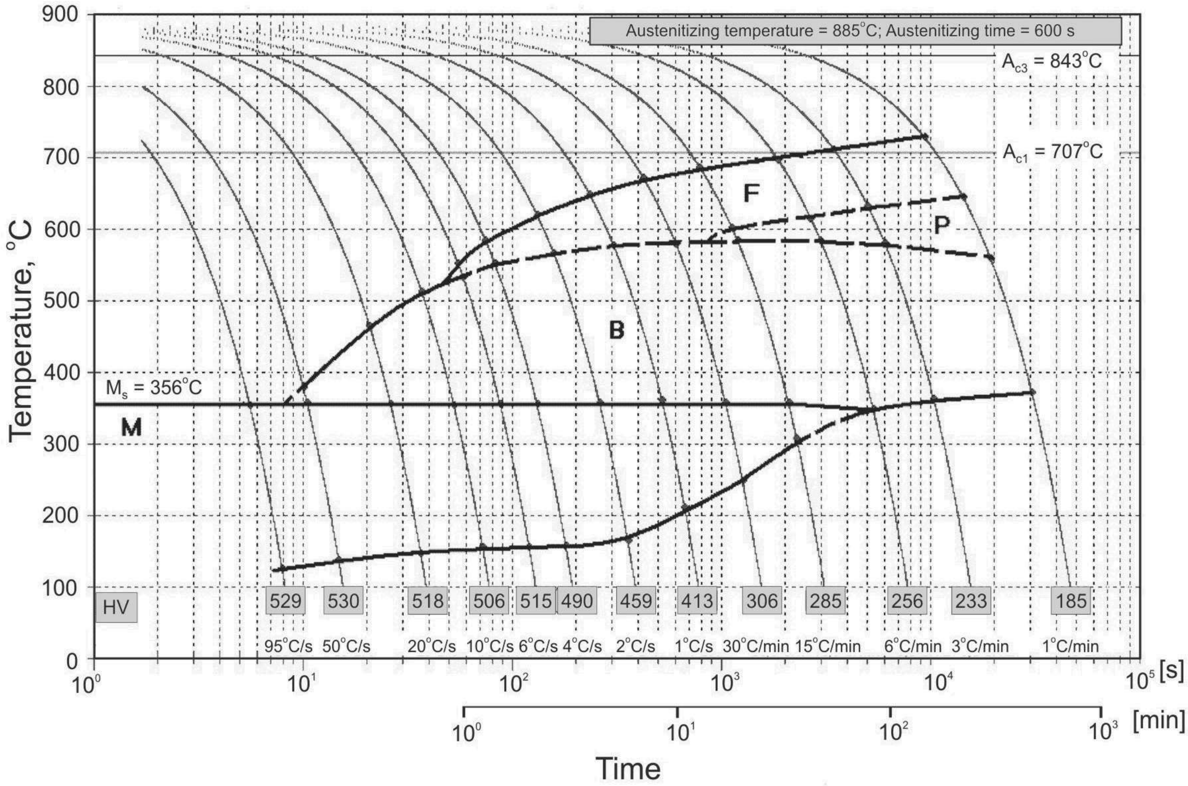

Fig. 3. Diagram of supercooled plastically-deformed austenite transformations of investigated steel; deformation temperature: 885°C

Table 2.
Results of the analysis of dilatograms of supercooled austenite transformations plastically deformed at the temperature of 885°C and results of the hardness test for studied steel specimens

Cooling rate	Hardness HV10	Start and finish temperature of transformation, °C							
		M_s	M_f	B_s	B_f	P_s	P_f	F_s	F_f
95°C/s	529	356	125						
50°C/s	530	357	138	379	357				
20°C/s	518	356	147	465	356				
10°C/s	506	353	156	512	353				
6°C/s	515	355	156	533	355			551	533
4°C/s	490	355	157	551	355			584	551
2°C/s	459	357	164	566	357			620	566
1°C/s	413	360	211	576	360			648	576
30°C/min	306	358	249	579	358			672	579
15°C/min	285	357	307	584	357	601	584	686	601
6°C/min	256			582	348	615	582	697	615
3°C/min	233			580	363	632	580	711	632
1°C/min	185			559	371	644	559	728	644

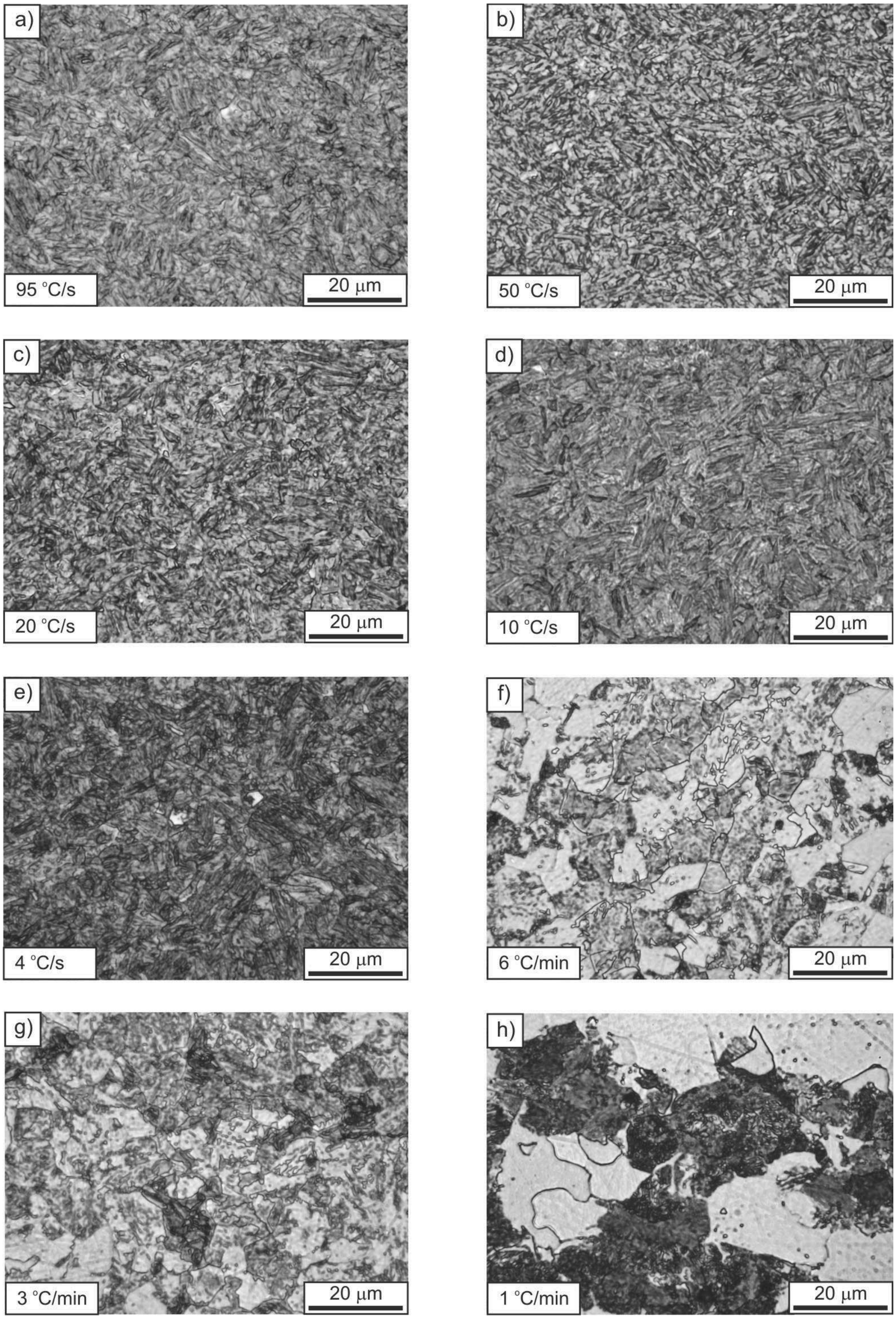

Fig. 4. Structures obtained after cooling the specimens from the deformation temperature of 885°C with a rate: a) 95°C/s, b) 50°C/s, c) 20°C/s, d) 10°C/s, e) 4°C/s, f) 6°C/min, g) 3°C/min, h) 1°C/min

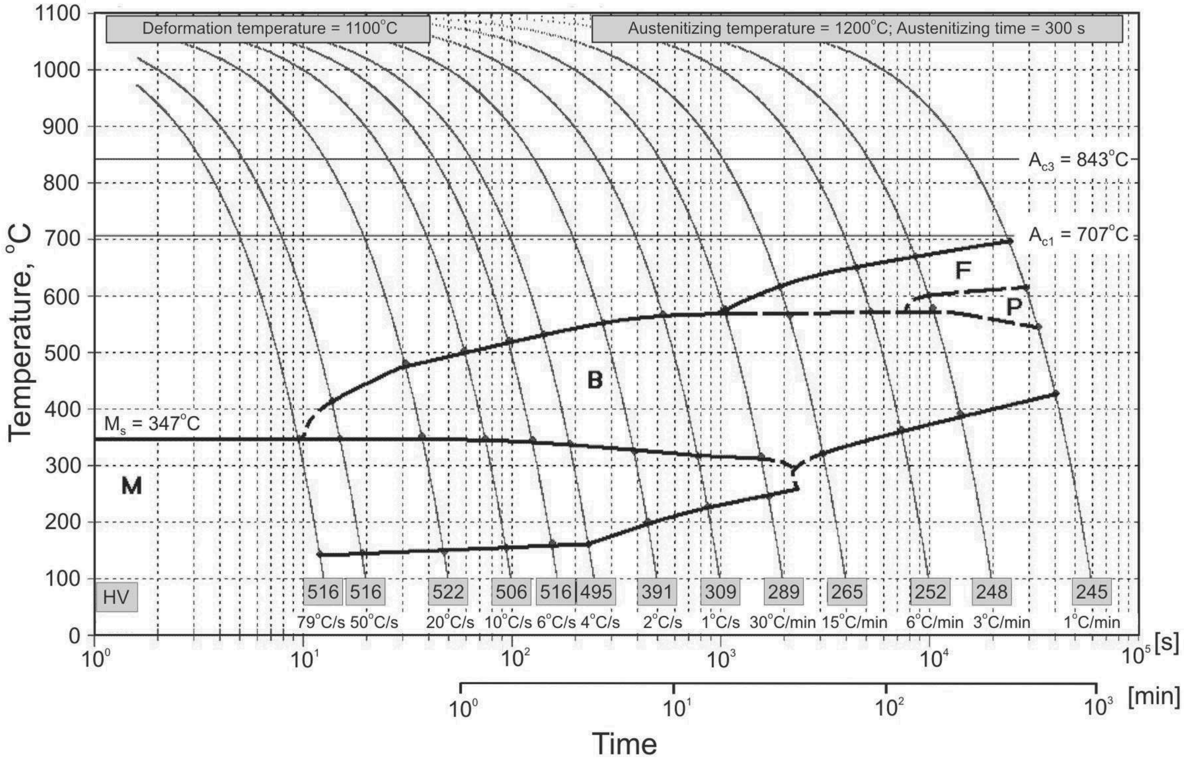

Fig. 5. Diagram of supercooled plastically-deformed austenite transformations of investigated steel; deformation temperature: 1100°C

Table 3.
Results of the analysis of dilatograms of supercooled austenite transformations plastically deformed at the temperature of 1100°C and results of the hardness test for studied steel specimens

Cooling rate	Hardness HV10	Start and finish temperature of transformation, °C							
		M_s	M_f	B_s	B_f	P_s	P_f	F_s	F_f
79°C/s	516	347	143						
50°C/s	516	347	144	413	346				
20°C/s	522	350	147	480	350				
10°C/s	506	346	153	502	346				
6°C/s	516	345	162	519	345				
4°C/s	495	337	160	531	337				
2°C/s	391	324	200	552	324				
1°C/s	309	316	226	566	316				
30°C/min	289	315	245	568	315			575	568
15°C/min	265			565	356			615	563
6°C/min	252			572	363			650	572
3°C/min	248			578	391	599	578	670	599
1°C/min	245			545	426	614	537	696	614

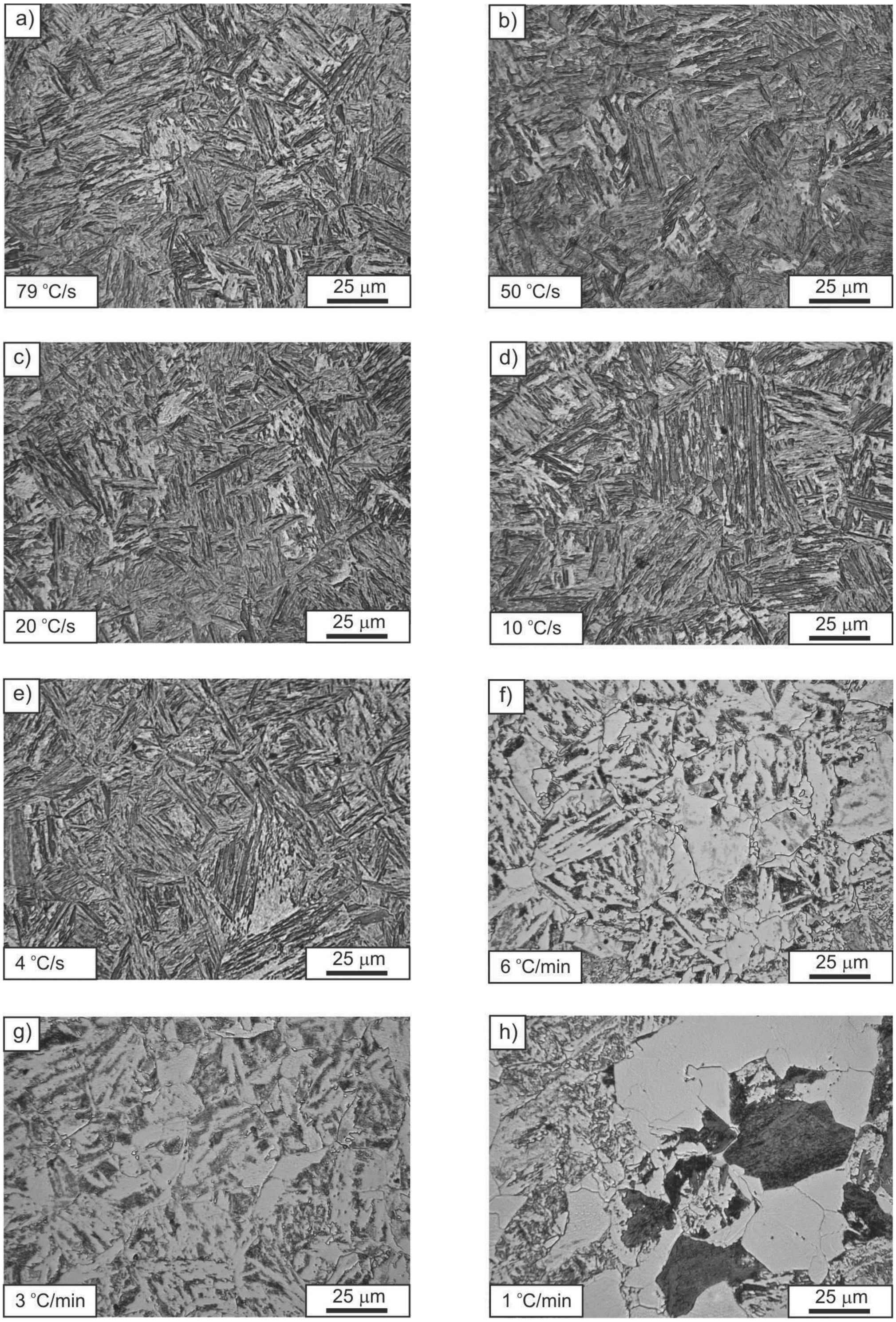

Fig. 6. Structures obtained after cooling the specimens from the deformation temperature of 1100°C with a rate: a) 79°C/s, b) 50°C/s, c) 20°C/s, d) 10°C/s, e) 4°C/s, f) 6°C/min, g) 3°C/min, h) 1°C/min

Estimated portion of these phases in mentioned range of cooling rates determined with the use of dilatometric method changes from 85% to 5% – for martensite and from 15% to 95% – for bainite, yet hardness of specimens decreases from 516 to 309 HV (Table 3). Reducing the cooling rate to 30°C/min results in appearance of ferrite in microstructure, but its fraction is minor and equal about 3%. In a range of the cooling rate from 15°C/min to 6°C/min, the microstructure of studied steel is bainitic-ferritic with dominant portion of bainite. Ferrite, bainite as well as pearlite in estimate amounts of these phases equal 75%, 20% and 5%, respectively, are present in microstructure of investigated steel cooled at the rate of 1°C/min after plastic deformation at the temperature of 1100°C. Hardness of the sample cooled at that rate is equal 245 HV.

Comparing diagrams of transformations of supercooled austenite plastically deformed at the temperature of 885 and 1100°C, no substantial differences regarding temperature-time areas of martensitic and bainitic transformations were found. It's significant that in a very wide range of cooling rates, especially in case of plastic deformation realized at the temperature of 1100°C, steel demonstrates martensitic-bainitic microstructure.

What can be noted when analysing the diagram of transformations of supercooled austenite of steel plastically deformed at the temperature of 885°C (Fig. 3) is that there is a very distinct translation of ferritic transformation bay in the direction of shorter time in respect to the limits of this phase transformation of austenite plastically deformed at the temperature of 1100°C. Moreover, it was found that increase of deformation temperature results in distinct decrease of $\gamma \rightarrow \alpha$ transformation start temperature. As per example, for the cooling rate of 30°C/s the ferritic transformation start temperature is equal 672°C and 575°C – for cooling at this rate after plastic deformation realized at the temperature of 885°C and 1100°C, respectively. The difference regarding $\gamma \rightarrow \alpha$ transformation start for both cases decreases along with decrease of the cooling rate. Reduction of plastic deformation temperature resulted also in apparent acceleration of pearlitic transformation. No significant differences regarding hardness of samples deformed in different temperatures, successively cooled at the same rate, have been observed.

Plastic deformation of steel at the temperature of 885°C prior to the start of phase transformations slightly changed the form of the diagram of supercooled austenite transformations. Determined M_s temperature of plastically deformed austenite is equal 356°C and is lower than martensite start temperature of non-deformed γ phase. Plastic deformation of austenite prior to the transformation results in distinct acceleration of pearlitic transformation and slight translation of bainitic transformation towards shorter times. No significant influence of plastic strain on ferritic transformation has been found. Additionally, it was found that samples plastically deformed at the temperature of 885°C prior to their controlled cooling demonstrate higher hardness values in respect to hardness values of non-deformed samples cooled at the same rate. It's a result of grain-refinement of microstructure in the whole range of the cooling rate.

Very clear translation of the ferritic bay towards right after plastic deformation at the temperature of 1100°C in respect to its position on the CCT diagram of supercooled non-deformed austenite is a result of increased austenitizing temperature equal 1200°C.

Elaborated curves of supercooled austenite transformations of studied steel fully predispose it to production of forgings quenched directly from forging finish temperature and successively subjected to high temperature tempering.

Determined diagrams of transformations of supercooled non-deformed and plastically deformed austenite will be the basis for elaboration of conditions of thermo-mechanical treatment for investigated steel forgings.

4. Conclusions

Performed investigations allowed to evaluate the influence of plastic deformation and cooling rate on the course of curves of supercooled austenite transformations of newly elaborated steel. Performed dilatometric research revealed that the steel is characterized with A_{c3}=843°C, A_{c1}=707°C and relatively low M_s temperature equal 370°C. The course of CCT curves of supercooled austenite transformations indicates that microstructure of the steel is martensitic in a wide range of cooling rates. Even after cooling the steel at relatively low rate, i.e. 2°C/s, the fraction of α' phase in microstructure is equal over 60%. It indicates that the steel possesses high hardenability, guaranteed by microaddition of boron and its shield against formation of BN in the form of titanium microaddition. Boron microaddition, introduced into steel in the amount of 0.003%, dissolved in a solid solution, causes a decrease of energy of these lattice defects, delays nucleation during $\gamma \rightarrow \alpha$ transformation and decreases the critical cooling rate while segregating on austenite grain boundaries.

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