

Metallurgical products of microalloy constructional steels

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Materials

ABSTRACT

Purpose: The aim of the paper is a monographic presentation, based on investigations connected with the elaboration of new kinds of microalloy steels and a technique of heat and thermo-mechanical treatment in the production of a selected group of metallurgical products.

Design/methodology/approach: The influence of the chemical composition and metallurgical purity on the homogenisation of austenite was analysed, as well as thermally activated processes and phase transformations in the course of deformation and controlled cooling conditions, concerning microadditives Nb, V, Ti, B and metallurgical products.

Findings: The chemical composition of the obtained constructional steels with microadditives was quoted, as well as the technological conditions of forging and rolled products required for industrial purposes.

Research limitations/implications: The results of investigations concerning the structure and mechanical properties have been presented, as well as the resistance to fracture, mainly of thick plates and hand-forged and drop-forged products and also elements of sheet structures for the automotive industry, made of AHSS steels.

Practical implications: The results of the author's own investigations concerning microalloy steels indicate the actual possibility of their application in practice in many steelworks, particularly in plants producing semi products and final metallurgical products, among others in hot rolling plants and forging shops, first of all in machine building and automotive industry.

Originality/value: The application of modern metallurgical technologies and metalforming has been suggested for selected metallurgical products of microalloy steels of the type HSLA with a ferrite-bainite, bainite or tempered martensite structure and also of HSS steels and UHSS steel with a wide range of mechanical properties and technological formability, which is essential in the case of products of the automotive industry.

Keywords: Microalloyed constructional steels; Mechanical properties; Heat treatment; Thermomechanical treatment

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1. Introduction

Microalloyed steels belong to the group of structural materials susceptible to attaining the required usable properties at relatively low costs of production. The possibility of controlling the

influence of the conditions of production on their structure decides about their application as fundamental material for the construction of various structures, machines and devices operating at varying temperature and in various stress conditions.

The industrial development has made it necessary to elaborate new techniques in metallurgy and processing, permitting to get

products of the new generation of constructional weldable steels with a high strength and warranted resistance to fracture under complex load conditions, particularly at low temperature.

Microadditives introduced into the steel permit at adequately chosen conditions of hot plastic treatment to obtain a fine-grained structure of recrystallized austenite, which in the course of controlled cooling undergoes further transformations, resulting in fine-grained products. These structures contribute considerably to the strengthening of metallurgical products and to an increase of their ductility, accompanied by a drop of brittle fracture transition temperature.

As essential group of products obtained from these steels are thick plates and forged machine elements with a yield point exceeding 800 MPa. High strength plates are made of microalloy steels with a medium content of carbon containing additionally such elements as Mn, Nb, Ti, V and N, B, whose high values of the yield point, strength and hardness HRC, obtained in the process of quenching and tempering or thermo-mechanical treatment. Both high-strength quenched and tempered plates as well as forged elements of machines are modern engineering materials, applied in the production of numerous responsible structures and machine elements and devices, operating under dynamical loads and varying climatic conditions.

The technology of plates and forged products of microalloyed constructional steel are still dealt with in scientific investigations at many research centers all over Europe and the World. This problem has also been developed for many years at the Institute of Engineering Materials and Biomaterials, Silesian University of Technology in Gliwice, particularly at the Division of Constructional and Special Materials Engineering, until September 2006 headed by the late Professor Jan Adamczyk. This publication is devoted to the memory of Professor Jan Adamczyk by the co-authors and co-workers of this Institute.

The aim of the paper is a monographic presentation, based on investigations connected with the elaboration of new kinds of microalloyed steels and the technique of heat and thermo-mechanical treatment in the production of a selected group of metallurgical products.

2. Characteristics of constructional microalloyed steels

Microalloyed steels were developed together with weldable HSLA (High Strength Low Alloy) steels, containing up to about 0.3% C and about 2% Mn. Manganese is in such steels not treated as a component of the alloy. Besides deoxidation and binding with sulphur to MnS, it involves a drop of the transition temperature of the brittleness of the steel. It also affects an increase of the hardenability of the toughening steels. At a concentration of more than 1.5% in the absence of other elements forming stable interstitial phase with N and C, leading to its tendency of coarsening the austenite grains in the course of heat treatment.

Unexpected possibilities of utilizing the fine-grain structure for the purpose of increasing the yield point in compliance with Hall-Petch's equation, accompanied by a drop of the transition temperature of the steel into the brittle state of and an improvement

of its ductility, turned up in the middle of the previous century, when slight amounts of niobium was introduced into the steel which was to be rolled into strips. Further investigations resulted in the development of qualitatively new constructional microalloyed steels of the type HSLA displaying a high strength containing microadditives very similar to C and N, i.e. Nb, Ti and V in an amount of about 0,1%. Interacting with C and N, metallic elements form stable fine-grained interstitial phases MX with a regular lattice of the type NaCl, permitting to obtain a stable fine-grain austenite structure after its hot plastic working under adequately chosen conditions of deformation, i.e. temperature and rate of deformation.

The discovery of abounding resources of Nb ores after 1945 in Canada and Brasil resulted in a violent increase of interest in its affecting the structure and properties of steel. Since then HSLA steel has been developing, simulated by various factors, as, for instance, the crud-oil crisis and the increasing pressure on environment protection, as well as the development of new techniques in metallurgy, metal processing and welding (Fig. 1).

Due to the technical and economical aspects, resulting from the production of metallurgical products of such steels, and also their adaptability in the construction of various structures, exploited in extremely difficult climatic conditions, this group of materials and the improvement of their technology have gained much interest among scientists, which is proved by numerous publications and regularly organized conferences [1-6], as well as monographies [7-9].

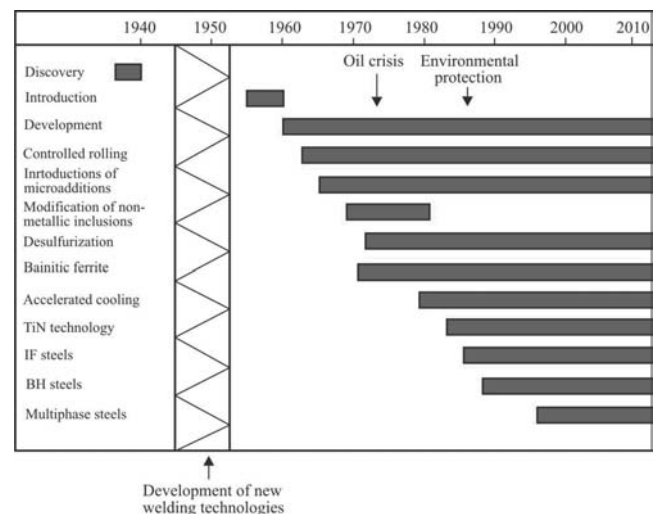


Fig. 1. Schematic diagram of the development of HSLA-steels from 1940 to 2012 [7]

Among the transition elements of the groups IVb and Vb in the periodic table, Nb, Ti and V treated as microadditives to constructional steels, containing sometimes also N or B. Besides their affinity with nitrogen and carbon, metallic microadditives also display a strong chemical activity in the presence of oxygen and sulphur, particularly however Ti and Zr, and in the case of boron also in presence of nitrogen (Fig. 2). This means that liquid steel should be introduced previous to the introduction of microadditives be adequately deoxidized and desulphurized. Otherwise, the

microadditives take over the function of deoxidants and pass as oxides into the slag, and some part of them remains in the steel after skulling in the form of non-metallic inclusions. Titanium, on the other hand, if not adequately desulphurized in the metallic bath, is bound in TiS and (Mn,Ti)S, the TiS remaining in the steel after skulling reacts at the temperature of about 1200°C, partially undergoing a transition into $Ti_4C_2S_2$.

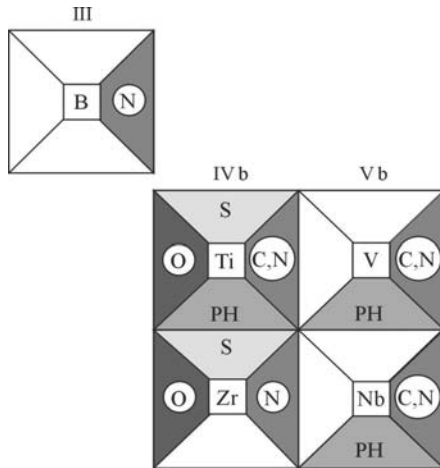


Fig. 2. Chemical affinity of Ti, Zr, V and Nb in relation to O, S, C and N (decreasing with the blackening degree), as well as of B in relation to N and the potential influence of metallic microadditions on the precipitation hardening of steel (PH) [6]

2.1. Metallurgical process and casting

The microadditives Nb, Ti, and V introduced into the steel become under adequately selected conditions of hot plastic forming dispersive particles of interstitial phases MX (M-Nb, Ti and V; X-N and C), restricting the growth of the grains of austenite crystallized in the course of metallurgical processes. The aforesaid microadditives, and also the sometimes applied Zr and B at a concentration of up to 0.005% display a considerable affinity with O, N, Ti, Zr and also with sulphur. In order to deoxidize and desulphurize the steel properly after its dephosphorization in a converter-oxygen process, it is subjected to a finishing treatment (secondary metallurgy) in a ladle-furnace (Fig. 3), in conditions which warrant energetic chemical reactions with a share of alkaline slag deposited on the liquid metal and pulverised substances blown into the bath, as well as alloy components and microadditives, for the purpose of deoxidation, desulphurization and modification of non-metallic intrusions. Previous to their introduction to the bath these substances must be calcinated (roasted) in order to remove moisture and crystalline water. Otherwise, there may occur an excessive enrichment of hydrogen in the metal bath, resulting in a hydrogen killed steel.

In the ladle furnace process the bath is deoxidized by means of Fe-Si, Fe-Mn and Al; and desulphurized by means of slag with a high concentration of CaO and a share of up to 30% Al_2O_3 and 5% CaF_2 and pulverized Ca, CaO, CaS, or, rarely, $CaCO_3$ and CaC_2 , blown into the bath. Alloy components and microadditives

were introduced in batches which permitted to solve them in liquid metal, and the chemical composition was refined and homogenized by blowing a flux of Ar or (rarely) N_2 through the bath. The steel was subjected to vacuum circulation degassing RH or vacuum batch degassing RD (Fig. 4); finally the liquid metal was adapted to the assumed conditions of casting.

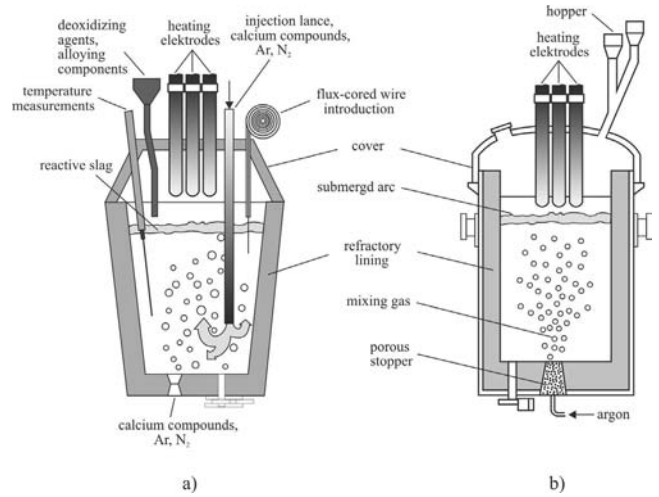


Fig. 3. Schematic drawing of a ladle furnace: a) with an injection lance supplied from the top, b) supplied with neutral gas through a porous stopper in the bottom of the ladle [10]

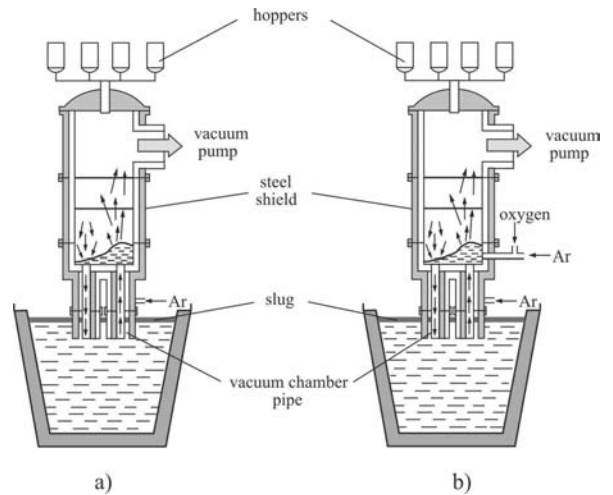


Fig. 4. Schematic drawing of the vacuum degassing process of steel in the RH device (a) and RH-OB device (b) [10]

After ladle furnace treatment and vacuum degassing the liquid metal is cast continually in argon atmosphere. The ladle furnace containing the liquid metal is transported to the station of continuous casting, consisting of an intermediate ladle, in which a constant metalostatic pressure is kept, a copper crystallizer with double walls, between which cooling water is flowing, a set of guide rolls, a part of which is provided with electromagnetic mixers, restricting the growth of columnar grains and the

formation of an axial segregation in the ingot, rolls bending and straightening the ingot, and shears (Fig. 5).

A continuous ingot solidifies at a given length of the crystallizer, and the deficiency of liquid metal in its core, due to the shrinkage of metal in the course of crystallization, is replenished from the intermediate ladle under metalostatic pressure. In newer solutions some part of the rollers in the leading system satisfies also the function of rollers burnishing the ingot with a liquid core. This permits to change considerably the cross-section and restricts the formation of a contraction cavity and shrinkage porosities in the ingot. The cross-section of the ingot must be properly matched to the cross-section of the ready product, taking into account the required plastic treatment. Having been cut in its hot state and additionally reheated to the required temperature, the ingot is subjected to plastic hot treatment.

2.2. Hot plastic working

The influence of microadditives on the formation of fine-grained structure of the steel requires some knowledge concerning their share in the formation of interstitial phases MX and their stability while the steel is being plastically hot-treated. The microadditive M dissolved in austenite binds at a given temperature T with the metalloid X forming the interstitial phase MX with a regular lattice of the type NaCl, in compliance with the reaction:



where: [M] and [X] - restrictive shares of the microadditive M (Nb, Ti, V, Zr or B) and the metalloid X (N or C) dissolved in the solid solution γ at the temperature T, K.

The solubility of the phase MX in the solid solution, taking into account the energy of activation of their formation (solution) is expressed by the reaction:

$$\log[M][X] = -Q/2,303RT + A'/2,303, \quad (2)$$

where: Q - activation energy of the phases MX, R - gas constant, T - temperature, K, A' - constant depending on the kind of phase.

Equation (2) is usually presented as:

$$\log[M][X] = B - A/T, \quad (3)$$

which permits to calculate the temperature at the beginning and the end of the precipitation of the phase MX at a given concentration of the microadditive M into the steel. The value of the constants A and B in equation (3), concerning selected phase MX have been gathered in Table 1.

The knowledge of temperature at the beginning and end of the precipitation of the phase MX in the austenite is of much importance in designing the conditions of the hot plastic working of microalloy steels. The effect of dispersive particles of the phases MX on the formation of a fine-grained structure can be explained as due to changes of the structure of steel in the course of hot plastic working, e.g. multi-pass rolling at elevated temperature. of recrystallization T_R of austenite. The charge heated up to the temperature at the beginning of the plastic treatment displays a coarse-grained structure. In the course of rolling with a sufficiently high ratio of deformation (about 15%) the austenite grains undergo a plastic deformation and elongate in the direction of the flow of metal (Fig. 6), simultaneously being dynamically recovered, which involves a decrease of the flow stress.

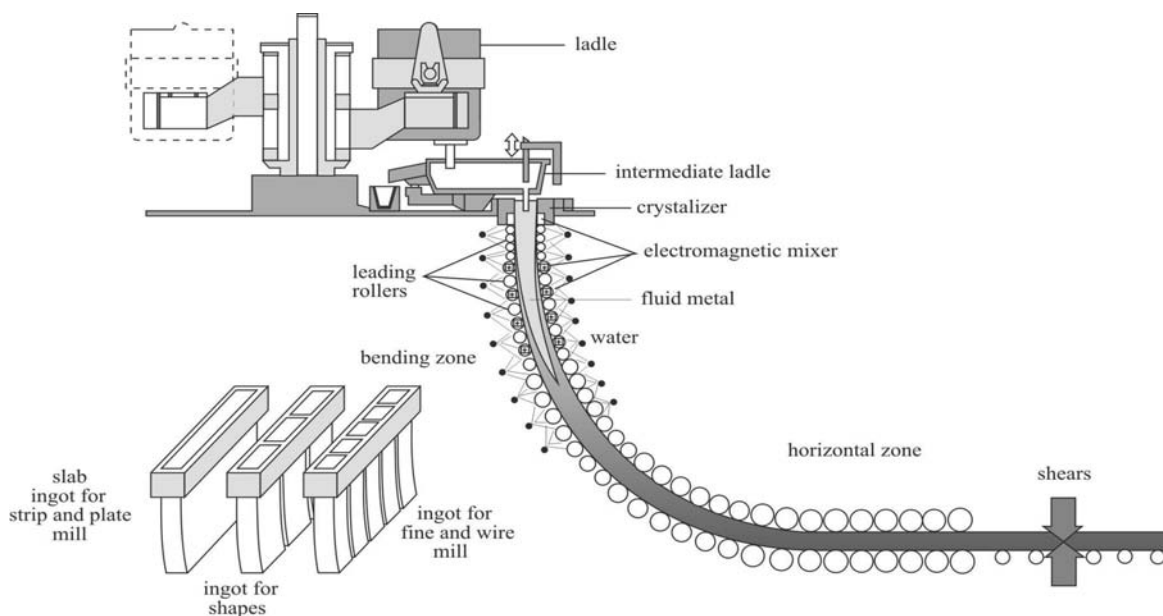


Fig. 5. Technological equipment for continuous casting of steel [9]

Table 1. Values of the constants A and B in the equation (3) for selected carbides and nitrides [11,12]

MX const	AlN	VC	VN	TiC	TiN	NbC	NbN
A	7184	9500	7840	10745	8000	7290	8500
B	1.79	6.72	3.02	5.33	0.32	3.04	2.80

After rolling in the plastically deformed metal occurs its static recovery and static recrystallization leading to a nearly absolute decay of consolidation and formation of a fine-grained structure of the phase γ which is stable due to the large surface of the grain boundaries, and its unfavourable relation to the volume of the grains is metastable. Therefore, after the recrystallization the austenite grains grow unfavourably. In the case of microalloy steels this process can be stopped by the dispersive particles of the phase MX, arising in the austenite during the plastic deformation, restricting the mobility of the grain boundaries. Therefore, after multi-pass rolling of the charge of microalloy steels the size of austenite grains is considerably smaller than in the case of steel without any microadditives (Fig. 7). The dispersive particles of the phase MX not only restrict the mobility of grain boundaries, but also result in their precipitation hardening of steel.

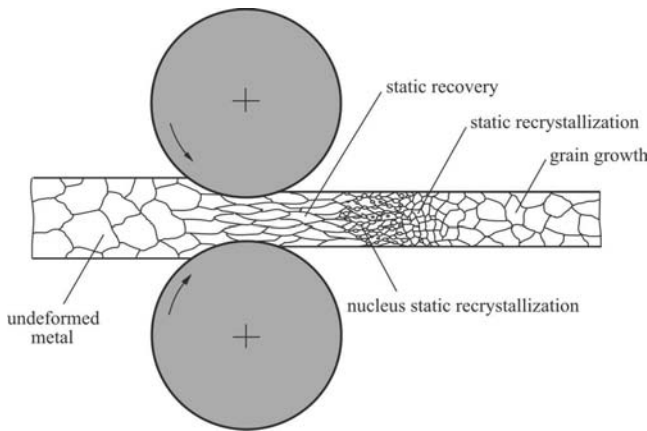


Fig. 6. Schematic diagram of structure changes of deformed metal in the course of rolling and after its completion with a strain higher than ϵ_{cs} [9]

The phases MX of various metallic microadditives differ in their temperature stability and effect on the properties of the steels. The optimal concentrations of Nb in low-carbon steel amounts to 0.04% and permits to increase the yield point of products attained in adequately chosen conditions of plastic working by about 120 MPa thanks to the grain refinement, and about 160 MPa thanks precipitation hardening (by the dispersive particles of NbC, whereby the impact transition temperature is simultaneously set down by nearly -40°C (Fig. 8). A lower influence on the strengthening of the products is affected by Ti at an optimal concentration of 0.08% because the phases MX of this metal, i.e. TiN, may be formed to some part already in the course of the crystallization of the steel, and the in the solid state, after

which Ti(C,N) is precipitate, and after the exhaustion of nitrogen TiC at a considerably higher temperature than NbC. With the drop of temperature, these phases coagulate, due to which their influence on the grain refining and precipitation hardening is reduced, whereas the phases VN, particularly however VC, begin to precipitate in the vicinity of the temperature A_{r3} of steel and therefore their influence on grain refining is inconsiderable, though essential for the precipitation hardening of the products.

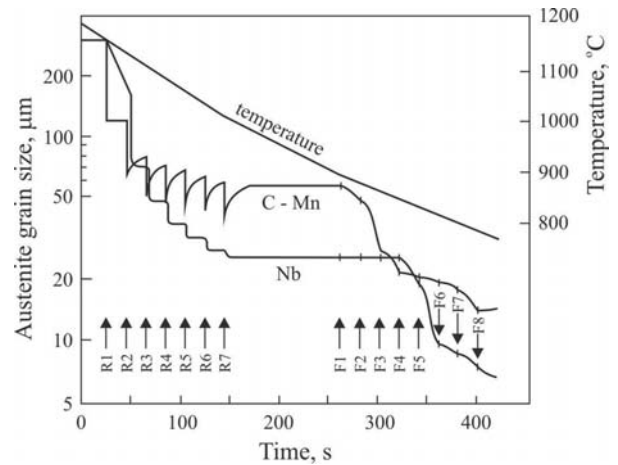


Fig. 7. Predicted changes of austenite grain size in the course of rolling 200 mm thick slab into 20mm thick plates of C-Mn steel and such a steel containing 0.04% Nb [13]; R₁ to R₇ and F₁ to F₈ - subsequent passes during roughing and finishing rolling, respectively

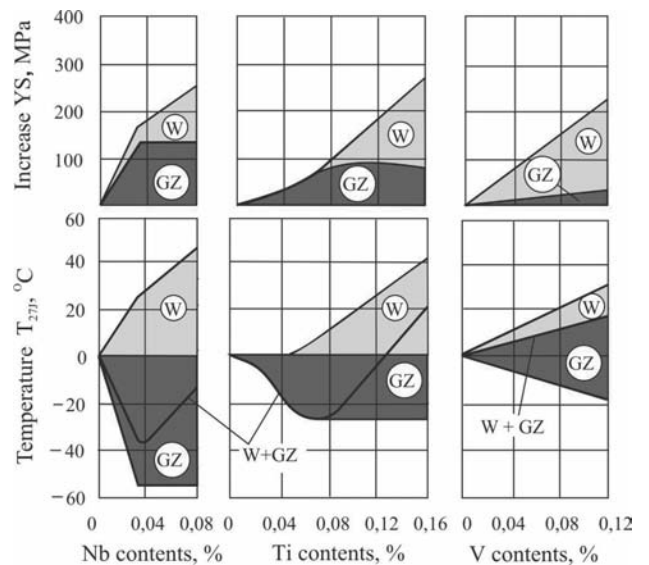


Fig. 8. Influence of the Nb, Ti and V content on the increase of the field point and variation of the impact transition temperature of low-carbon steel [14]: GZ - influence of grain refining, W - influence of precipitation hardening

The temperature range of precipitation of the phases MX depends on the concentration of C and N in the steel and on the applied microadditive. Thus for instance, at the concentration of 0.1% V in the steel containing 0.2% C and 0.02% N, the temperature at the beginning of the precipitation of VN in the austenite is nearly 1100°C, and at the end less than 900°C (Fig. 9a). The knowledge of the temperature range of the precipitation of various phases MX is of essential importance in designing the conditions of the hot plastic working of steel. In order to increase the temperature range of the precipitation of the phases MX several microadditives are introduced into the steel (Fig. 9b).

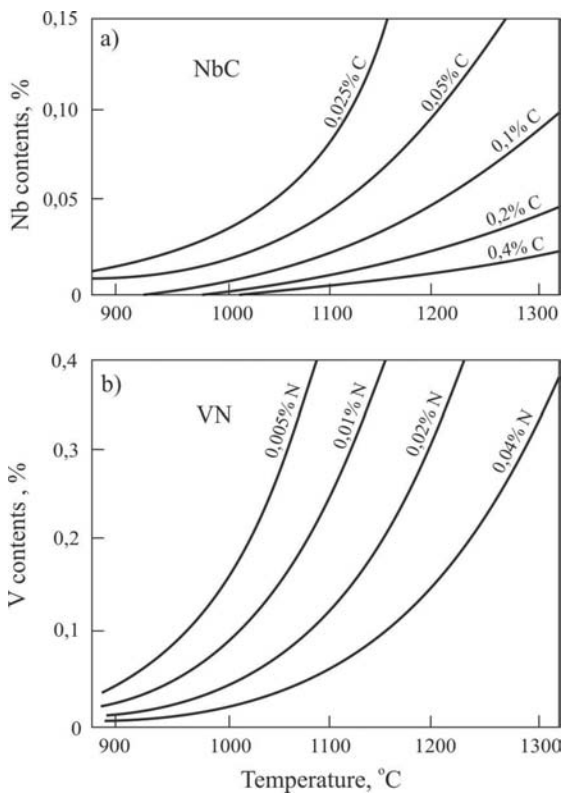


Fig. 9. Solubility curves of NbC and VN in austenite as a function of temperature and content of carbon and nitrogen in steel [8]

Up till now numerous relations have been dealt with describing the size of recrystallized austenite grains as well as the time $t_{0.5}$ that is required for the formation of 50% of the fraction of the recrystallized phase γ , expressing the effect of the preliminary size of the grains in this phase and the conditions of plastic deformation, i.e. temperature and strain rate and degree of deformation. These relations have been dealt with in detail in [8,9] and have proved to be of use in designing the conditions of hot plastic working, e.g. multi-pass rolling of microalloy steels (Fig. 7). In order to obtain metallurgical hardware of microalloy steel with high mechanical properties the conditions of hot plastic working must be adapted to the kinetics of precipitation of the interstitial compounds MX in the austenite of the introduced microadditives in compliance with equation (3). Keeping in mind

that the predominant part of metallurgical products is obtained by rolling, in the case of microalloy steels controlled rolling is applied, mainly concerning steels with a ferritic and ferritic-perlitic structure, as well as rolling connected with thermomechanical treatment viz. thermally improved martensitic and bainitic steels.

Controlled rolling consists in the preliminary rolling of the charge in the upper region of temperature of plastic working, similarly as in the case of common (ordinary) steels, and finishing rolling in the lower range of temperature, corresponding to the precipitation of the phases MX, which restrict the expansion of recrystallized austenite grains between the successive rolling passes. Finishing rolling usually ends at a temperature somewhat lower than the temperature of recrystallization T_R of the austenite up to somewhat exceeding that of A_{F3} or A_{F1} of the steel, followed by cooling the products in open air (Fig. 10).

Due to the presence of slip bands and deformation bands in uncrystallized austenite the population of spots favourable for nucleation in the phase transformation $\gamma \rightarrow \alpha$ increases leading to the formation of more fine-grained structure of metallurgical products.

Thermomechanical rolling (with controlled recrystallization) consists in the rolling of the charge in the temperature range of precipitation of the MX phases including breaks between the successive rolling passes, permitting a complete recrystallization of the austenite and an isothermal holding of the products at an adequately chosen temperature while finishing the plastic treatment for the time $t_{0.5}$ - required for the formation of 50% of the fraction of recrystallized austenite, followed by their intermediate hardening connected with an adequate rate of cooling. The isothermal holding of the products at the temperature towards the end of rolling for the time $t_{0.5}$ is inevitable, because the considerable density of dislocations in the plastically strained austenite increases the critical rate of cooling due to a rather large acceleration of diffusive changes in this phase and reduces the hardenability of the steel because of the impoverishment of the over-cooled austenite with alloy components and the dispersive effect of the carbides. Products manufacture in this way are exclusively subjected to high tempering or aging.

3. Experimental procedure

3.1. Controlled hot forging of machine elements

Microalloyed constructional steels containing about 0.3 to 0.5% C, 0.7 to 1.5% Mn, 0.2 to 0.8% Si and microadditions of Nb, Ti, V and N (up to about 0.1%) in a form of forged parts with small cross section or thickness, are important structural material used in cars, mine machine and equipment, transport equipment, agricultural machinery, etc. These steels grades are brittle fracturing resistance, especially at lower temperatures, without need of applying expensive heat treatment.

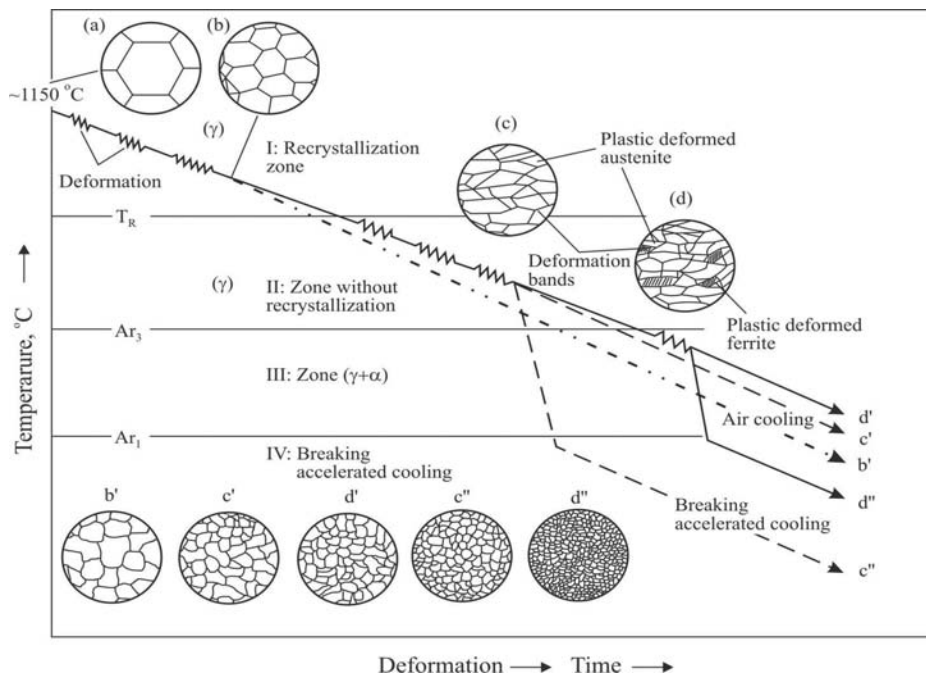


Fig. 10. Schematic diagram of the influence of accelerated cooling on the microstructure of low-carbon microalloy steel products during controlled rolling [15]

Obtaining of these mechanical properties is possible provided steel processing is carried out in conditions enabling the precipitation in the hot formed austenite of the dispersive particles of interstitial phases e.g. nitrides, carbonitrides or carbides of the steel microalloying elements [16]. These precipitations inhibit the growth of the austenite grains when the metadynamic, postdynamic or static recrystallization is over, before the transformation of this phase occurs during cooling [3,4].

Rational use of microalloying requires adjusting the conditions of hot forming of the steel to the kinetics of the solubility of the given interstitial phases in the austenite. For instance, in the V-N microalloyed steels the solubility of the VN nitride depends on the nitrogen content in the steel.

In the case of steel containing about 0.2% C and 200 ppm N, at the final hot forming temperature, about 1000°C, the austenite dissolves only 0.05% V (Fig. 9b). This portion of vanadium is bonded during cooling with N and C into dispersive VN, V(C, N) and VC particles, causing precipitation hardening and making an important contribution to the steel. Ti microalloying has a refining influence on the metal bath due to its affinity to sulphur and is used for continuous cast steels of high metallurgical purity.

The objective here was to determine the influence of austenitizing conditions, hot plastic forming by forging and cooling on the VN nitride precipitation process and transformation of the supercooled austenite also on the structure and mechanical properties of the newly developed V-N microalloyed medium-carbon constructional steel scheduled for forgings and certain sections. The aim of the work was also to evaluate the influence of undissolved V(C,N) precipitates during reheating on structure and mechanical properties of this steel produced by controlled hot forging and accelerated cooling.

Two melts were prepared of V and N microalloyed grades melted at various processes (Table 2). Melt A has been melted in electric arc furnace (20 tons) and ladle VAD treated. Melt B has been melted in an open hearth furnace (100 tons) with no ladle treatment. Following deoxidizing of liquid metal by Al, for both melts V and N additions were introduced in form of Fe-80%V and NITROVAN (79% V+7% C+12% N) - the product of Carbide Union, USA. Liquid metal was bottom poured to 1000 kg (melt A) and 2000 kg (melt B) ingots, which were clogged into sq. 140 mm billets and after reheating rolled to sq. 42 mm bars. Finishing temperature of rolling was about 1050°C.

Table 2. Chemical composition of the steels studied

Melt code	Chemical composition, wt.%					
	C	Mn	Si	P	S	Cr
A	0.27	1.40	0.35	0.012	0.009	0.22
B	0.25	1.40	0.33	0.022	0.030	-
	Ni	Mo	V	Al	N	Cu
A	0.08	0.02	0.20	0.025	0.016	0.09
B	-	-	0.15	0.011	0.018	-

Test pieces cut from the rolled bars were closed and open die forged into the shape of flat bars 12 mm thick with 30 and 70% deformation rate, respectively, in the finish forming temperature range 1000-900°C. The deformation rate was at closed and open die forging 70 and 50s⁻¹, respectively. The charge heated to a temperature of 1180 to 1000°C was air cooled to the forging temperature. After hot forming the flat bars were air-cooled or blast separately to 500°C, or else water-cooled to about 650°C with subsequent air-cooling [17].

Investigations carried out showed that the temperatures of the phase transformations A_{c1} and A_{c3} are 750 and 850°C, respectively, irrespective of the austenitizing temperature of the test pieces in the temperature range 900 to 1050°C. The differentiated quantities of microalloying V and N elements, dissolved in the solid solution and bonded as a VN nitride, do not indicate any influence on the kinetics of the supercooled austenite transformation (Fig. 11). Yet almost complete dissolution of the VN nitride in the solid solution during the austenitizing of test pieces at high temperature followed by air-cooling causes the formation of a coarse grained pearlitic-ferritic structure with enhanced pearlite content. It is the effect of the vanadium alloying, influence on the eutectoid composition that shifts the carbon solubility limit in austenite towards lower values of carbon content.

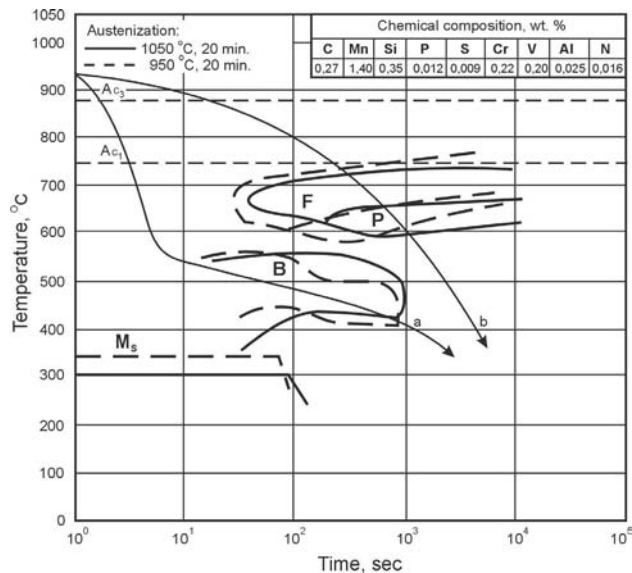


Fig. 11. Phase transformation diagram of undercooled austenite showing schematically cooling conditions: a - ferrite-pearlite structure, b - bainitic structure [16,18]

The steel may obtain ferritic-pearlitic structure with comparable quantities of ferrite and pearlite but exhibiting unsatisfactory mechanical properties, after austenitizing at high temperature and forming in the upper range of test temperature. For instance, the test pieces austenitized at temperature 1180°C, after closed die forming at a temperature of 1000°C followed by air-cooling, exhibit properties comparable with results reported in [18], i.e. U.T.S. about 900 MPa, Y.P.(0.2) 570 MPa, elongation (El) 15%, KCV 23 Jcm⁻² and HB about 270.

Partial dissolving of the VN nitrides in the solid solution due to the lowering of the reheating temperature and plastic deformation in the vicinity of the lower limit of the test temperature range, causes the forming of a fine-grained ferritic-pearlitic structure, with small local quantities of upper bainite (Fig. 12), with increasing influence of the VX (where X-N and C) particles precipitating in the deformed austenite, at the $\gamma \rightarrow \alpha$ deformation front and during the eutectoidal transformation. The fine-grained structure of the ferrite and of the ductile dense

lamellar pearlite (Fig. 13) together with the strengthening influence of the dispersive precipitations of the interstitial VX phases result in improvement of the mechanical properties of the steel. These advantageous properties, including U.T.S. about 900 MPa, Y.P.(0.2) about 690 MPa, El about 17%, KCV more than 30 Jcm⁻², and HB about 270, are acquired by the steel after austenitizing at a temperature of 1000°C, forming at 900°C and air-cooling.

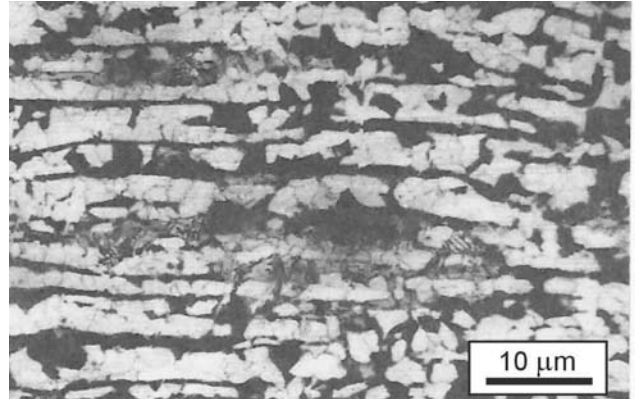


Fig. 12. Fine grained ferritic-pearlitic structure of steel A after austenitizing at temperature 1100°C and closed die forging at 900°C with subsequent air-cooling

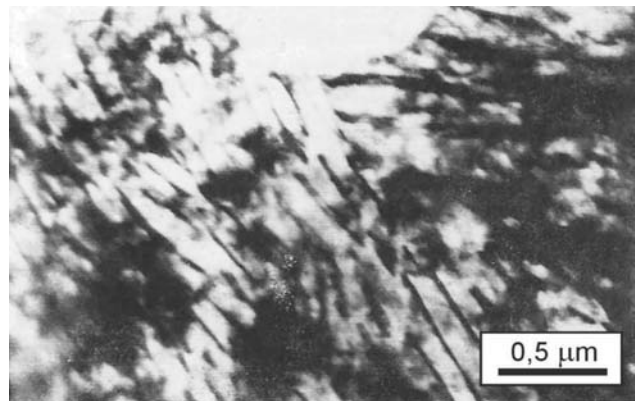


Fig. 13. Structure of a dense lamellar pearlite of steel A

Intensive air-cooling of the test pieces from the forging temperature gives a ferritic-pearlitic-bainitic structure (Fig. 14) with slightly improved mechanical properties but with lower plasticity and brittle fracturing resistance

Reducing the deformation rate from about 70 s⁻¹ (closed die forging) to about 50 s⁻¹ (open die forging), does not change the pattern of the steel plastic deformation process or the phase transformations of the supercooled austenite. Test pieces closed die forged and open die forged at the same austenitizing temperatures, plastic deformation and cooling acquire comparable structure and mechanical properties, although a somewhat more advantageous set of properties is acquired by the steel formed at the higher deformation rate [18].

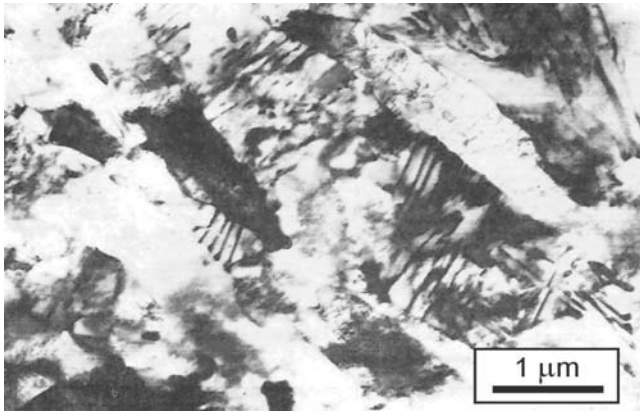


Fig. 14. Ferritic-pearlitic-bainitic structure of the blast cooled steel A from the hot forging temperature

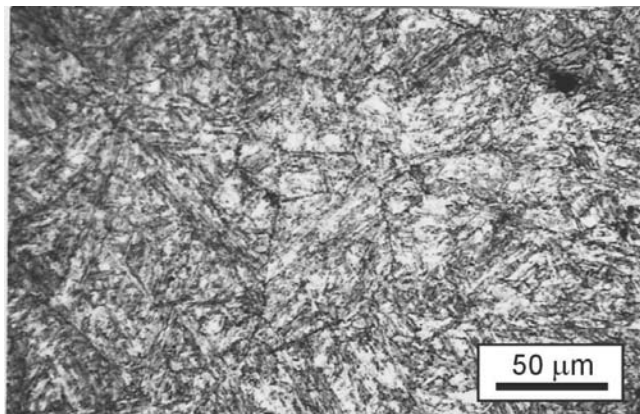


Fig. 15. Bainitic structure of steel A after being water cooled from the hot forging temperature to 650°C with subsequent air cooling

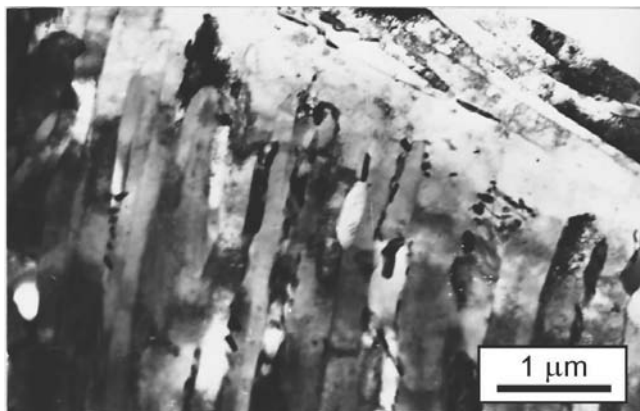


Fig. 16. Bainitic structure of with lath cementite and grainy VN nitride precipitations with the privileged lay-out on the lathes' boundaries (steel A)

Water-cooling of the test pieces from the forming temperature to about 650°C followed by air-cooling gives the steel a bainitic structure (Figs 15 and 16) of irregular shape caused by the influence of the dispersive VX particles on the longitudinal and lateral lath growth.

The morphology of the primary austenite grains flattened in the direction perpendicular to the load as confirmed by results of investigation of test pieces water-cooled directly after forming, indicates that in the test conditions of plastic deformation, dynamic recrystallization does not take place. Dynamic and static recovery of the austenite precede the $\gamma \rightarrow \alpha$ eutectoidal, and bainite transformations along with simultaneously precipitation of the dispersive isomorphous VN, V(C,N), and VC particles. The increasing quality of precipitates with decreased austenitizing temperature and plastic deformation, diminishes the role of precipitation hardening in bainite strengthening but causes significant improvement of the steel notch toughness. An advantageous set of properties, i.e. U.T.S. about 880 MPa, Y.P.(0.2) about 650 MPa, elongation (El) about 18%, KCV more than 60 Jcm⁻² and hardness about 270-280 HB, is acquired by the steel after austenitizing at temperature 1000°C, finish forming at 900°C and water-cooling to 650°C with subsequent air-cooling. Slightly better mechanical properties but less ductility is observed in steel formed at lower deformation rate. Brittle fracture resistance of the steel decrease linearly with decrease of the test temperature, and depending on conditions of austenitizing, plastic deformation and cooling, the steel maintains a notch toughness at anobut KCV 10-20 Jcm⁻² at 0-60°C. The changes in notch toughness found are fully reflected in the course of the steel brittle fracturing process Fractography analysis of test piece fractures showed that transcrystalline brittle fracturing of the steel takes place with average distributive fracture quantity increasing with the increase of temperature of austenitizing and plastic deformation and with decrease in the test temperature. Relatively good brittle fracturing resistance at lower temperatures is continued by the fully developed fracture surfaces of the test pieces with a line grained ferritic-pearlitic structure, and still more for those with bainitic structure.

Refinement of austenite grains by decreased reheating and forming temperatures improve toughness of forgings with bainitic and tempered martensite structures. Particularly good combination of mechanical properties is attained in the steel with upper bainite lath structure, formed during interrupted cooling to about 500°C followed by air cooling (Table 3). Forgings with such structure, compared to fine-grained ferrite-pearlite one present improved mechanical properties and appropriate toughness even at minus 40°C, despite of limited plasticity.

Table 3. Effect of structure on the mechanical properties of the steel reheated at 1000°C, formed at 900°C and control cooled

Melt code	Structure	UTS MPa	YS MPa	El %	KV _{50°C} J	KV _{-40°C} J	HB
A	ferrite-pearlite	905	694	17	25.8	14.1	270
B	ferrite-pearlite	705	564	23	57.1	25.6	230
	lath bainite	835	742	13	45.2	40.7	295

From the results of the investigations carried out energy saving technology was developed for plastic forming of the steel products exhibiting high mechanical properties and a desirable brittle fracture resistance without the need, to carry out any heat treatment [20]. Result show that differing conditions of austenitizing and its plastic deformation with deformation rates equivalent to forging rates and cooling after hot forming, significantly influence the structure and mechanical properties of the V-N microalloyed constructional steel.

Austenitizing of the steel at high temperature giving complete dissolution of the microalloying V and N elements in the solid solution is of no advantage since it gives rise after hot forming and subsequent cooling, to a coarse grained ferrite-pearlitic or bainitic structure with the enhanced mechanical properties but with reduced hot-workability and brittle resistance.

In the investigated conditions of hot forming of the steel, dynamic recrystallization does not occur but only dynamic and static recovery with simultaneous precipitation in the austenite of the isomorphous nitrides, carbonitrides and vanadium nitrides, which have a significant influence on the process of nucleation and growth of the phases in the following transformations: $\gamma \rightarrow \alpha$, pearlitic and bainitic.

The content of interstitial phases, increasing with lowering austenitizing and plastic deformation temperatures, which are precipitated before the transformation occurs during cooling, causes decrease in the contribution of precipitation hardening in steel strengthening but enhancing its hot-working properties and notch toughness due to the influence of the dispersive VX particles in forming of a fine grained ferrite-pearlitic or bainitic structure.

The range of optimal conditions for austenitizing and hot forming of the steel may be extended by a rational selection of concentration of vanadium and nitrogen in the steel to secure the precipitation of only the VN nitride during plastic deformation and cooling.

Presented study reveals that share of undissolved V(C,N) precipitates controlled by reheating conditions, influence to lesser extend mechanical properties but apparently improve plasticity and toughness of ferrite-pearlitic also bainitic structures.

It is therefore not recommended to apply high reheating temperature at which V(C,N) completely dissolves, causing austenite grain growth and coarsening recrystallized austenite grain growth after forming.

To attain good toughness particularly at low temperatures, it is recommended to reheat blanks below 1000°C where fine grain structure of austenite persist pinned by large fraction of undissolved V(C,N) particles.

Coagulation of secondary phases leads to solution of small and growth of large V(C,N) particles with decrease in their total number - the process that decides on effectiveness of retardation of grain boundary mobility. This is the reason for sudden grain

growth at austenitizing temperatures above 1000°C, despite of the fact that undissolved V(C,N) are still present in large quantities.

Taking into account the large contribution of V(C,N) to precipitation hardening with simultaneous decrease in ductility of the steel, particularly with ferrite-pearlite structure, vanadium content should not exceed 0.15 % and nitrogen content should be limited to value necessary to form VN nitride.

The mechanical properties of the steel are lowered by large amount of elongated non-metallic inclusions promoting banding.

Forgings with structure of lath bainite present improved mechanical properties and toughness compared to ferrite-pearlite structure.

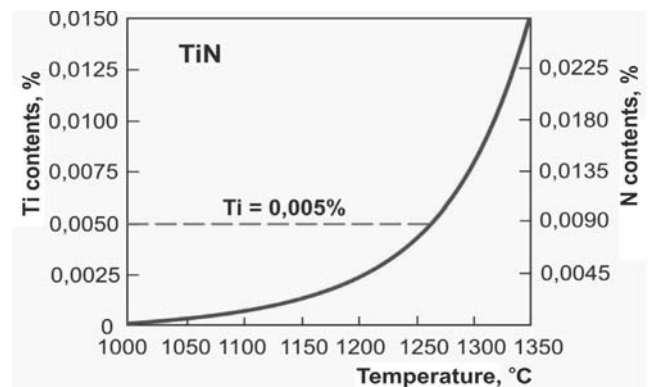


Fig. 17. Solubility curve of TiN nitrides in austenite of A type steel as a function of temperature

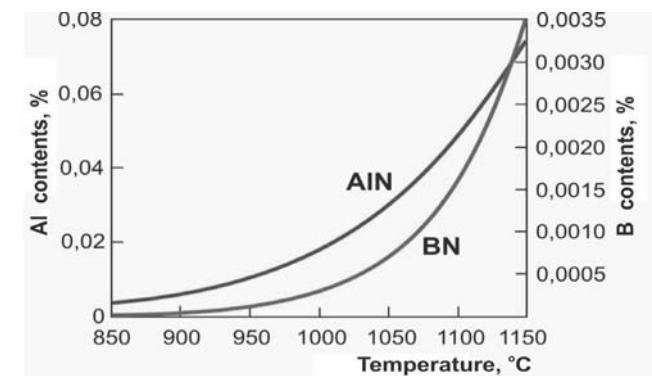


Fig. 18. Solubility curves of AlN and BN nitrides in austenite of A type steel as a function of temperature

Table 4.

Chemical composition of tested steels [21,22]

Steel	Chemical composition, wt. %														
	C	Mn	Si	P	S	Cr	Ni	Mo	Ti	V	B	Cu	Al	N	
A	0.21	1.02	0.25	0.018	0.005	0.18	0.07	0.03	0.005	0.008	0.002	0.11	0.024	0.009	
B	0.21	1.03	0.23	0.018	0.013	0.14	0.07	0.03	0.032	-	0.002	0.14	0.024	0.008	
C	0.32	1.17	0.25	0.015	0.007	0.22	0.07	0.02	0.035	0.026	0.003	0.23	0.025	0.010	

3.2. Production of element forged by thermomechanical treatment

The goal of this experiment are the structure and the mechanical properties of forged elements using thermo-mechanical treatment, made of C-Mn steel with microadditions of Ti, V, B and N (Table 4) melted: after furnace treatment of metal bath, and continuous casting ingots with a cross section of 100×100 mm. After their solidifications the ingots were rolled to bars with a dimension of about 36 mm, and after using the same conditions to bars with the dimension of 17 mm.

The range of controlled rolling temperature was based on calculation of solubility in austenite microadditions to the steel. In order to make this kind of calculation the authors used the kinetic equation (3). It was found that a microaddition of Ti introduced to the steel type A with a concentration of 0,005% is completely dissolved in austenite at a temperature of about 1250°C (Fig. 17) and during cooling does not absorb the whole N from the steel to TiN. In this case of an excess of N is creating nitrides of AlN and BN during cooling the steel (Fig. 18).

Boron is fixed to stable nitride BN, practically not increasing the hardenability of the steel. On the other hand, nitride from steels type B and C is completely fixed in TiN. Total dissolving of that phase in the austenite is possible only at temperatures of 1400 and 1450°C, respectively for steel type B and C. The rest of Ti introduced to those steels amount is fixes itself to carbonitrides Ti(C,N) and carbides TiC. However, Al remained in dissolved state and B is fixing to $M_{23}(C,B)_6$ during cooling of the steel which could be dissolved in the solid solution at temperatures a little higher than temperature A_{c3} of the steel.

The charge heating temperature for forging was calculated basing on the primary austenite grain size index of the samples quenched at the programmed increasing austenitizing temperature (Fig. 19). As could be seen, steel type A with a small concentration of Ti and a low volume of the TiN phase displays a distinct growth of austenite grains just after crossing 900°C.

Whereas the growth of grains of the γ phase of the steel type B proceed softly up to 1100°C, in steel type C it grows even higher than to 1150°C. Basing on data which were quoted, the charge heating temperatures were adjusted for steel A type - 950°C, and for steel B and C respectively 1000 and 1150°C. The time needed for the static recrystallization of the γ phase after finishing plastic deformation at the temperature at the end of forging was determined basing on the primary austenite grain size of the forged samples with $\epsilon = 0.25$ at a temperature of 900°C with strain rate of 14 s^{-1} , before quenching held in water for 0 up to 24 s (Fig. 20). As has been shown in that diagram steel type A reaches the finest austenite grain size after holding the samples at this temperature for 3 s, whereas steels B and C for 12 s and 16 s respectively.

Thermo-mechanical treatment was realized thanks to open die forging of experimental segments with 17 mm diameter and 150 mm length in the temperature range of 950 to 900°C and 1000 to 900°C - respectively for steel type A and B to rods with intersection of 12 x 12 mm. Before quenching in water these rods were held at a temperature at the end of forging for 3 and 12 s. Quenched rods were tempered at a temperature of 600°C for 1 h, whereas rod segments of steel type C with an intersection of 24 x 24 mm were forged in the range of temperature from 1150 up to 900°C to the shape of rods with an intersection of 12 x 12 mm. Before quenching in water the rods were held at a temperature at

the end of forging for 16 s, and then were tempered at a temperature of 500 and 600°C for 1 h.

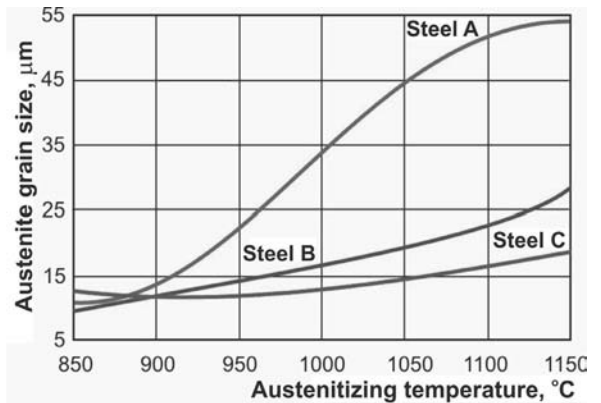


Fig. 19. Influence of the austenitizing temperature on the grain size of austenite

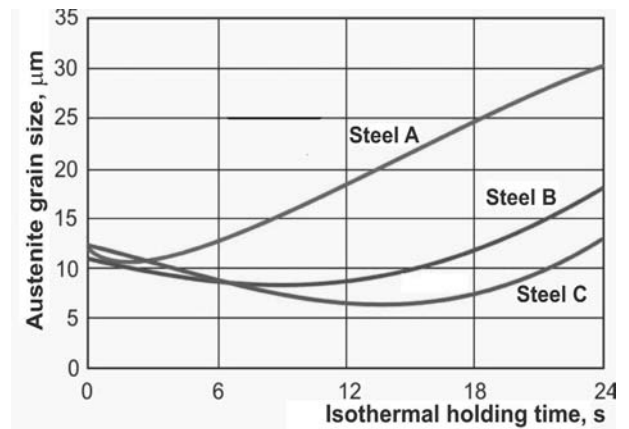


Fig. 20. Influence of the isothermal holding time of the specimens after completing hot-working at a temperature of 900°C before water quenching on the grain size of austenite

Investigations have proved that after thermo-mechanical treatment and after quenching the examined steels have a fine-grain structure of primary austenite (Figs 21, 22) with a grain size about 10, 5 and 8 μm and a martensitic-bainitic structure (Fig. 23) and hardness 42, 44 and 49 HRC - respectively for steel type A, B and C. The hardness of the steels decreases after high tempering from 220 to 250 HBW, which does not create difficulties, during the mechanical treatment of forged elements.

Examinations of the structure of thin foils performed in TEM (transmission electron microscope) showed that type C steel (quenched at a temperature of the end of forging - 900°C and held previously at this temperature for 16 s) have a lath martensite structure (Fig. 24). Inside the martensite laths the presence of dispersive particles of cementite (Fig. 25) has been proved, whereas at the boundaries of primary austenite $M_{23}(C,B)_6$ dispersive particles were found (Fig. 26), which had occurred in the steel during the self tempering process.

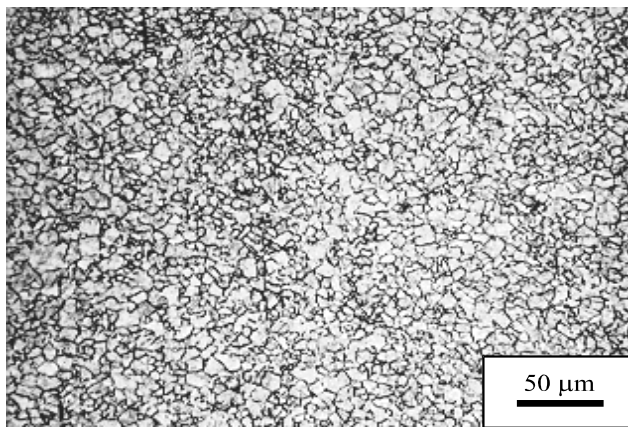


Fig. 21. Fine-grained structure of statically recrystallized austenite; finishing forging temperature 900°C; isothermal holding time 12 s (steel B)

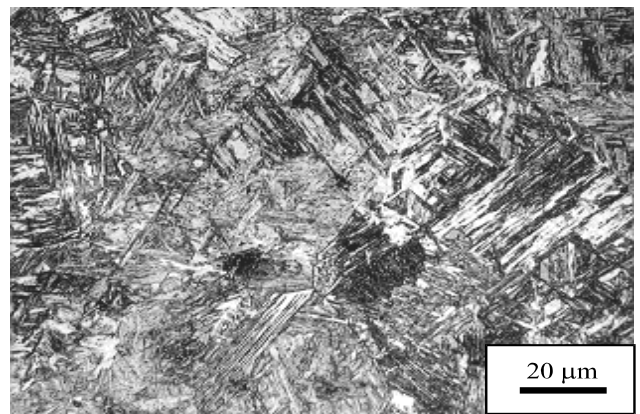


Fig. 23. Martensitic-bainitic structure of steel B water quenched at a finishing hot-working temperature of 900°C after holding for 12 s

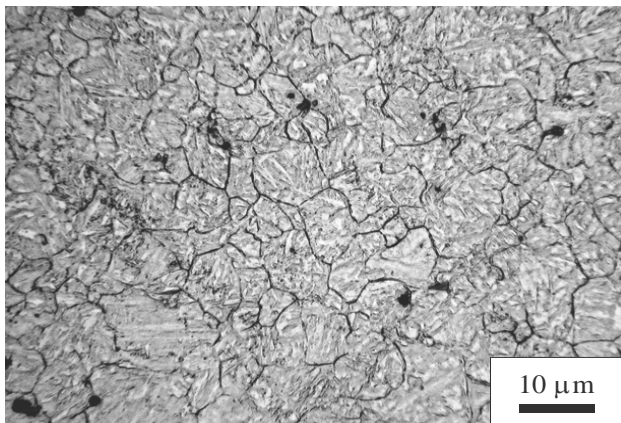


Fig. 22. Fine-grained structure of statically recrystallized austenite; finishing forging temperature 900°C; isothermal holding time 16 s (steel C)

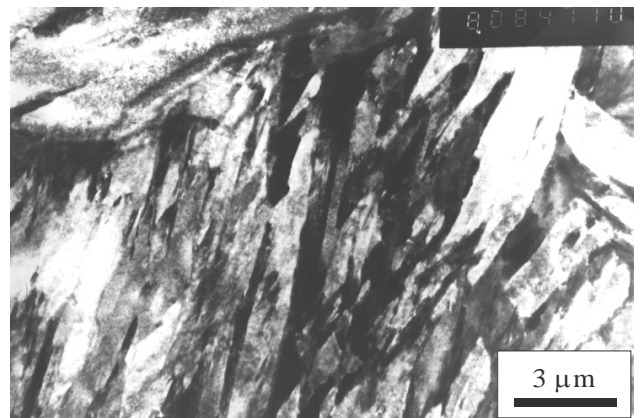


Fig. 24. Lath martensite structure of steel C quenched from a finishing hot working temperature of 900°C after holding for 16 s

Particular attention should be focused on the high mechanical properties of the steels in their high tempered state and especially on their crack resistance also at low temperatures (Table 5). This table also shows the ideal diameter D_I , characterizing their hardenability. The ideal diameter was calculated in compliance with the procedure described in the ASTM A255-89 standard. A microadditive of boron introduced into the steel type A is totally fixed in nitride BN and because of this it does not increase hardenability of the steel. On the contrary in steels B and C a micro-additive of boron dissolved in solid solution affects the hardenability. It has been noticed that the calculated ideal diameter D_I of steel type A is equal to 28 mm, and the calculated ideal diameter of steels type B and C, taking into account the influence of micro-additive of boron D_{IB} is equal to 55 and 96 mm respectively. This suggests that steel type A is useful for forging relatively small intersections and the steels B and C type for big ones.

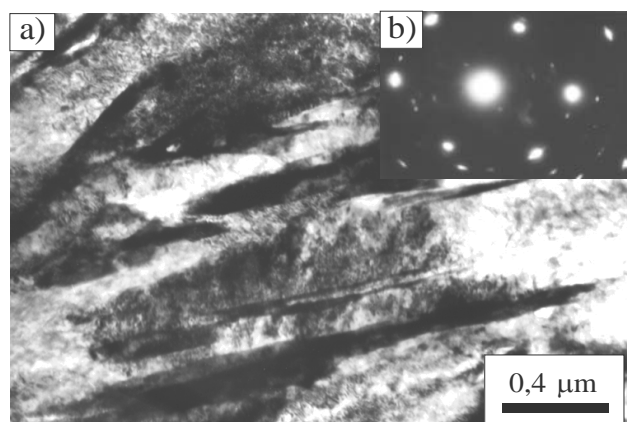


Fig. 25. Lath martensite structure with dispersive Fe_3C precipitations (steel C): a - light field, b - diffraction pattern according to Fig. 25a $[111]Fe\alpha$ and $[012]Fe_3C$

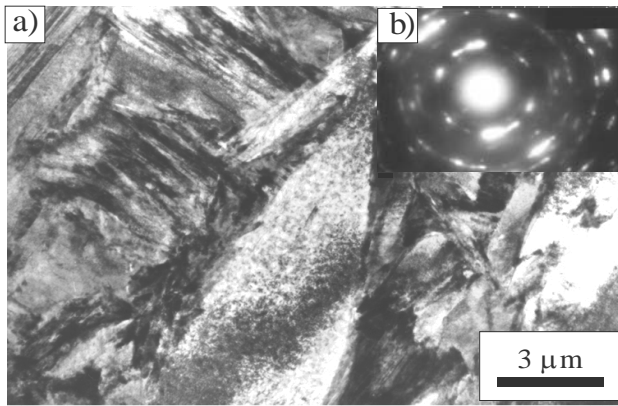


Fig. 26 Dispersive $M_{23}(C,B)_6$ precipitations at the grain boundaries of primary austenite (steel C); a - light field, b - diffraction pattern according to Fig. 26a $[111]Fe\alpha$ and $[001]M_{23}(C,B)_6$

Investigations carried out show the full usability of micro-alloyed steels for the production of forged machine parts, with high strength and good cracking resistance, using the energy-saving thermo-mechanical treatment method. This thermo-mechanical treatment, allows to obtain a fine-grained austenite

structure by hot plastic deformation, and gives us forged elements obtaining the yield point $YS_{0.2}$ over 690 MPa, UTS over 770 MPa, hardness 220 up to 250 HBW and breaking energy KV over 180 J.

Designing forging technology of micro-alloys steels needs the accommodation of heating conditions to the kinetics of precipitation processes of MX phases in austenite, without growing grains in the γ phase. Thanks to investigations of the influence of austenitizing temperature on the primary austenite grain size, and kinetics of dissolving interstitial MX phases in austenite it was found that the heating temperature for forging elements made of

steel A should not exceed 950°C, and at elements made of steels B and C could have even 1150°C. This shows that a charge of steels B and C could be heated up to a considerably higher temperature than A_{c3} for that steel, keeping its fine-grained structure. This increases the live time of the dies.

Introducing micro-additives of boron to the steel (which increase the hardenability) made of fine-grained steel needs a Ti shield in a quantity indispensable to fix all nitrogen to BN. When the concentration of Ti is too small to consume the whole nitrogen from the steel, boron is turned off from affecting the hardenability of the investigated steels. Also a small amount of TiN nitrides does not protect steel against the growth of austenite grains, at a high temperature used to heat the charge [23-25].

3.3. Controlled rolling of thick plates

Controlled rolling of thick plates is performed on steel A (Table 6), whose chemical composition has been developed basing on data quoted in literature and materials advertised by the firms Thyssen and Svenkst Stål, concerning steels for the production of plates with proof stress $YS_{0.2}$ from 550 to 900 MPa, tensile strength UTS from 670 to 1100 MPa, elongation at fracture (El) from 18 to 12% and energy KV_{-40°C} above 40 J [8]. Steel smelted in compliance with open hearth melt process has subjected to steel refining in order to get the required chemical composition and metallurgical purity. These attempts provided only a partial success, because the moist materials into the bath with the purpose of desulphuriaztion and modification of non-metallic inclusions involves an excessive increase of the concentration of hydrogen and unkilld steel. Due to the impossibility of vacuum degassing of the bath attempts have been made to reduce the excess of hydrogen by means of low cooling of the flat ingots in the course of the solidification cast from the top (16 hours) and their cooling after the disassembly of the moulds for also 16 hours. The rolled blooms, 60 mm thick, were subjected to dehydrogenation for 32 hours at 650°C.

Table 5. Results of the mechanical properties and hardenability of the investigated steels

Steel	Treatment type		Mechanical properties							
	Treatment conditions	Tempering temperature, °C	$YS_{0.2}$ MPa	UTS MPa	El %	RA %	KV J	HBW	D_1 mm	
A	900°C/3s/water	600	695	770	22	69	186	230	28	
B	900°C/12s/water	600	700	786	24	75	197	220	55	
C	900°C/16s/water	500	1021	1115	13	50	72*	275	96	
		600	852	932	18	65	108*	250		

* - specimen breaking energy tested at temperature of - 20°C

Table 6. Chemical composition of the investigated steels

Steel	Mass contents, wt. %													
	C	Mn	Si	P	S	Cr	Ni	Mo	Nb	Ti	V	Al	B	N
A	0.17	1.30	0.27	0.018	0.003	0.53	0.91	0.41	0.025	0.025	0.008	0.03	0.003	0.008
B	0.17	1.37	0.26	0.012	0.001	0.24	0.05	0.48	0.025	0.004	0.020	0.06	0.003	0.004

Table 7.

Effect of heat treatment conditions on the mechanical properties of the plates of steel A

Temperature, °C		YS _{0.2} , MPa	UTS, MPa	El, %	YS _{0.2} / UTS	KV _{50°C} , J		KV _t /KV _l
quenching	tempering					KV _l	KV _t	
890	600	869	914	13.8	0.95	37.4	31.8	0.85
	620	796	864	14.8	0.92	36.5	33.4	0.91
	640	785	831	15.0	0.94	49.6	34.5	0.69
	660	739	795	15.3	0.93	40.6	32.9	0.81
940	600	895	954	14.3	0.94	33.4	28.6	0.85
	620	839	888	14.4	0.84	33.8	27.5	0.81
	640	727	859	13.5	0.93	39.5	28.1	0.71

KV_t - transverse specimens, KV_l - longitudinal specimens

Defectoscopically selected blooms were subjected to controlled rolling at a temperature of 1050-850°C to sheets with a thickness of 20 mm. Basing on repeated defectoscopic investigations from the rolled sheet samples were cut out, toughened by quenching from 890 or 940°C after their austenitization for half an hour and tempering for 1 hour within the range of temperature from 600 to 660°C.

Metallographic investigations of the tempered samples have shown that the obtained sheets display a fine-grained structure of primary austenite with a slight share of non-metallic inclusions of globular morphology, which proves that the ladle furnace refining of the liquid steel was carried out correctly. In the direction of rolling narrow decarbonized zones were detected affected by hydrogen. In quenched and tempered samples in these zones a slight consolidation of dispersive carbides was to be observed [26,27].

Tensile tests carried out on samples situated laterally to the direction of rolling proved that steel sheets hardened at a temperature of 890°C after toughening display more favourable mechanical properties (Table 7). After tempering within the range of temperature from 600 to 660°C such sheets indicate YS_{0.2} from about 870 to 740 MPa, UTS from 914 to 795 MPa, El from 13.8 to 15.3% and the ratio YS_{0.2}/UTS from 0.95 to 0.93, as well as a higher energy of breaking Charpy V samples, lateral and longitudinal KV_{50°C} of lateral samples to longitudinal ones. The value of this ratio characterizes the anisotropy of the plastic properties of the steel sheets, which is in the case of sheets quench hardened from the temperature 890°C is smaller than after quenching from 940°C.

The data gathered in Table 7 indicate that the mechanical properties of the obtained steel sheets are comparable with those of WELDOX and XABO steels sheets with the same strength, delivered from abroad. Higher technological requirements, the introduction of vacuum degassing and continuous casting of ingots as well as acquisition of quality accreditations by insurance companies will permit to start the production of thick weldable high strength steel plates in Poland, competitive with those produced by foreign firms.

3.4. Rolling of thick plates by means of thermomechanical treatment

The investigations were made using imported heat-treatable steel B (Table 6), delivered as 40 mm thick plates. Secondary

metallurgy was employed in the melting process of the molten steel along with the modification of nonmetallic inclusions, using calcium compounds and vacuum degassing of the molten metal. Cast strands were initially rolled to a 40 mm thick plate.

The phase transformation temperatures and impact of the plastic deformation on the phase transformation of austenite were determined, using dilatometric methods. Basing on the determined A_{c3} temperature for steel, equal to 875°C, the austenitizing temperature for the specimens was assumed as 900°C. Plots of transformations of supercooled austenite TTT were made and of the austenite plastically deformed by compressing the specimens at the temperature of 900°C with the 50% deformation and deformation rate of 3.5 s⁻¹.

The investigations of the hot working of steel and the kinetics of the strain hardening decay of the austenite at intervals between the deformation cycles of specimens with the same value of $\varepsilon = 0.2$ at the rate 3 s⁻¹ were carried out in conditions close to plate rolling. The method of specimen twisting was employed using the SETERAM-7N torsional plastometer. O.D. 6 mm, 10 mm long specimens were subjected to the torsion test. The specimens fixed in plastometer jaws, shielded by a quartz tube with a flow of argon, were induction-heated to the temperature of 1150°C, and after holding for 10 minutes they were cooled to the test temperature. Torsional torque, axial load, and temperature as a function of the number of revolutions made to obtain the required deformation or to the specimen damage were logged. The plastometer made it possible to stop the test at the required deformation and immediate cooling of the specimens with water to freeze the structure of the plastically hot-deformed austenite. The activation energy of hot plastic deformation of the steel was determined basing on the continuous twisting of the specimens to their damage. The strain hardening decay kinetic curves, corresponding roughly to the curves of static recrystallization kinetics of the austenite were determined, basing on the intermittent torsion test for the required deformation $\varepsilon = 0.2 < \varepsilon_m$ (ε_m - deformation corresponding to the maximum yield stress value) with isothermal holding of the specimens between two deformation stages for 0.2 to 120 sec.

The results of investigations of austenite recrystallization kinetics were used to determine the times t_R and $t_{0.5}$, necessary for developing a 50% fraction of the recrystallized phase γ as a function of temperature. These investigations made it also possible to work out the conditions of the thermo-mechanical treatment - rolling with controlled recrystallization - of a flat bar

with a transverse section of 200 x 40 mm to a plate 15 mm thick in five passes within the temperature range of 1000 to 900°C with decrements of 50°C. Pass reductions of 20% were applied in the first three passes, and 15% in the last two ones.

The thermo-mechanical treatment of plates was carried out according to two variants. In the first one, the strip was air-cooled between the passes to the temperature of the successive pass and held for the time $t_{0,5}$ in stable conditions after completing the rolling, before quenching in water. In the second variant, however, a retention shield was used, making it possible to adjust the intervals between the successive passes to a time close to t_R and to hold the plate isothermally at the temperature at the end of rolling for the time $t_{0,5}$ before its quenching.

The test specimens of plates were tempered within the temperature range of 550 to 650°C. To evaluate the effect of the conditions of plastic deformation of the structure and mechanical properties of the steel, the conventional heat treatment of steel was carried out, i.e., quenching of specimens at the austenitizing temperature of 900°C, proper for this steel, and tempering in conditions identical with those applied to the rolled and quenched test plates.

A comparison of the transformations plots of the plastically undeformed and deformed supercooled austenite (Fig. 27) reveals that plastic deformation before its transformation results in a significant promoting of diffusion transformations, i.e., of the ferritic and pearlitic ones, and in shifting the bainitic transformation towards the temperature axis with the simultaneous lowering of the temperature M_s of the steel.

The increased diffusion rate in the plastically deformed supercooled austenite and the significant density of locations convenient for heterogenous nucleation of the diffusion transformation products of this phase, feature the decisive factor causing a shift of the boundary of phase transformations of the supercooled austenite in its both states. However, a slight lowering of the temperature M_s of the plastically deformed steel is connected with enriching the austenite with carbon before the front line of the bainitic transformation.

Continuous torsion tests of the specimens made it possible to determine the effect of temperature within the range of 1100 to 900°C on the shapes of the σ - ϵ curves and deformations ϵ_m - corresponding to the maximum yield stress value (Fig. 28), and therefore, the evaluation of the deformation necessary to originate the dynamical austenite recrystallization $\epsilon_{cd} \approx 0.8 \epsilon_m$. The calculated energy of the plastic deformation process for the investigated steel is 306,95 kJ·mol⁻¹. The value of this energy is comparable with the self-diffusion energy, which means that the dynamic recovery is a heat-activated mechanism, controlling the plastic deformation of the steel in the investigated conditions, which occurs mostly by dislocation climbing [28,29].

Torsion tests of the specimens made to obtain the required deformation revealed, that depending on temperature and the time of holding the specimens between two deformation stages, partial or total decay of strain hardening occurs. It is the result of recovery and static recrystallization processes of the plastically deformed austenite. The strain hardening decay kinetics curves obtained for the austenite (Fig. 29) demonstrate that the time of the total austenite recrystallization t_R at temperature of 1100 °C is about 50 sec and increases to about 300 sec at a temperature of 900°C. However, the $t_{0,5}$ time required for developing of 50%

fraction of the recrystallized austenite at a temperature of 1100 °C is 9 sec, and increases to 60 sec at a temperature of 900 °C.

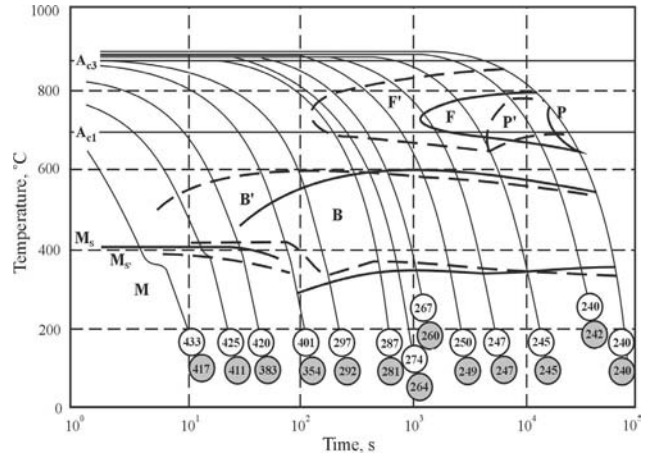


Fig. 27. Effect of the plastic deformation on the supercooled austenite transformation curves of steel B; solid line - non-deformed austenite; broken line - plastically deformed austenite

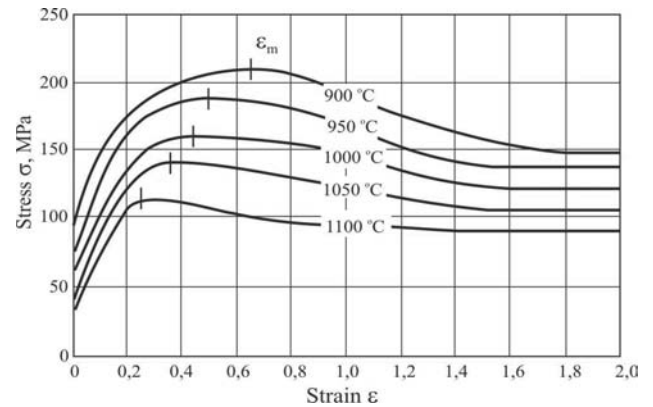


Fig. 28. Effect of temperature on the σ - ϵ curves in the twist test with the deformation rate of 3 s⁻¹ (steel B)

Basing on Fig. 29, the times of total recrystallization t_R and semirecrystallization $t_{0,5}$, as well as the recrystallized austenite grain size, were determined as functions of temperature (Fig. 30). The grain size $d_{RS} = f(T)$ shown in this figure corresponds roughly to the model relationship given by Hodgson [30], whereas, the time $t_{0,5} = f(T)$ corresponds to the Beynon and Sellars relationship. The metallographic examinations of the specimens deformed in cycles revealed that if the interval between the consecutive deformations is smaller than the time of the total austenite recrystallization t_R , then because of the superposition of the consecutive deformation cycles the shearing bands develop in the austenite, in which of carbon, chromium, molybdenum, niobium, titanium and vanadium are segregated. Similar segregation bands were revealed in the primary austenite (Fig. 31) and in martensite (Fig. 32) of the test sections of plates using the first variant of the thermo-mechanical treatment.

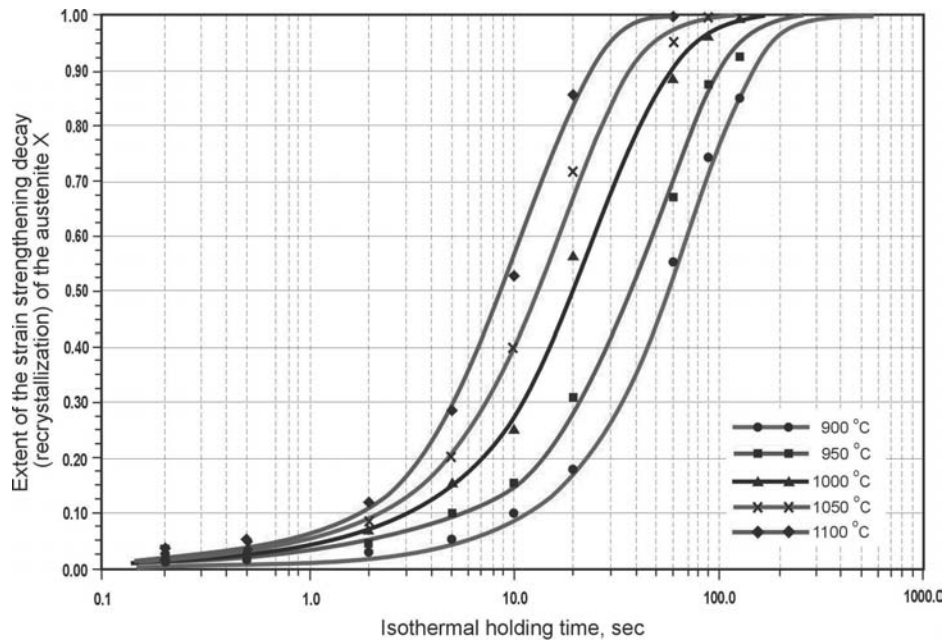


Fig. 29. Effect of the test temperature of the steel B on the strain hardening kinetics decay curves of the plastically deformed austenite in the torsion test at the rate of 3 s^{-1} ($\epsilon = 0.2$)

Table 8.

Intervals between the consecutive passes t_p and the relevant extents of the austenite X strain strengthening decay in both strip cooling variants

No	Temperature range, °C	Time t^*_R , sec	Variant I		Variant II		Time at the temperature of the consecutive pass		
			t_p , sec	X	t_p , sec	X	$t_{0.5}$, sec	$t_{0.8}$, sec	$t_{0.9}$, sec
1	1100-1050	70	53	0.95	70	1.00	12	28	40
2	1050-1000	110	36	0.80	60	0.95	20	45	60
3	1000-950	165	26	0.45	53	0.88	40	80	100
4	950-900	225	14	0.15	40	0.52	60	110	170

t^* - calculated as the arithmetic average of the austenite t_R times at the temperatures of the current and successive passes

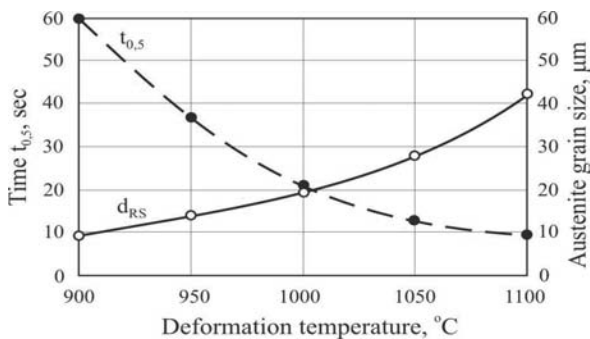


Fig. 30. Time $t_{0.5}$ determined experimentally, and the recrystallized austenite grain size d_{RS} as a function of test temperature [31]

The significant density of dislocations in the developing bands is the reason for segregation of C, Cr, Mo, Nb, Ti and V, and therefore, the martensite developed in these regions has

distinctly a higher hardness than the surrounding steel matrix (Fig. 33). Segregation bands do not develop if the interval between the successive deformation cycles makes total recrystallization of the plastically deformed austenite possible. Martensite originated due to quenching of the test plate segments from the finish-rolling temperatures has a lamellar structure (Fig. 34) with a non-uniform lamellar thickness, often curved because their growth is affected by the statically recovered austenite with a varying dislocation density and dislocation grain boundaries. The segregation bands occurring in the martensite do not disappear during tempering of the steel but transform into the layered granular cementite precipitations (Fig. 35).

Adjusting the interval times between the consecutive deformation cycles to the time t_R , by employing the retention shield (Table 8), and the isothermal holding of the rolled strip for a time $t_{0.5}$ at the temperature of the end of hot working before quenching, has added to an improvement of the homogeneity of the chemical composition. The steel structure developed in these conditions contains about 20% of the fine-grained, statically recrystallized austenite, and the remaining portion of the phase in

the significantly larger grains of the statically recovered austenite, with no traces of segregation bands (Fig. 36).

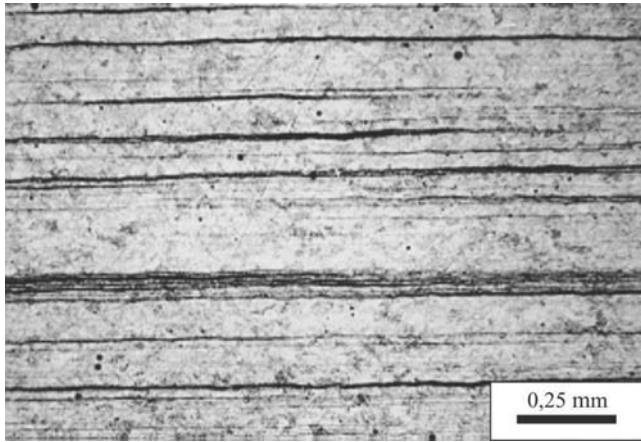


Fig. 31. Austenite structure with distinct segregation bands [32]

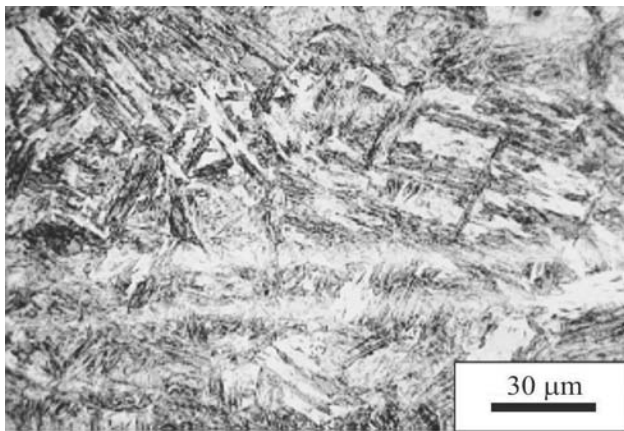


Fig. 32. Low-carbon lath martensite with distinct segregation bands (steel B)

This means that the time $t_{0,5}$, determined experimentally basing on the strain hardening decay kinetics does not lead to an origination of 50% fraction of the recrystallized austenite, as the static recovery occurring in these conditions has a significant input to the decrease of the yield stress of steel.

Steel quenched from the finish-rolling temperature after the time $t_{0,5}$ from completing the plastic working, has - similarly as the one presented in Fig. 34 - a lath martensite structure with dispersive precipitations of the cementite within the lamellae (Fig. 37), originated in the self-tempering process, and after high tempering, granular precipitations of this phase are distributed on the recovered ferrite lathes' boundaries (Fig. 38). The differentiated steel structure, both in the quenched state for both thermo-mechanical working variants and in the tempered state has a small effect on the mechanical properties, however, a significant one on the Charpy V-notched specimens' breaking energy at a temperature of -40°C (Table 9) and anisotropy of the plastic properties of the plate test segments, expressed by the ratio of the specimens transverse to longitudinal $KV_{-40^{\circ}\text{C}}$.

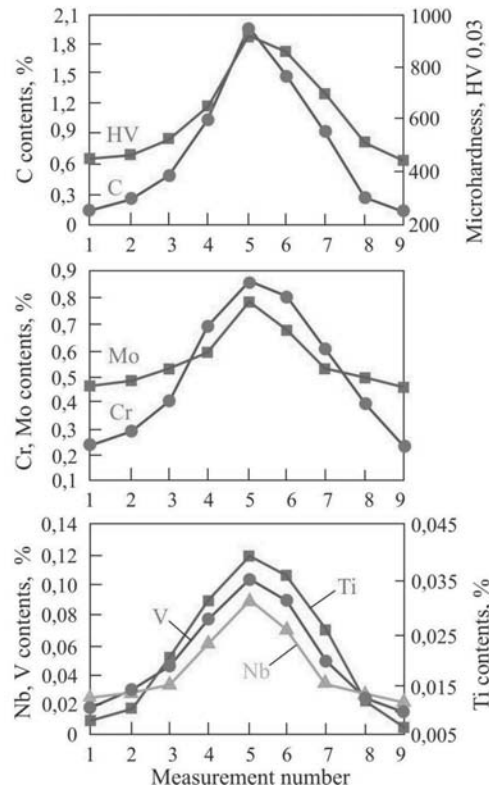
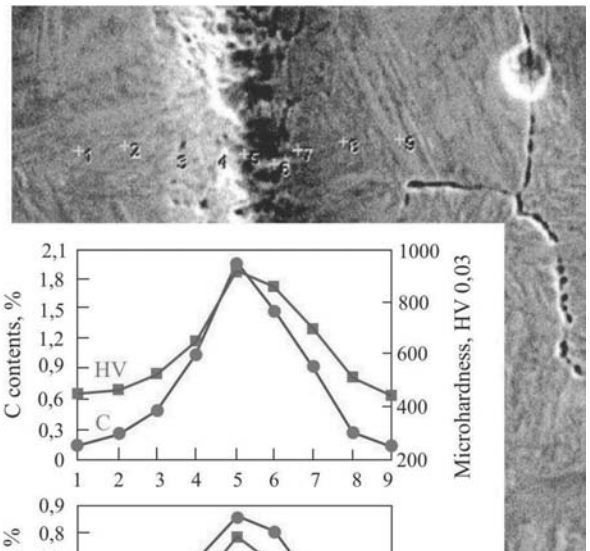


Fig. 33. C, Cr, Mo, Nb, V and Ti concentration and hardness distributions in the segregation band and in the adjacent quenched steel zone

Tensile test results indicate that the test plate segments made in both thermo-mechanical working variants, tempered in the temperature range 550 to 650°C 's demonstrate comparable mechanical properties, that is: yield stress $YS_{0,2}$ from 1000 to 1089 MPa, Ultimate Tensile Strength (UTS) from 1030 to 1120 MPa, elongation (El) from 17 to 19%, and reduction of the area at a fracture (RA) from 60 to 69%. Top $YS_{0,2}$ and UTS values refer to the test plate segment using the first thermo-mechanical working variant, whereas, the bottom ones refer to the second variant. Moreover, transverse specimens have, in general, slightly better strength properties, and lower plastic ones.

However, conventionally quenched steel, after austenitizing the specimens at the temperature of 900°C , demonstrates properties comparable with those given above for the thermo-mechanically worked specimens, only after tempering at a temperature of 550°C . Increasing the tempering temperature to 650°C results in decreasing the yield stress $YS_{0,2}$ to 990 MPa with UTS of about 1040 MPa, El about 19% and RA from 67 to 70%.

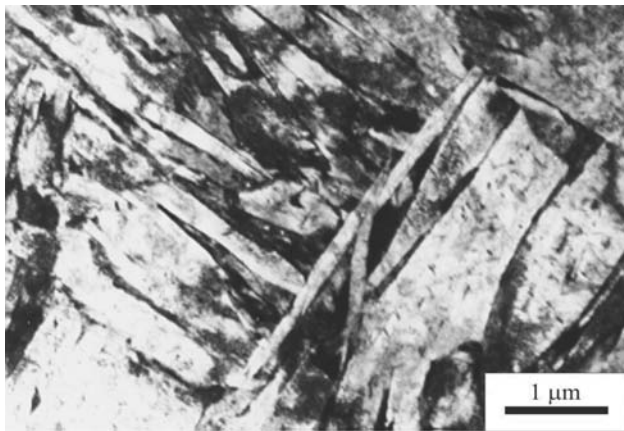


Fig. 34. Lath martensite structure with the non-uniform lath thickness (steel B)

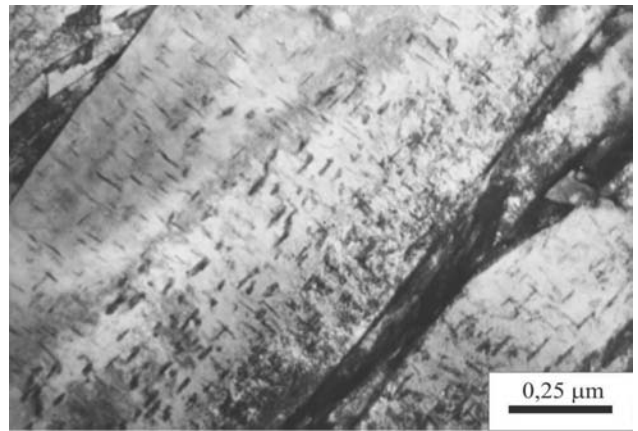


Fig. 37. Lath martensite structure with the dispersive cementite precipitations inside lathes; 900°C/60 sec/water (steel B)

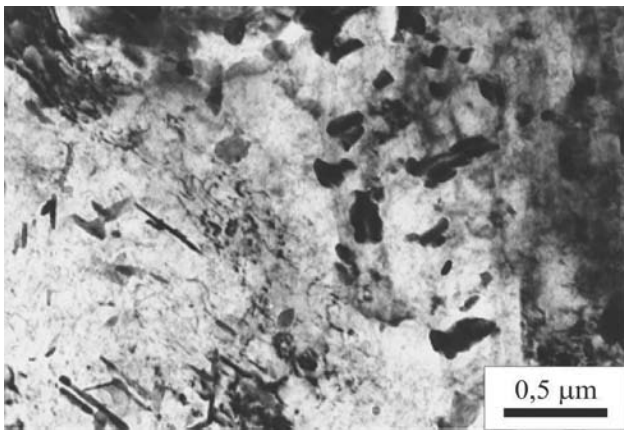


Fig. 35. Tempered martensite structure with the lamellar and coagulated Fe_3C particles in the segregation zone after tempering the steel B at the temperature of 650°C

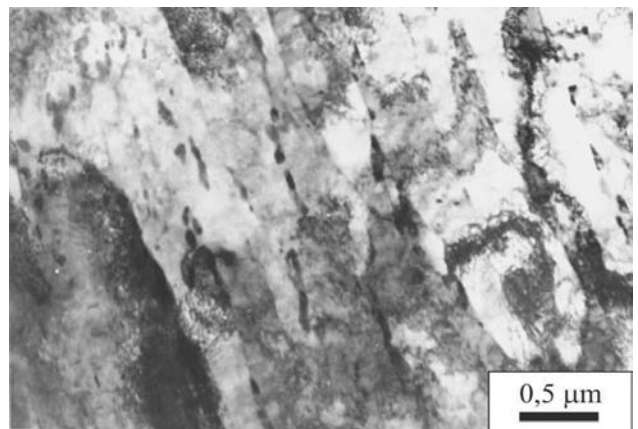


Fig. 38. Tempered martensite structure with the granular Fe_3C precipitations at the recovered austenite grain boundaries; 900°C/60 sec/water, tempering temperature 650°C (steel B)

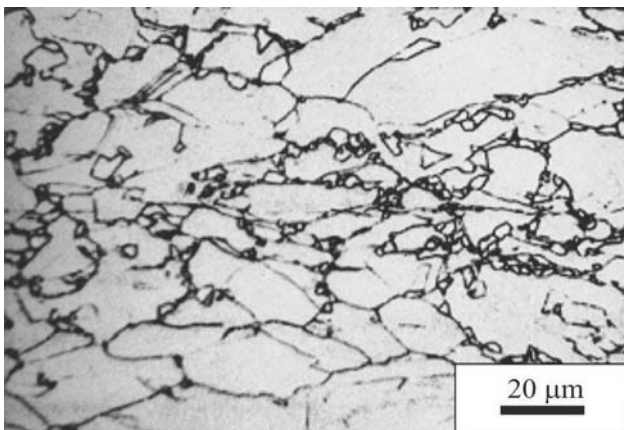


Fig. 36. Fine grains of the statically recrystallized austenite distributed at the grain boundaries of the statically recovered austenite; deformation temperature 900°C, $t_{0,5} = 60$ sec; (steel B)

The significant differences refer to crack resistance of the test segments of the plate made using both variants of thermo-mechanical working, and after conventional quenching and tempering of the steel. It turns out, from V-notched Charpy test results at the temperature of -40°C that energy KV, for both transverse specimens and for the longitudinal ones, increases with the increase of the tempering temperature. However, the energy values of breaking the specimens taken from the segment of the plate manufactured in the conditions of total recrystallization of plastically deformed austenite at intervals between the successive rolling passes and isothermal holding for the time $t_{0,5}$ at the temperature of finish-rolling prior to quenching, is nearly two times larger than the energy of breaking the specimens taken from the test segments of the plate made in the first thermo-mechanical working variant. The high value of the KV_t/KV_l ratio should be stressed - transverse specimens to the longitudinal ones, from 0.82 to 0.87, characterizing the anisotropy of plastic properties of the plate in the second variant of thermo-mechanical working, as

compared to the value of this ratio of 0.60 to 0.74 for the test segment of the plate in the first variant of this working, and 0.72 to 0.74 for the conventional heat treatment of the specimens taken from the test segment of the plate.

The investigations carried out revealed that the process of plastic deformation of steel in the temperature range from 1100 to 900°C in conditions close to heavy plate rolling is controlled by

the dynamic recovery flow, where the strain hardening decay of austenite between the successive deformation cycles occurs by the static recovery and static recrystallization. The nonuniform plastic metal flow is the reason for the origination of the deformation bands (shearing) in the plastically deformed austenite, spreading in the direction of the load, disappearing during the static recrystallization of the austenite.

Table 9. Effect of tempering on the mechanical properties and specimen breaking energy thermo-mechanically worked and traditionally quenched

Treatment conditions	Tempering temperature, °C	Test direction	YS _{0.2} MPa	UTS MPa	El %	RA %	KV _{-40 °C} J	KV _t /KV _l
900°C/60 s/water (no shield)	550	longitudinal	1038	1097	18.0	66.4	35.9	0.70
		transverse	1089	1116	17.3	62.3	25.0	
	600	longitudinal	1010	1078	18.7	66.3	42.9	0.74
		transverse	1060	1080	18.3	66.9	26.8	
	650	longitudinal	1005	1049	18.9	68.5	45.3	0.60
		transverse	1038	1045	18.6	67.0	32.8	
900°C/60 s/water (retention shield)	550	longitudinal	1036	1095	18.5	64.1	73.5	0.84
		transverse	1042	1112	17.2	63.0	61.4	
	600	longitudinal	1021	1074	18.9	66.5	103.0	0.87
		transverse	1025	1080	17.8	64.2	89.9	
	650	longitudinal	1000	1029	19.0	67.3	156.4	0.82
		transverse	1005	1036	18.4	66.9	128.9	
quenching 900°C/water	550	longitudinal	1016	1092	16.5	67.1	35.5	0.72
		transverse	1017	1095	18.6	69.1	25.7	
	600	longitudinal	996	1066	19.3	67.4	37.3	0.72
		transverse	995	1068	18.7	70.2	27.0	
	650	longitudinal	989	1039	19.8	68.4	38.0	0.74
		transverse	991	1041	19.0	70.5	28.0	

* average value calculated from three measurements

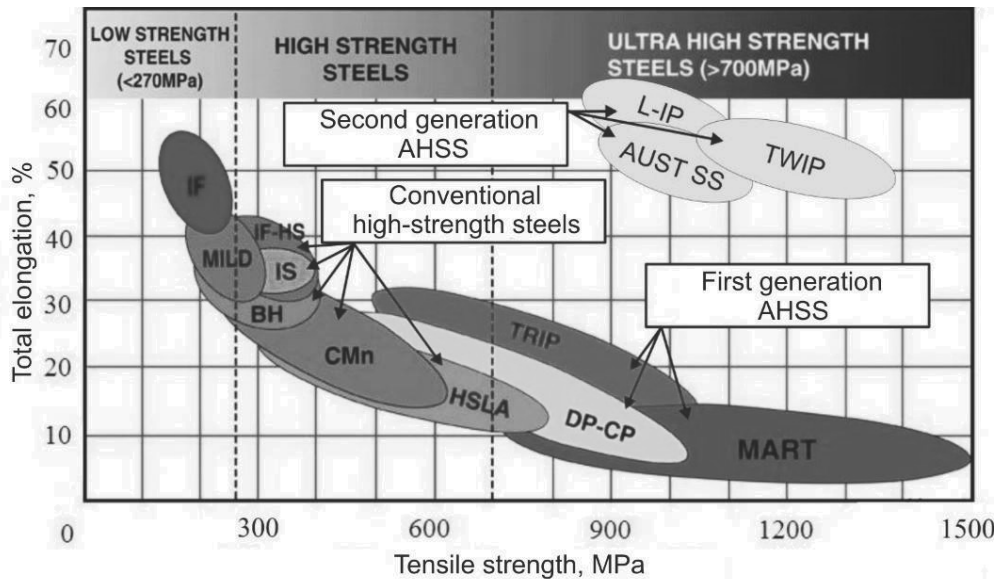


Fig. 39. Comparison of mechanical properties of conventional and advanced steels of the first and second generation (AHSS - Advanced High Strength Steels) used in the automotive industry: LSS (Low Strength Steels), HSS (High Strength Steels), UHSS (Ultra High Strength Steels) [35]

The time of total recrystallization t_R of the plastically deformed austenite of $\varepsilon = 0.2$ at a rate of 3 s^{-1} at the temperature of 1100°C is 50 sec, and increases to about 300 sec at the temperature of 900°C . However, the time $t_{0.5}$ - necessary to develop a 50% fraction of recrystallized austenite after plastic deformation of steel at the temperature of 1100°C , determined basing on the strain hardening decay kinetics - is 9 sec and increases to 60 sec at the temperature of 900°C . The time $t_{0.5}$ evaluated in this way does not lead to a development of the 50% fraction of recrystallized austenite, as the statical recovery of the plastically deformed austenite has a significant input to the decrease of the yield stress of hot steel.

If the intervals between the plastic deformation cycles are shorter than the time of statical recrystallization of the austenite then overlapping of the nonuniform flow of plastic metal in the successive deformation cycles leads to a development of the shearing bands in the austenite, enriched by segregation with carbon and chromium, that do not disappear during the isothermal holding of the products at the finish-strain temperature for the time $t_{0.5}$, nor during austenitizing the steel in its recommended conditions. These segregation bands, preserved in steel during quenching from its finish-rolling temperature, immediately after concluding its plastic working or after the time $t_{0.5}$, and also after conventional austenitization, are the reasons for development during the high tempering of the layered carbides, causing a lower crack resistance and increase of the anisotropy of the plastic properties of plates.

The deformation (shear) bands do not develop if the time between the consecutive deformation cycles makes total recrystallization of the plastically deformed austenite possible.

Adjusting the interval times between the deformation cycles in the multi-pass rolling process to the total recrystallization time of the plastically deformed austenite, by employing the retention shield, and the isothermal holding of the product at a finish-rolling temperature for the time $t_{0.5}$ prior to quenching, ensures manufacturing the plates obtaining after tempering in the temperature range 550 to 650°C the yield stress $YS_{0.2}$ from 1036 to 1000 MPa, UTS from 1095 to 1029 MPa, elongation from 18 to 19%, reduction of area from 64 to 67%, and the breaking energy $KV_{40^\circ\text{C}}$ of the longitudinal specimens from 73 to 156 J and the KV_T/KV_1 ratio (transverse to longitudinal specimens) from 0.82 to 0.87.

The investigation results obtained make it possible to work out the industrial scale technology for the weldable heavy plates with high strength and crack resistance at the lowered temperature, using the energy-saving thermo-mechanical method integrated with ladle metallurgy of the molten steel and continuous slab casting [33,34].

3.5. Steel sheets for the automotive industry

A variety of produced cars and other mechanical vehicles decides about the necessity to produce weldable hot and cold-rolled sheets and plates, characterized by diversified yield stress, tensile strength and formability, especially drawing and strain hardening dependent on microstructure. Steels for car sheets production used nowadays and to be used in the future are shown in Fig. 39.

The trend of the improvement of mechanical properties of steels used for car sheets production realized during the last tens of years concerns steels with A2 lattice matrix, among which the strongest possibilities of further improvement of strength and ductility correlation are noted for multiphase steels.

The development of mentioned steel groups, elaborated in order to meet the needs of automotive industry, in last thirty years, has started from **HSLA microalloyed steels** (High Strength Low Alloy), containing microadditions with strong chemical affinity to carbon and nitride. Hot-rolled sheets made of HSLA steel with fine-grained ferritic-pearlitic or ferritic microstructure, sometimes with particular portion of ferritic bainite, are manufactured by controlled rolling technology, usually in integrated lines with application of accelerated cooling from the temperature of the end of plastic working. Steels used for the production of these sheets consist of 0.05 to 0.07% C, 0.4 to 1.5% Mn, 0 or 0.4% Si, 0.03 to 0.15% V, up to 0.03% Nb and to 0.020% N. Sheets made of these steels 9 to 4 mm thick, or even thinner, are rolled in the temperature range adjusted to the kind and concentration of Nb, V and N microadditions introduced into the steel, reach the proof strength YS (0.2% offset) from 390 to 590 MPa, tensile strength UTS from 450 to 660 MPa, total elongation from 28 to 22% and uniform elongation from 20 to 14%, depending on the concentration of Mn, Si and V in the steel. Sheets with the listed mechanical properties are used for bus and truck chassis, wheel disks, motor bases, bumper reinforcements, suspension elements and others.

BH-type low-carbon steels (Bake Hardening) are no-silicon steels, consisting of less than 0.1% C, from 0.1 to 0.5% Mn, from 0.010% Nb, 0.005% Ti and to 0.001% B, often with a higher concentration of P, up to 0.030%, improving the strength of sheets and their normal plastic strain ratio. Sheets which are cold-rolled after recrystallization annealing in protective atmosphere are generally subjected to hot galvanizing in galvannealing type or to a different continuous line with accelerated cooling from the temperature of the process. Products obtained from these sheets, such as boot lids, doors, bonnets, suspension elements and others, drawn or cold-bended, are subjected to lacquering. During holding parts with applied automotive lacquer or organic covers at the temperature of 170 to 250°C , substantial hardening connected to the course of strain ageing is observed. This hardening is called bake hardening effect and usually causes an increment of yield stress and tensile strength of car parts by around 20 to 70 MPa. According to PN-EN 10325:2006, this increase is described as BHI (Bake-Hardening Index). This index is defined as a difference between yield stress of the steel subjected to 2% plastic strain and holding at the temperature of 170°C for 20 minutes and the stress necessary for plastic deformation of the steel to a value of 2% [35,36].

IF-type steels (Interstitial Free) with a slight concentration of interstitial elements, silicon and sulphur, consisting of up to 0.02% Nb, 0.03 to 0.07% Ti to 0.003% B and to 0.003% S, 0.2% Si and to 0.060% P, are used for the production of cold-rolled sheets, showing particularly strong susceptibility to deep drawing [36,37]. These steels require an accurately performed metallurgical process; Ti and Nb microadditions, and sometimes also B, are added in the final stage of off-furnace processing only in concentrations necessary to bind C and N into stable MX interstitial phases in the steel. Sheets made of IF steel, cold-rolled

in integrated lines and covered with zinc coating in hot galvanizing line with galvannealing method or different, are characterized by a strong susceptibility to deep drawing and in this state they reach a proof strength YS (0.2% offset) from 160 to 310 MPa, tensile strength UTS from 300 to 450 MPa, total elongation from 45 to 40% and uniform elongation from 26 to 22 [37]. The presence of dispersive TiN and NbC particles of interstitial phases causes strong hardening of drawn parts, expressed by a high value of the strain hardening exponent $n > 0.22$. Strong susceptibility to deep drawing and relatively low yield stress allow to use sheets made of these steels for parts particularly difficult in drawing, e.g. oil sumps and car bodies [35-37].

IS-type steels (Isotropic Steels) are a separate group of the discussed steels with a chemical composition similar to IF. They are characterized by a fine-grained ferrite microstructure with a properly configured fibrous structure being a consequence of controlled recrystallization rolling at the temperature to 700°C. In result of the applied technological operations, IS-type steels show isotropic properties during plastic strain [35].

Of particular interest are steel sheets with a high strength and required formability. This group consists of DP-type sheets (Dual Phase) with a ferritic-martensitic microstructure, TRIP-type (TRansformation Induced Plasticity) with ferritic-bainitic microstructure with retained austenite, which undergoes martensitic transformation during technological working and CP-type (Complex Phase) with properly selected portions of ferrite, bainite, martensite and retained austenite. A condition necessary to produce sheets with the described microstructure, in particular including a big enough portion of ferrite, deciding about their plastic properties and formability, is an adequate selection of the chemical composition of the steel assuring a form of supercooled austenite transformations diagram with $\gamma \rightarrow \alpha$ transformation put forward to short times and $\gamma \rightarrow$ bainite transformation in the opposite direction (Fig. 40). Translation of $\gamma \rightarrow \alpha$ transformation towards short times occurs with increasing concentration of Si, Al and P in the steel, and $\gamma \rightarrow$ bainite transformation in the opposite direction together with increasing amount of C, Mn and Cr [38-40].

Increased concentration of carbon causes a disadvantageous decrease of ferrite in the steel microstructure, therefore the value of this element in steels assigned for the production of multiphase sheets is limited to around 0.2% and Si and Al are added instead, slightly influencing the carbon concentration in the eutectoid mixture

DP-type steels (Dual Phase) generally consist of 0.1 to 0.2% C, 1.2 to 1.6% Mn, 0.3 to 0.6% Si, $\leq 0.030\%$ P and $\leq 0.030\%$ S and 0.1 to 0.15% V, $\leq 0.025\%$ N and 0.02 to 0.06% Ce, used in order to modify non-metallic inclusions decreasing the plastic anisotropy of sheets. Sheets with ferritic-martensitic microstructure made of these steels are produced nowadays using controlled rolling applying the accelerated cooling method. Depending on the chemical composition of the steel and portion of martensite located in forms of islands on ferrite grains boundaries, 3 to 6 mm thick sheets achieve a yield stress of about 300 to 380 MPa, tensile strength from 520 to 760 MPa, total elongation from 32 to 24% and uniform elongation from 26 to 18%. The presence of hard martensite islands causes strong hardening of cold-formed products made of these sheets, e.g. drawn truck wheel discs which present high fatigue strength.

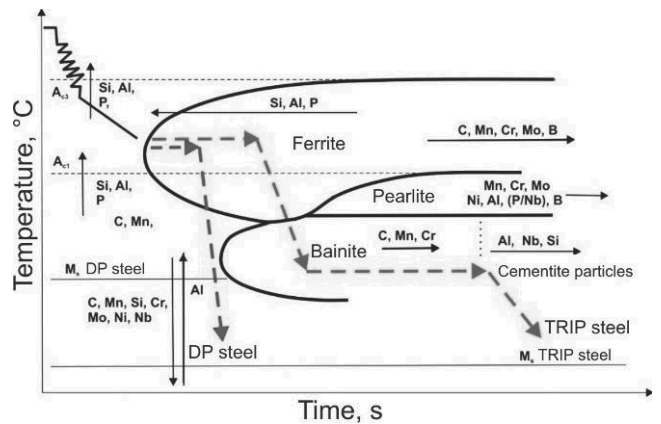


Fig. 40. Scheme of the influence of elements on the continuous cooling transformations diagrams with a course of the cooling for DP and TRIP-type sheets after hot rolling [39]

TRIP-type steels (TRansformation Induced Plasticity) are used for the production of sheets with a ferritic-bainitic-martensitic microstructure with retained austenite, which undergoes martensitic transformation causing grand hardening during cold forming of products. It was assumed initially that this type of steels can consist of: 0.2 to 0.4% C, 1 to 2% Mn and 1 to 2% Si [40]. Advantageous exploitation properties are achieved with a significant portion of ferrite which together with non-transformed austenite decides about their plastic properties. Therefore the carbon concentration in these steels doesn't exceed 0.2% and silicon can be partially replaced by Al and P, simultaneously the total concentration of Si + Al doesn't usually exceed 1.7% [41,42]. Silicon and aluminium, apart from influencing the forms of supercooled austenite transformation curves, cause a decrease of C activity in austenite during bainitic transformation in a temperature range from 300 to 450°C. The course of austenite transformation in the mentioned conditions leads to ferritic bainite formation with small participation of granular or fine Fe_3C carbides, with simultaneous enriching with carbon of non-transformed austenite with increasing stability along with lowering the temperature of the transformation. It's a result of tending of the arrangement of thermo-dynamical equilibrium between bainite and austenite enriched with carbon up to a particular concentration limit, temperature dependent. If this austenite doesn't yield carbon impoverishment, e.g. as a result of carbides precipitation, then bainitic transformation will be stopped [43-49]. Limiting carbon concentration in non-transformed austenite decreases together with transformation lowering the temperature, because the presence of graphitizing elements, i.e. Si and Al slows down nucleation and increases cementite precipitations. The rest of carbon-enriched austenite keeps its stability during slow cooling of sheets from the temperature of isothermal transformation to room temperature and undergoes martensitic transformation during cold plastic deformation of sheets.

TRIP-type sheets are made by cold rolling with successive austenitizing at the temperature slightly higher than that of A_{c1} of steel and isothermal quenching in the bainitic transformation temperature range, i.e. slightly higher than M_s of steel up to

around 450°C and cooling to room temperature (Fig. 41). However, techno-economic interest concerns the production of these sheets using the thermo-mechanical processing method, consisting in controlled hot rolling of the steel and controlled cooling of the sheets in $\gamma \rightarrow \alpha$ transformation temperature range as well as direct isothermal quenching at the temperature higher than M_s of steel, allowing the occurrence of bainitic transformation for the time needed to enrich non-transformed austenite in carbon reaching limiting concentration and cooling of sheets. In the last case, sheets are made of TRIP-type steels with Ti, Nb and V microadditions, and the temperature range of controlled hot rolling of sheets is selected adequately to the kind and concentration of microadditions made to the steel.

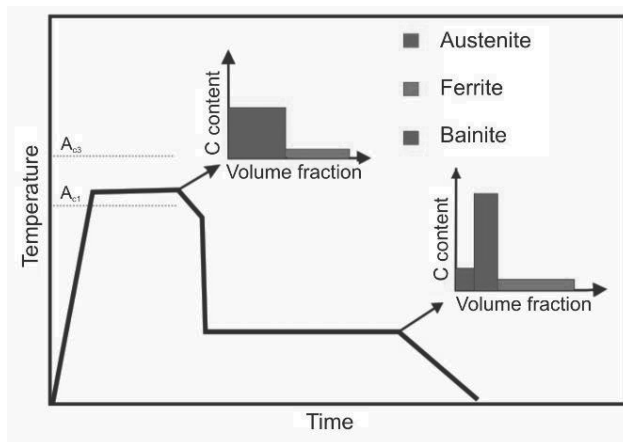


Fig. 41. Scheme of the heat treatment of TRIP-type steel after cold rolling together with carbon concentration changes for particular structural constituents [44]

Basing on the analysis of Fig. 39, it can be stated that the group of steels with a selected complex way multiphase microstructure CP-type with highest possibilities concerning improvement of mechanical and plastic properties is one of the groups of steels used at present in automotive industry. In case of tensile strength, they yield only to steels with a martensitic microstructure. However, as distinct from them, they have also good plastic properties with total elongation from 15 to 20%. It results in a very profitable product of UTS·TEL, which is a measure of the formability and the ability to absorb the energy in case of road collision. High mechanical properties are the consequence of the presence of a large fraction of hard phases in the microstructure of these steels, i.e. bainite and martensite, at the cost of ferrite and retained austenite portions responsible for the particular level of steel ductility. Obtaining the assumed multiphase microstructure with adequate portions of individual phases requires a precise selection of the chemical composition of the steel and monitoring of hot-working conditions and successive several-stages cooling from the end of rolling. Smaller participation of ferritic matrix is a result of slightly higher carbon concentration in these steels when compared to concentrations used in the case of TRIP-type steels. Apart from this, participation of α phase can be formed by properly selected course of cooling after the end of rolling. Smaller portion of retained austenite

derives from a shorter time of steel holding in the conditions of isothermal bainitic transformation. As a result, retained austenite doesn't reach full thermal stability during further cooling to room temperature, because its temperature M_s is higher than room temperature. For this reason, a part of γ phase undergoes martensitic transformation, essentially improving the mechanical properties of these steels, yet at the cost of decrease of steel ductility. The residual part of retained austenite undergoes martensitic transformation during cold forming, similarly to TRIP-type steels.

The higher concentration of carbon in these steels as well as required smaller portion of retained austenite, allows a decrease of the silicon concentration, responsible for proper stabilization of γ phase during sheet holding in the bainitic range. It's profitable because silicon concentration of around 1.5% - commonly used in steels with TRIP effect - was the cause of the lack of adequate wettability by liquid zinc in the process of hot galvanizing, which did warrant corrosion resistance of sheets. It was found that the concentration of silicon effectively prevents cementite precipitation during isothermal holding of steel in the bainitic range equal to around 0.8% and the other, necessary for conserving required portion of retained austenite in the microstructure, can be complemented by using Al or P. Aluminum also delays precipitation of cementite, accelerates bainite formation and rises the temperature of martensitic transformation.

In the last few years an increased interest in **high-manganese austenitic steels** has been observed (Fig. 39). High-manganese austenitic steels consist of 15 to 30% of manganese, from 0.02 to 0.1% of carbon and around 3% of aluminum and 3% of silicon added, among others, in order to decrease the density which for this group of steels is equal to about 7.3 g/cm³. These steels achieve profitable mechanical properties, i.e. UTS = 600-900 MPa, YS (0.2% offset) = 250-450 MPa, UEI = 35-80% which strongly depends on the chemical composition, especially the concentration of Mn. The role of silicon and aluminum is solution hardening of the steel, and carbon is an element that stabilizes austenite. Apart from high mechanical and plastic properties they have a particular ability to absorb significant amounts of energy in conditions of plastic strain with high rate and formation of parts with complex shape.

In the case of manganese concentration equal from 5 to 10%, steel shows the microstructure of ferrite and austenite. Together with further increase of manganese concentration in range from 20 to 30%, a total stabilization of austenite occurs. In the case of manganese concentration below 25% it's possible to use TRIP effect (Transformation Induced Plasticity) consisting in steel hardening in the consequence of $\gamma_{A1} \rightarrow \epsilon_{A3}$ or $\gamma_{A1} \rightarrow \epsilon_{A3} \rightarrow \alpha'_{A2}$ martensitic transformation occurring during cold forming. Then, high strength of steel, around 900 MPa at elongation of around 40% is achieved. Martensite ϵ with A3 hexagonal lattice is formed during plastic strain only when the stacking fault energy SFE of austenite at room temperature is lower than 20 mJ/m². An addition of aluminium into steel increases SFE and austenite stability which leads to suppressed influence on martensitic transformation. While the addition of silicon decreases SFE and allows $\gamma \rightarrow \epsilon$ transformation. The chemical composition of steels should also take into account the fact that the stacking fault energy depends on temperature and increases along with its rising.

Moreover, the TRIP effect is present in metastable austenite when $\Delta G^{\gamma \rightarrow \epsilon}$ free enthalpy is negative and equal ≤ -220 J/mole depending on chemical composition. Transformation of metastable austenite into martensite with A3 hexagonal lattice during cold plastic deformation can be considered similarly a allotropic transformation of cobalt. The change of A1 lattice into A3 hexagonal lattice can occur through a change of plane sequence with thick atoms packing. Formation of stacking fault in every second plane (111) of A1 lattice causes the sequence change of these planes from ABCABC... type - adequate for this lattice sequence ABAB... - of (0001) planes corresponding to A3 hexagonal lattice. This change of plane sequence with thick atoms packing can be replaced by displacement of Shockley's partial dislocation $a/6 \langle 211 \rangle$ in every second plane (111) of A1 lattice. Shearing that occurs in (111) plane will leave this plane free from deformation and twisting, which makes it a crystal habit plane.

When the manganese concentration in the steel exceeds 25%, the stability of austenite during plastic strain is maintained. The strength achieved in such a situation is equal to around 650 MPa and elongation up to 80%, essentially contributing in plastic strain through mechanical twinning mechanism - TWIP effect (TWinnig Induced Plasticity). Deformation twins formed in austenite substantially increase the strain hardening rate being an obstacle for a swift of dislocations, like grain boundaries. Steels with TWIP effect show uniform elongation up to fracture and in particular they possess the ability to absorb the energy over twice bigger when compared to conventional steels used for deep drawing. Moreover, these steels present high impact strength, apart from testing temperature in the wide range from -196°C to 400°C . In the paper it was found that along with the increase of manganese concentration from 15 to 25%, the tensile strength from 930 to 610 MPa decreases and increase uniform elongation from 46 to 88%. Further increase of Mn concentration doesn't cause any increase of plastic properties of the steel, which are the highest for steels with 3% Al and 3% Si. In the range of small deformations, up to around 0.15, steels are hardening in a similar way. When the actual strain is exceeded by around 0.15 steel with 15% Mn with initial $\gamma + \alpha + \epsilon$ the microstructure is subjected to transformation of austenite into ϵ and α' martensite. It's manifested with much stronger strain hardening when compared to steel with 25% of Mn, subjected only to intense mechanical twinning.

An essential influence on deformation twins and ϵ and α' martensite formation in the microstructure of high-manganese austenitic steels, besides chemical composition, exert temperature, reduction and strain rate.

In the following sections of the paper a few examples of investigations on steels for the automotive industry are described. They concern mainly hot-working resistance, phase transformations during multistage cooling and structural aspects of thermo-mechanically processed or cold-worked steel sheets.

Dual phase steels

The chemical composition of the investigated low-carbon structural steel is given in Table 10. The steel contains increased concentrations of manganese, silicon, molybdenum, titanium and a boron microaddition to improve the hardenability of steel and a

small concentration of impurities. The established heat treatment conditions are schematically shown in Fig. 42.

Table 10.

Chemical composition of the investigated steel

Chemical composition, wt. %											
C	Mn	Si	Ni	Cr	Mo	Ti	P	S	B	Al	N
0.09	1.50	0.26	0.07	0.06	0.14	0.113	0.014	0.009	0.003	0.029	0.012

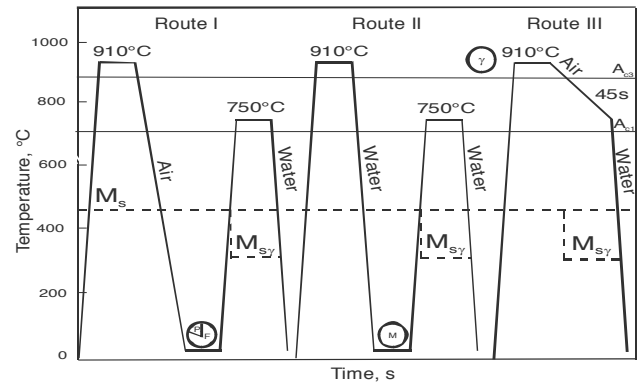


Fig. 42. Schematic representation of the heat treatment of DP-type steel according to routes I, II and III

The specimens subjected to the heat treatment routes shown in Fig. 42 have various initial structures. In two of three cases the steel was annealed in a $\gamma + \alpha$ range and water quenched. The soaking temperature 36°C higher than A_{c1} of the steel was selected on the basis of initial investigations concerning the influence of austenitizing temperature on a martensite fraction of the quenched steel. The authors of the works suggest, that DP-type steels obtain the best mechanical properties for the α' fraction from 20 to 25%. In the third variant the specimens were initially air cooled for 45s to a temperature of 750°C and then water quenched. The cooling time of the specimens in air was determined experimentally based on thermocouple measurements.

The specimens for structure investigations were conventionally prepared and etched using nital and a saturated aqueous solution of picric acid with CuCl_2 addition. The Leica MEF4a light microscope was used for observations of the obtained structures after the heat treatment and to evaluate fractions of ferrite and martensite using the automatic image analyzer. Thin foils were investigated in the JEM-200CX transmission electron microscope at an accelerating voltage of 120kV. The specimens with a diameter of 6mm and a gauge length of 30mm were subjected to the tensile test using the Zwick Z/100 machine. A various initial structure of the investigated steel aimed at the determination of its influence on morphology of structural constituents in a final dual phase structure. The initial structures of the specimens subjected to the heat treatment according to three routes in Fig. 42 were: ferrite with some fraction of pearlite, low-carbon martensite and primary austenite with a grain size of about $12 \mu\text{m}$ - respectively for the routes I, II

and III. It was found that the martensite morphology exhibits substantial differences dependent on the initial structure of the steel. The specimens heat-treated according to the route I have a ferritic matrix and the martensite is located on the boundaries of the α phase as an irregular envelope (Fig. 43). The location of martensite is strongly dependent on the distribution of the austenite formed due to carbon enrichment of the boundary-zones of ferrite connected with a decomposition of pearlite grains. During annealing of the investigated steel at a temperature of 750°C the privileged diffusion of carbon on the boundaries of the α phase occurred. Underhardening from an increasing annealing temperature leads to an increase of the martensite volume fraction, keeping the network distribution of this phase on grain boundaries of the α phase (Fig. 44). The martensite fraction in the steel quenched from a temperature of 750°C equals about 22% and increases to 40% after quenching from a temperature of 810°C. A result of the increase of martensite fraction together with increasing the underhardening temperature is unfavourably decreasing the carbon concentration in austenite and increasing the M_s temperature of this phase. For 22% volume fraction of martensite, the carbon concentration in the γ phase equals about 0.47% and the M_s temperature of this phase about 290°C. The austenite enriched in carbon increases the hardenability of steel, too. An additional influence in improving the hardenability of steel display alloying elements dissolved in the austenite, i.e. manganese, silicon, molybdenum, nickel, chromium and especially a boron microaddition. Its positive influence is guaranteed due to titanium fixed the nitrogen entirely. It is a very important element of manufacturing the DP-type steels. The underhardening of austenite and the smaller precipitation hardening by martensite islands are obtained after the quenching of steel from a temperature of 810°C. The carbon concentration in the austenite is about 0.2% and M_s increases to 390°C, at this temperature. A different morphology of the structure have the specimens quenched twice, i.e. from a temperature of 910°C and 750°C. In this case, during heating the steel to an underhardening temperature of 750°C, the nucleation of austenite occurs mainly on the boundaries of martensite laths formed after primary quenching of the investigated steel from a temperature of 910°C. Here, the martensite fraction in the form of a network on grain boundaries of the α phase is not large. The predominating martensite fraction occurs mainly as thin fibres located in a surroundings of grain boundaries and inside ferrite grains (Fig. 45).

Moreover, in the surroundings of martensite, especially at a boundary zone of large grains of the α phase, small grains of the recrystallized ferrite can be identified. They are the result of plastic deformation connected with volume changes accompanying the martensite transformation.

Apart from the morphology differences after the heat treatment of steel according to the routes I and II a difference in grain size of ferrite is observed. The more fine-grained structure has a steel heat-treated according to the path II. The average grain size of the α phase is in this case about 7 μm , while for the route I it is about 3 μm larger. This is due to the increased number of places convenient for the nucleation of ferrite and also a partial course of the recrystallization of the specimens with an initial structure of martensite. After underhardening for both variants of the heat treatment the martensite fraction is near the same. For the route I equals about 22% and for the route II it is about 2%

higher. Such a fraction of martensite and ferrite being a complement to 100% influences the enrichment of austenite in carbon and also lowers the M_s temperature of austenite to $M_s = 286^\circ\text{C}$ and 297°C , respectively for the routes I and II.

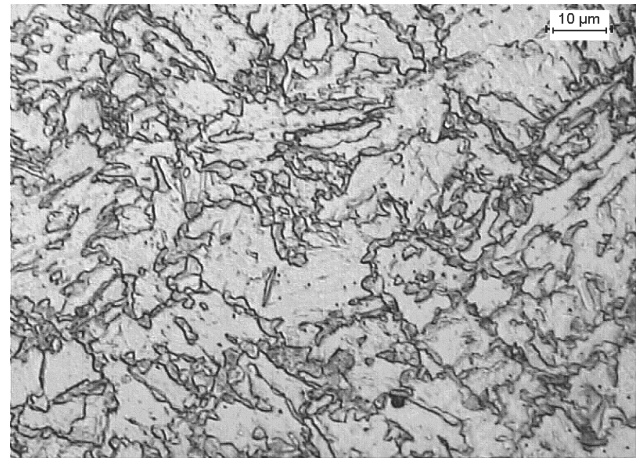


Fig. 43. Ferritic-martensitic structure of the steel quenched from a temperature of 750°C

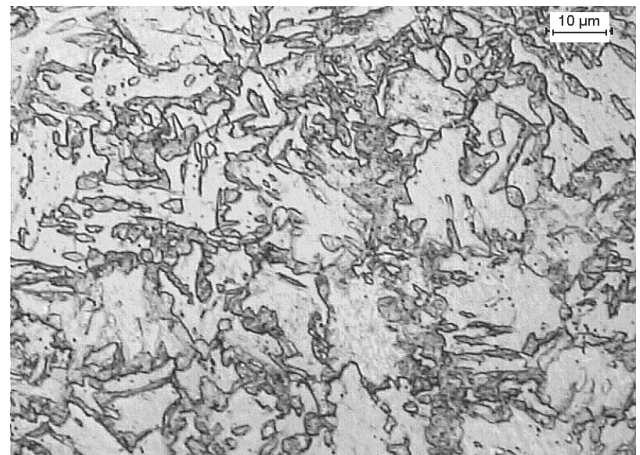


Fig. 44. Ferritic-martensitic structure of the steel quenched from a temperature of 810°C

A different type of the structure specimens was found after the realization of the heat treatment according to the route III (Fig. 46). In this case after austenitizing at a temperature higher than A_{c3} for a steel the specimens were initially air cooled for 45 s or 35 s to a temperature of 750°C and 785°C - respectively, and next water quenched in order to transform the austenite fraction of various carbon concentrations to martensite. This variant of heat treatment can simulate cooling after rolling, e.g. during thermomechanical processing. As distinct from other heat treatment routes, martensite does not occur as a network or thin fibres but in form of islands of diversified size. They are located on the grain boundaries of the α phase. The optimum fraction of martensite averaging 20% occurs after air cooling of the specimens for 45 s (Fig. 46).

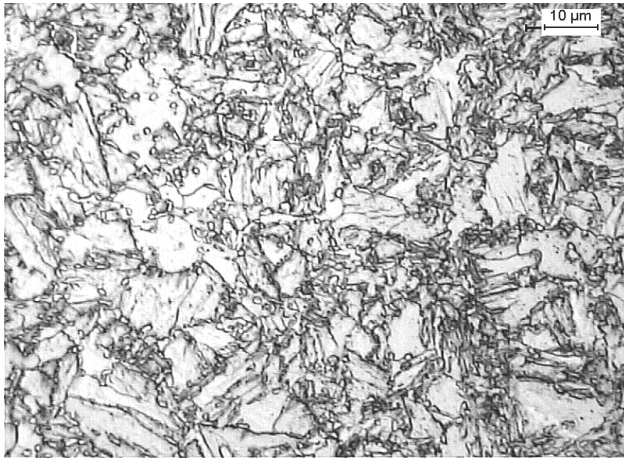


Fig. 45. Ferritic-martensitic structure of the steel twice quenched from the temperatures of 910°C and 750°C

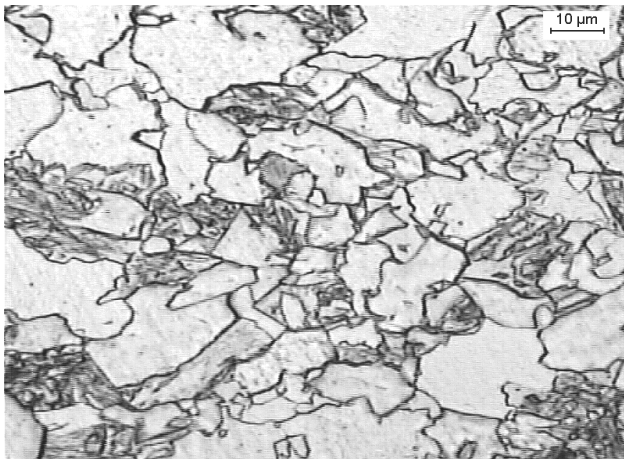


Fig. 46. Ferritic - martensitic structure of the steel quenched from a temperature of 750°C after previous air cooling for 45 s from 910°C

The optimum cooling time of specimens in air equaling 45 s (route III) was not selected but only based on the comparable martensite fraction to another variants.

A reduction of the cooling time increases the fraction of diffusionless products of undercooled austenite, but at a lower carbon enrichment of the austenite. It lowers the hardenability of steel (austenite depleted in carbon) and can lead to a transformation of some part of austenite to undesirable bainite.

In order to investigate in detail the structure of DP-type steels investigations of a thin foil structure were carried out by the use of the transmission electron microscopy. Fig. 47 shows a transmission electron micrograph of the structure of water quenched steel following austenitizing at a temperature of 750°C. The martensite has a lath morphology and occurs as regular islands, surrounded by ferrite of a high dislocation density. The dislocation density decreases with water quenched increasing distance from the islands (Fig. 47). The increased dislocation

density in ferrite is a result of increasing the volume of diffusionless products of austenite transformations for the carbon concentration in the γ phase about 0.47%. A lack of the presence of bainite in the structure of the steel proves the good hardenability of the steel. Apart from the enrichment of the austenite in carbon, a substantial effect on the increase of the hardenability of steel have Mn, Si, Cr, Mo, first of all boron dissolved in the solid solution. A positive influence of boron is possible due to the effective shield by titanium, fixing the total nitrogen in the steel as TiN. It was also found that between martensite laths the films of retained austenite can occur locally (Fig. 48a). This phase is a desirable constituent of the structure of DP-type steels. Its transformation into martensite proceeding during technological forming of the elements can positively increase the strengthening of steel, as in the case of the steels showing the TRIP effect.

On the basis of the tensile test, it was found that the morphology of heat-treated DP-type specimens has a fundamental effect on mechanical properties of steel (Table 11). The best connection of strength and ductile properties has the heat-treated steel according to the route II, i.e. quenched from a temperature of 750°C following primary quenching from a temperature of 910°C.



Fig. 47. Martensitic island on the boundaries of three ferrite grains of high dislocation density

For this variant of heat treatment, the yield point is 516 MPa, tensile strength 800 MPa, total elongation about 20% and uniform elongation 16%. Steel quenched from a temperature of 750°C with an initial ferritic-pearlitic structure has lower values of the yield point and tensile strength by about 50 MPa as well as total and uniform elongations by about 4%, i.e. TEI = 15% and UEI = 12%. The highest strength properties has steel air cooled for 45 s from a temperature higher than A_{c3} to a temperature of 750°C. It leads to a transformation of about 79% austenite into ferrite and a residual fraction of the austenite enriched in carbon to $C_{\gamma} = 0.5\%$ into martensite. The yield point of the specimens cooled according to this route achieves 635 MPa and tensile strength about 1000 MPa. However, the high strength properties influence the ductile properties, which are slightly lower. They are comparable to the properties obtained for the specimens

cooled according to the route I (TEI = 15%, UEI = 12%). The shortening of the time for the realization of the $\gamma \rightarrow \alpha$ transformation from 45 s to 35 s leads to an increase of the martensite fraction to 32%. As a result of that, the strength properties ($YS_{0.2} = 690$ MPa, UTS = 1180 MPa) increase but the ductile properties decrease.

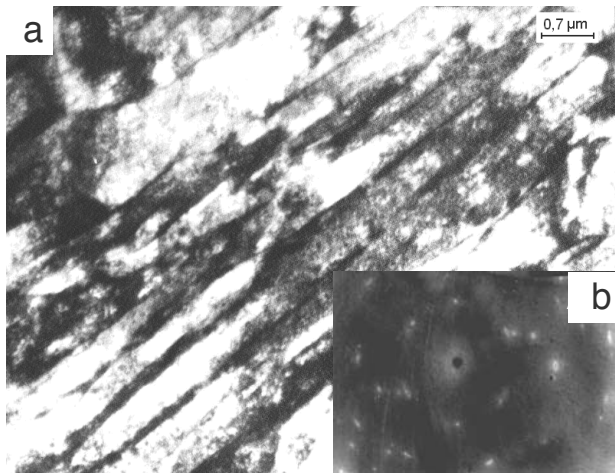


Fig. 48. Structure of lath martensite with the retained austenite located between martensite laths, a) light field, b) diffraction pattern

Table 11.
Mechanical properties of the investigated DP-type steel

	Value	UTS, MPa	YS, MPa	UEI, %	TEI, %	YS/UTS	UTS · U _{El} , MPa · %
Route I	Mean	751	501	12.3	15.0	0.66	9282
	±	12	64	3.5	3.0	0.08	1202
Route II	Mean	798	516	16.0	19.3	0.65	12781
	±	14	54	1.1	1.2	0.06	1009
Route III	Mean	998	635	11.7	14.6	0.63	11625
	±	35	25	1.1	1.1	0.02	1180

The DP-type structure produced under investigated conditions of the heat treatment ensures a favourable value of the $YS_{0.2}/UTS$ ratio, equaling 0.58 to 0.67. It indicates an intensive course of strain hardening of steel after exceeding the yield point.

The carried out investigations showed that the mechanical properties of DP-type steels depend on many factors. They are: ferrite and martensite morphology, properties of these phases and their volume fractions. The temperature-time conditions of the heat treatment were chosen in order to get a comparable fraction of martensite for all the variants. Taking it into account, the carbon supersaturation for all of the variants is comparable too. In this case, the dominant factors affecting the properties of the steel are: the martensite morphology and ferrite properties. To determine the

properties of the α phase, its behaviour during plastic deformation caused by the martensitic transformation has to be taken into account. The difference in the strength of ferrite is most probably the main reason of the difference in the strength of specimens with different structures. The network structure (route I) was produced by martensitic transformation of the austenite located along grain boundaries of coarse-grained ferrite. The limitation of the transformation under these conditions is much weaker. It results in a much smaller density of dislocations in the α phase and the lowest strength properties.

The authors of suggest, that the deformation of ferrite in the DP-type structure with the martensite in the form of fine fibres is much larger. There is a dislocation substructure in ferrite, increasing the strength properties of this phase. Additionally, a factor deciding about the highest strength properties of the specimens heat-treated according to the routes I and III is the fine-grained structure of ferrite. The grain size of this phase is about 3 μm (route II) and 1.5 μm (route III) smaller in comparison to the ferrite grain size of heat-treated specimens according to the route I.

Mechanical properties of steels with various kinds of DP-type structure indicate their different suitability to metalforming operations, especially sheet-metal forming.

4. Conclusions

Constructional microalloy steels are still the basic material used in the construction of various kinds of welded structures, machines and transport installations, as well as technological products applied in conditions of varying external loads and temperature.

Conventional highly strengthening steels of the type HSLA (High Strength Low Alloy) with a fine ferritic-pearlitic, bainitic or martensite tempered structure, previously used in the production of thick steel sheets, are now - in order to meet the actual requirements of the automotive industry - enriched as modern multiphase constructional steels of the first and second generation type AHSS (Advanced High Strength Steels).

Nowadays, beside HSLA steels generally steels of the type BH (Bake Hardening) are applied, strengthened by means of cold-work ageing, as well as IF (Interstitial Free) and IS (Isotropic) with good deformability to press forming. The most effective solution, ensuring properties of the steel sheets required by the automotive industry, and due to the possibilities of producing them in the conventional production lines, provide the main constructional microalloy steels with a multiphase structure. These comprise ferritic-martensite Dual Phase steels (DP), ferritic-bainite steels with retained austenite of the type TRIP (Transformation Induced Plasticity), strengthened in the course of technological forming due to the martensite transformation of the γ phase, applied in the production of elements displaying a high ability of absorbing energy in the case of crush of automotive vehicle. Complex Phase steels (CP) with a complex share of various phase, and also steels with a martensitic structure of the type MART (Martensite Steel).

The quoted data indicate that microalloy steels are a new generation of weldable constructional low-alloy steels, which can

be applied to produce hardware characterized by their high strength and resistance to cracking, making use of energy-saving techniques. An important achievement is the mass production of hardware with stable utility properties in integrated production lines with a controlled metallurgical process and metallurgical processing, the conditions of which can be adapted to the kind and concentration of the microadditives and alloy components introduced into the steel. This permits to reduce the number of installations connected with a considerable reorganization of the steelworks. The financial outlay for this purpose is refunded in a short time thanks to the production of goods required for the erection of modern structures, machines and appliances in a rationally selected mass and a prolonged time of exploitation. This indicates also possibilities of reducing the production of steel in its raw state as well as energy and raw materials inevitable for the production of this material.

An analysis of selected problems concerning constructional microalloy steels, carried out for the purpose of studies, does not exhaust the vast and complex problem of the development and applications of these steels, as required to meet the demands of the market and the necessity of improving the structure of conventional and newly designed and produced steels. Such an analysis indicates, however, distinctly the necessity of investigations in this line and scientific achievements, as well as their contribution to the design of modern materials and the improvement of metallurgical techniques and processing.

These problems were analyzed fundamentally and fully appreciated also by the late Prof. J. Adamczyk, who stimulated most of the quoted investigations, taking active part in their realization. His perspective vision of the development of investigation concerning this problem is presently continued by a team of his former co-workers. A further determinant of the aim of the perspective investigation is the criterion of the quality and technical usability of the developed techniques of production. The up-to-date applicability of the erected structures and the rational application of steel depends on the actual meeting of the requirements concerning the utility of the products, including their geometry, the dimensional tolerance, the state of surface and hers protection.

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