

Development of the microalloyed constructional steels

J. Adamczyk *

Division of Constructional and Special Materials Engineering,
Institute of Engineering Materials and Biomaterials, Silesian University of Technology,
ul. Konarskiego 18a, 44-100 Gliwice, Poland

* Corresponding author: E-mail address: jan.adamczyk@polsl.pl

Received 15.11.2005; accepted in revised form 31.12.2005

Materials

ABSTRACT

Purpose: The aim of the paper is to present the development and the technical importance of the HSLA-type microalloyed constructional steels (High Strength Low Alloy) in selected industry branches.

Design/methodology/approach: A mechanism of the interaction of Nb, Ti, V and B microadditions introduced into the steel on mechanical properties of selected metallurgical products with the fine-grained structure, forming under the properly selected hot-working conditions with the of (M-Nb,Ti,V; X-N,C) interstitial phases MX-type is discussed.

Findings: The requirements concerning the metallurgical process, continuous casting of steel and the necessity of adjusting the hot-working conditions to the precipitation kinetics of the dispersive MX phases particles in austenite, in the controlled rolling or thermo-mechanical treatment processes are indicated.

Research limitations/implications: The continuation of investigations concerning the thermo-mechanical rolling of automotive sheets with the multiphase structure of microalloyed steels is planned.

Practical implications: The indicated data, coming also from the own research, are of practical use in relation to manufacturing the metallurgical products and machine elements of high strength and crack resistance, also at low temperatures.

Originality/value: The results contribute to the development of rolling and forging technologies of the microalloyed steel (HSLA) elements produced using the energy-saving thermo-mechanical treatment methods.

Keywords: Metallic alloys; HSLA steels; Metallurgical process; Continuous casting; Controlled rolling

1. Introduction

Mass use of steel as the main constructional material is connected directly with development by Bessemer in 1856 of smelting of the big mass of steel by blowing the molten pig-iron in the converter with the compressed air stream. Steel scrap accumulating over time, coming from the worn out constructions and devices, and from the technological waste, as well as depletion of the iron ore resources with low concentrations of P and S impurities decided the need to develop the new steel smelting methods. The open-hearth process, or Siemens-Martin process, was developed by Martin in 1865, consisting in steel smelting from pig-iron in the solid state and steel scrap in the

open-flame fired furnace using the large amount of fuel, indispensable for melting the charge and for maintaining the course of reaction between the molten metal with slag and combustion gases directed onto its mirror. Return to the converter steel smelting from pig-iron with high P concentration occurred due to employing the base refractory materials for furnace lining and the slag-forming materials by Thomas in 1877. Steel smelted in these processes and with the varying reduction degree was used commonly for riveted constructions.

The idea of joining the steel elements by welding using the electric energy and by the heat of chemical reactions of the combustible gases and thermite originated at the end of XIX century, after bonding used since long time. The vision of the considerable

construction manufacturing cost reduction connected with welding substitution for riveting decided the intensive development of welding technologies in the beginning decades of XX century, especially the electrical welding. However, the steels manufactured at that time were often the cause for the failures of welded constructions, especially painful during the World War II. Employment of welding instead of riveting of the hulls caused a big number of averages of the *Liberty* and *Victory* transport ships or their engulfments resulting in huge material losses and casualties.

The losses suffered decided the necessity of carrying out the thorough steel tests explaining the causes of the catastrophes that had happened. The examinations revealed that the constructional steels belong to the semi-short materials, i.e., that above a certain critical point T_{pk} they behave like the ductile materials - cracking after their initial significant plastic strain, and below that point - as short materials and crack at the minimal plastic strain. The T_{pk} temperature is not the material constant and rises unfavourably along with the concentration increase of O, N, H, and P, as well as of Cr and Si in steel, and also of C - recognised as the main constituent bringing about strengthening of the steel products. Elements like Ni and also Mn in concentration of up to about 2% have the opposite effect on T_{pk} .

However; S has a small affect on T_{pk} , but fixing the iron in FeS it causes hot-shortness of steel, and forms the nonmetallic MnS precipitations with manganese present in the steel, subject to plastic strain at elevated temperatures, being the cause of the anisotropy of the plastic properties of flat products and deterioration of steel weldability [1-8]

Results of these investigations revealed that the cause of the catastrophes that had occurred was the low metallurgical purity of steel, expressed by the increased concentrations of P, S, and N, as well as the high portion of the deformable non-metallic inclusions, employed auxiliary materials and adverse welding conditions, giving the possibility of oxidation and nitriding the welds and heat affected zone, and also development of the coarse grained structure of the welded joints [5-11].

Improvement of the metallurgical purity of steel resulted from development of the oxygen-blown converter in the fifties of the last century and high capacity electric arc furnaces used earlier for alloy steels smelting. Smelting of constructional steels in electric furnaces gains the growing importance due to ecological reasons.

In further research it was determined that the lower yield point limit value σ_y of the polycrystalline iron grows along the decrease of the average grain size, according to the Hall-Petch relationship [12,13]:

$$\sigma_y = \sigma_0 + k_y d^{-1/2}, \quad (1)$$

where:

σ_0 - constant corresponding to the stress value after approximation of the σ - $d^{-1/2}$ relationship to the value of $d^{-1/2} = 0$,

k_y - constant, representing blocking effect of grain boundaries on movement of dislocations at the initial stage of the plastic deformation.

The fine-grained structure causes also increase of the brittle cracking (cleavage fracture) of iron and low-carbon steels, described by formula [14, 15]:

$$\sigma_f = \sigma_{of} k_f d^{-1/2}, \quad (2)$$

where:

σ_{of} and k_f - experimentally determined constants, where $k_f > k_y$ from formula (1) has the following value:

$$k_f \geq [6 \pi \gamma G / (1-\nu)]^2, \quad (3)$$

where:

γ - surface energy of a crack, G - torsional modulus, ν - Poisson's ratio.

The ductile-to-brittle transition temperature T_{pk} is given with the formula [15]:

$$T_{pk} \approx \ln d^{-1/2}, \quad (4)$$

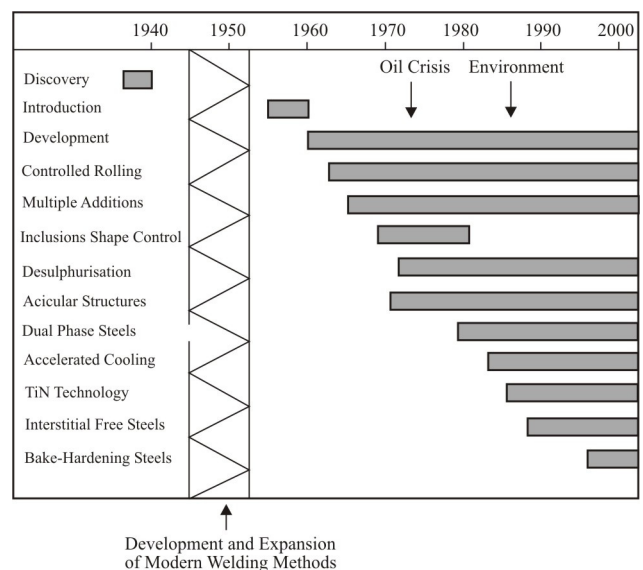


Fig. 1. Development of the HSLA steels from 1940 to 2000 [16]

Relationships (1) to (4) show that the fine-grained structure allows the significant improvement of the mechanical properties and crack resistance of the steel at the simultaneously lowered ductile-to-brittle transition temperature of this material from the ductile state to the brittle one. Steels manufactured at that time and steel industry processing technologies did not offer such possibilities. The unexpected possibility of manufacturing the fine-grained steel products appeared due to the incidental introducing of the low Nb concentration in the low-carbon steel for strip rolling. Further research, whose results have been widely published and included in proceedings of the international conferences [16-20] have resulted in development of the readily welded micro-alloyed steels of the HSLA (High Strength Low Alloy) type steels, whose development in years 1940 to 2000 is shown in Figure 1. Steel products made from the high strength, crack resistant steels are used nowadays for building contemporary constructions with the high technological load and rationally matched mass, for transport means consuming low energy amounts for their propulsion, as well as for various

machines and devices. Therefore the portion of the microalloyed constructional steels grows steadily and is 50% or more in the highly industrialized countries. Taking into account that portions of the unalloyed and alloyed steels together are more than 85% of the overall global steel production; introduction of the microalloyed steels contributes significantly to reduction of their use and lowering their manufacturing costs, and also to limiting the ecological damage caused by the environmentally harmful steel industry.

2. Metallurgical process and casting

The microalloyed steels trace their origin to the weldable unalloyed constructional steels containing up to 0.2% C and 1.8% Mn. The Nb, Ti, and V microadditions introduced into these steels in the amount to about 0.1% create in the properly selected hot working conditions the dispersive particles of the MX (M-Nb, Ti, and V; X-N and C) interstitial phases, limiting the recrystallised austenite grains coarsening during steel processing. The microadditions mentioned hereabove, and also Zr and B introduced sometimes in concentration of up to 0.005% have the significant affinity to O and N, and Ti and Zr also to sulphur. This means that before introducing the microadditions the molten steel needs proper deoxidation and desulphurisation. Therefore, the steel is subjected to the finishing treatment (secondary metallurgy) in the ladle (Fig.2), after the adequate dephosphorization in the oxygen converter process or smelted in the arc electric furnace, with the base slug directed onto the molten metal mirror and powder substances injected into the bath for deoxidizing, desulphurization, and modification of the nonmetallic inclusions, and also with alloying elements and microadditions. These substances should be roasted to remove humidity and water of crystallization before introducing them to the bath. Otherwise the excessive enrichment of the metal bath in hydrogen may occur, leading to the hydrogen brittleness of steel.

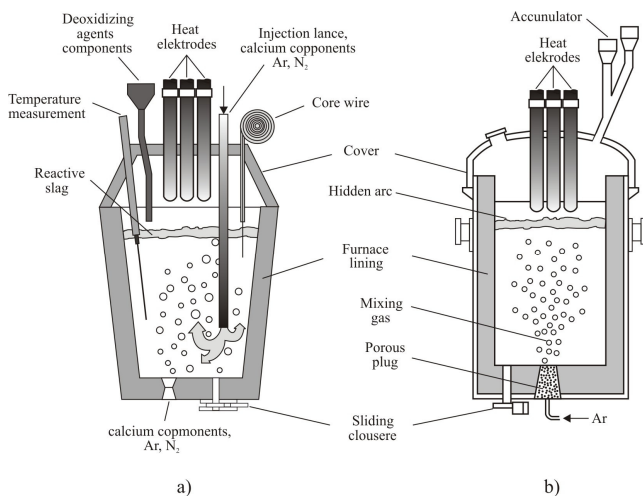


Fig. 2. Schematic drawing of a ladle furnace: a – with an injection lance supplied from the top, b – supplied with neutral gas through a porous stopper in the bottom of the ladle

The following occur in the ladle process: deoxidation of the bath with Fe-Si, Fe-Mn, and Al, desulphurisation with slag with the high concentration of CaO with portions of up to 30% Al₂O₃ and 5% CaF₂, and with the Ca, CaO, CaSi, or more rarely CaCO₃ and CaC₂ powders injected into the bath, introduction of the alloying elements and microadditions in portions that let them possible to dissolve in the molten metal, refining and homogenising of the chemical composition due to blowing them through with the Ar stream or more rarely with N₂, vacuum degassing of steel with the R-H circulation process or with the vacuum batch degassing (Fig.3), as well as preparing the molten metal for the assumed casting conditions.

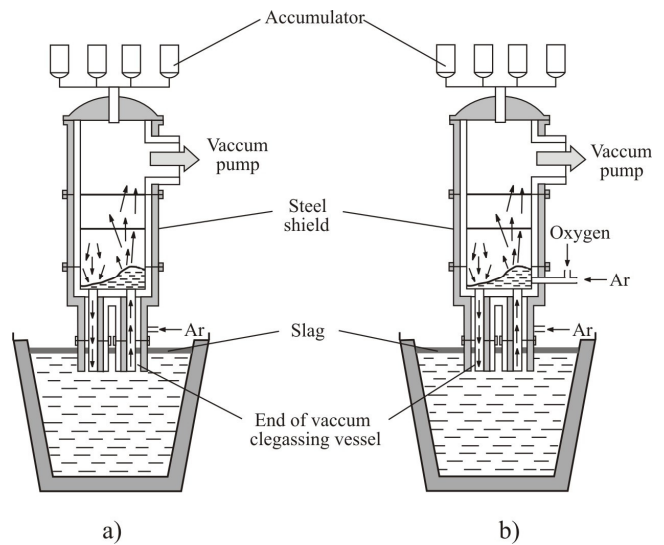


Fig. 3. Schematic drawing of the vacuum degassing process of steel by the RH process (a) and RH-OB process (b)

Molten metal achieving the right temperature after the ladle treatment and vacuum degassing is cast with the continuous casting method in the argon temperature, protecting the steel from the secondary oxidation and nitriding. The ladle with the molten metal is transported to the continuous casting system consisting of the tundish maintaining the continuous metalostatic pressure, copper mould with double walls between which cooling water flows, pinch rollers, some of which are equipped with electromagnetic stirrers - limiting development of columnar grains and development of the axial segregation in the slab, bending rollers and those straightening the slab, as well as of the shear (Fig.4). The continuous slab solidifies at a certain distance from the mould, while the molten metal deficiency in the slab core, connected with metal contraction during crystallisation, is replenished from the tundish under the metalostatic pressure. In newer solutions, some of the guiding system rollers fulfil also the role of the pinch rollers for the slab with liquid core, which makes a significant section change possible and limits development of the contraction cavity and shrinkage porosity in the slab. Slab section is, in fact, selected according to the final product shape, making allowance for the required rolling reduction ratio. The slab is subjected to hot-working after shear while hot and after additional heating up to the required temperature.

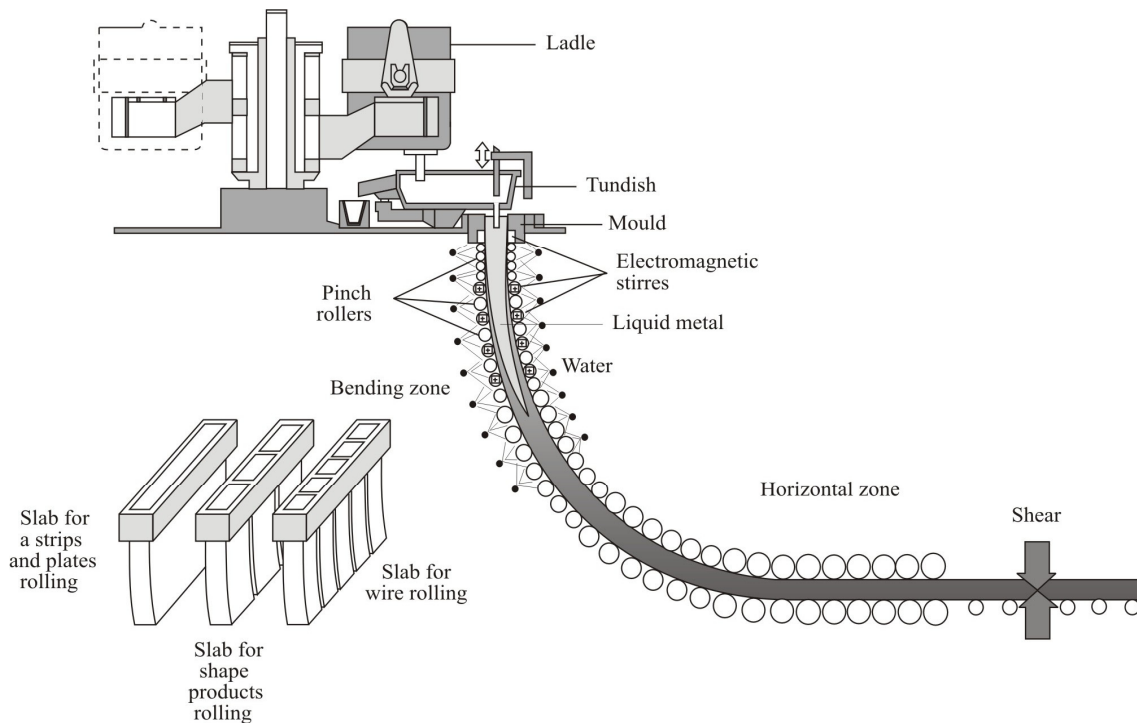


Fig. 4. Technological equipment for continuous casting of steel

3. Hot-working

3.1. Stability of interstitial phases

Determining the effect of microadditions on development of the fine-grained steel structure calls for understanding of development the MX interstitial phases containing them and of their stability in hot-working conditions. Microaddition M dissolved in austenite gets bonded at a certain temperature T with metalloid X into the interstitial phase MX with the cubic lattice of the NaCl type according to the reaction:



where:

[M] and [X]- microaddition M (Nb, Ti, V, Zr or B) and the metalloid X (N or C) portions respectively, dissolved in the solid solution γ at the temperature T, K.

Solubility of the MX phases in the solid solution allowing for the activation energy of their nucleation (dissolving) is described by the relationship:

$$\log [M][X] = -Q/2,303RT + A'/2,303, \quad (6)$$

where:

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

Equation (6) is presented usually as:

$$\log [M][X] = A - B/T, \quad (7)$$

making it possible to evaluate the temperatures of the beginning and end of the MX phase precipitation at a certain concentration of the microaddition M introduced into the steel. Values of constants A and B in equation (7) for the selected MX phases are given in Table 1. Knowledge of the temperatures of the beginning and end of precipitation of the MX phases in austenite is very important for designing the hot-working conditions for the microalloyed steels.

Table 1.
Values of A and B constants for selected carbides and nitrides

Constant	MX							
	AlN	VC	VN	TiC	TiN	NbC	NbN	BN
A	7184	7840	9500	10745	8000	7290	8500	13970
B	1.79	3.02	6.72	5.33	0.32	3.04	2.8	5.24

3.2. Effect of the MX phases

The effect of the dispersive particles of the MX phases on development of the fine-grained structure may be explained basing on steel structure changes during hot-working, e.g., multi-pass rolling at the temperature higher than the austenite recrystallisation temperature T_R . The feedstock heated to the temperature of the

beginning of hot-work has a coarse grained structure. During rolling with the draft degree big enough (about 15%) the austenite grains undergo the plastic deformation and elongate in the direction of metal flow (Fig.5) with their simultaneous dynamical recovery, causing the flow stress reduction. The static recovery and static recrystallisation occur in the plastically deformed metal after leaving the rolls, leading to the nearly complete work softening and to development of the fine-grained γ phase structure, which is metastable due to the big grain boundaries area and its disadvantageous ratio to the grain volumes. Therefore, the unfavourable austenite grains growth occurs when the recrystallisation is over. This process may be stopped in case of the microalloyed steels by the dispersive particles of the MX phases, developed in the austenite during plastic deformation, restricting the grain boundaries mobility. Therefore, the austenite grains size after the multi-pass rolling of the feedstock from the microalloyed steels is significantly smaller than in case of the steels without the microadditions (Fig.6). The dispersive particles of the MX phases cause precipitation hardening of the steel apart from restricting the grains boundaries mobility.

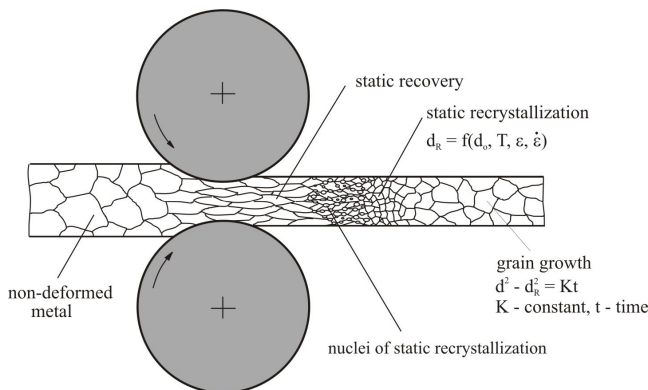


Fig. 5. Schematic diagram of structure changes of the deformed metal in the course of rolling and after its completion with a strain higher than ϵ_{cd}

The MX phases of various metallic additions have different thermal stability and differentiated effect on steel properties. The optimum Nb concentration in the low-carbon steels is 0.04% and makes it possible to increase the yield point of products fabricated at the correctly selected plastic working conditions by about 120 MPa due to refinement of grains and by about 160 MPa due to precipitation hardening by the dispersive NbC particles, with the simultaneous lowering of the T_{pk} temperature by about 40°C (Fig.7). Ti at the optimum concentration of 0.08% has a smaller effect on strengthening of products, as the MX phases of this metal, i.e., TiN may develop partially as early as during solidification of steel, and later in the solid condition, next Ti(C,N) precipitates, and finally TiC after depleting nitrogen at the temperature much higher than NbC. As the temperature gets lower, these phases coagulate and therefore their effect on refinement of grains and precipitation hardening decreases. On the other hand, the VN phases, especially VC, begin to precipitate close to temperature A_{r3} of the steel, whereby their effect on the refinement of grains is small, however significant for the precipitation hardening of products.

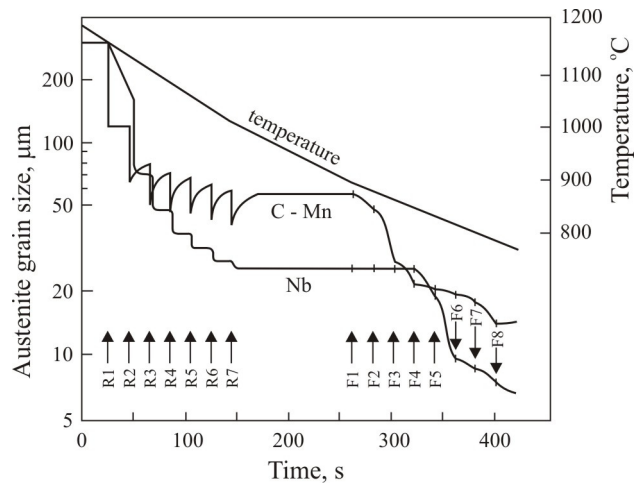


Fig. 6. Predicted changes of austenite grain size in the course of rolling 200 mm thick slabs into 20 mm thick plates of C-Mn steel and such a steel containing 0.04% Nb [27]; R₁ do R₇ and F₁ to F₈ - subsequent passes during roughing and finishing rolling, respectively

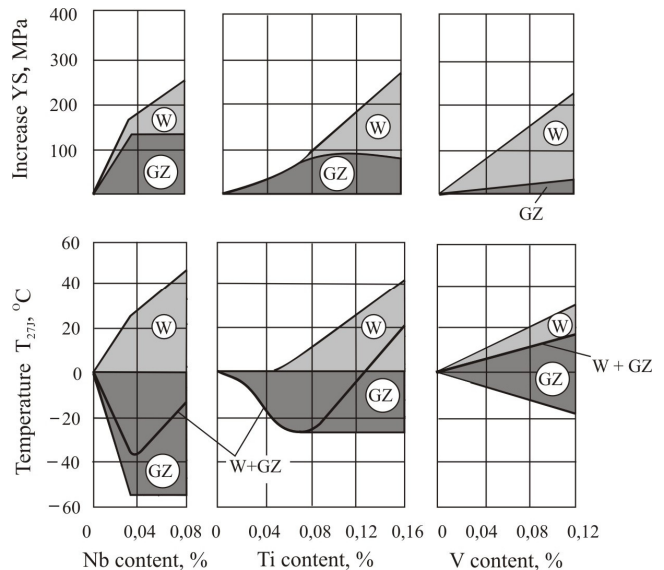


Fig. 7. Influence of the Nb, Ti, and V content on the increase of the yield point and variation of the impact transition temperature of the low-carbon steel: GZ - influence of grain refining, W - influence of precipitation hardening

The precipitation temperature range of the MX phases depends on concentrations of C and N in the steel and type of the introduced microaddition. For instance, at the V concentration of 0.1% in steel containing 0.2% and 0.02% N the temperature of beginning of the VN nitride precipitation is close to 1100°C and below 900°C for its end (Fig.8). Knowledge of the precipitation temperature range of the various MX phases is very important for designing the hot-working conditions for steels. Several

microadditions are introduced into the steel to increase the precipitation temperature range of the MX phases (Fig.9).

Many relationships have been worked out describing the recrystallised austenite grain size and the time $t_{0.5}$ necessary to develop 50% of the recrystallised γ phase fraction, taking into account the effect of the initial grain size of this phase and conditions of the plastic deformation, e.g., temperature and deformation rate and extent. These formulae, compiled in [25, 26], are useful for selecting the hot-working conditions, e.g., for the multi-pass rolling of the microalloyed steels (Fig.6).

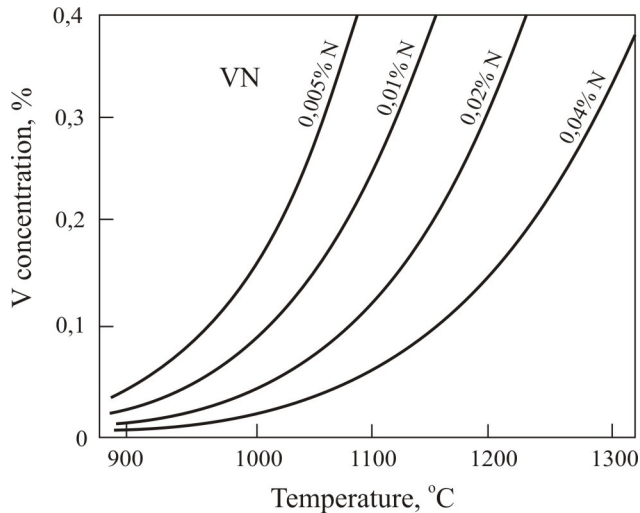


Fig. 8. Solubility curves of VN in austenite as a function of temperature and content of carbon and nitrogen in steel [23]

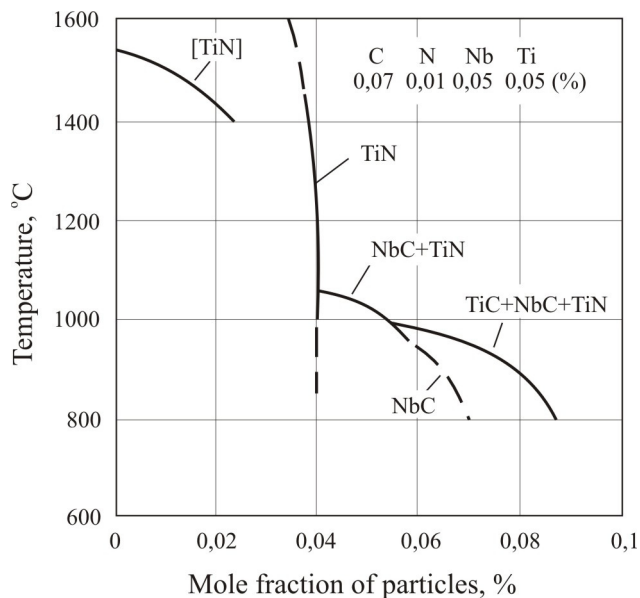


Fig. 9. The temperature sequence of nitrides and carbides precipitation in steels with microadditions of Ti and Nb [24]

3.3. Steel processing

Manufacturing of the steel products with the high mechanical properties from the microalloyed steels calls for adjusting the hot-working conditions to kinetics of precipitation in the austenite of the interstitial MX phases of the microadditions introduced to them, according to formula (7). Taking into account that the prevailing portion of steel products is manufactured by rolling then in case of the microalloyed steels two types of rolling processes are used - controlled rolling, employed usually for steels of the ferritic and ferritic-pearlitic structures, and the thermo-mechanical rolling for steels of martensitic and bainitic structures.

Controlled rolling consists in roughing the feedstock in the upper part of the plastic working temperature range, like in case of the unalloyed steels, and in finish rolling in the lower part of the temperature range for this process, corresponding to precipitation of the MX phases, limiting the recrystallised austenite grains growth in pauses between the consecutive passes. Finish rolling ends usually at the temperature slightly lower than the austenite recrystallisation temperature T_R and a little bit higher than A_{33} or A_{11} for the steel with the subsequent air cooling of products (Fig.10) Presence of the slip bands and deformation bands in the non-recrystallised austenite causes the increase of the population of locations convenient for nucleation in $\gamma \rightarrow \alpha$ transformation and leads to development of the more fine-grained structure of the steel products.

Thermo-mechanical rolling (with the controlled recrystallisation) consists in rolling the feedstock in the MX phases precipitation temperature range with pauses between the consecutive rolling passes making complete austenite recrystallisation possible as well as the isothermal holding of products at the properly selected temperature of the end of plastic working by time $t_{0.5}$ - needed for development of 50% fraction of the recrystallised austenite and their direct quenching with the appropriate cooling rate. Isothermal holding of products at the temperature of the end of rolling by time $t_{0.5}$ is necessary, as the high dislocation density in the plastic deformed austenite causes increase of the critical cooling rate due to the significant increase of the diffusion transformations of this phase and lowering hardenability of the steel because the alloying elements in the supercooled austenite are depleted and due to the effect of the dispersive carbides. Products manufactured with this method are subjected only to the high-temperature tempering or ageing.

Integrated production lines Connecting directly ladle treatment of the molten metal and continuous casting of slabs with the controlled rolling or thermo-mechanical rolling and their controlled cooling from the correctly selected end of hot-working temperature has the particularly significant effect on lowering the manufacturing costs of steel products with stable mechanical properties. Such integrated technological lines are used nowadays for production of a big amount of heavy plates and sheets, as well as of selected long sections. The first integrated manufacturing line for steel sheets named CDP (Compact Strip Production) was commissioned in 1989 in Nucor Steel Co in USA. This line consists of the continuous casting machine for concast slabs 40 to 50 mm thick, shear, roller table furnace, hydraulic mill scale breaker, roll train, laminar flow cooling, and a coiler (Fig.11). Next designs of this line contain two continuous casting machines with roughing train for slabs with liquid core and two roller table furnaces delivering feedstock alternately to the rolling train.

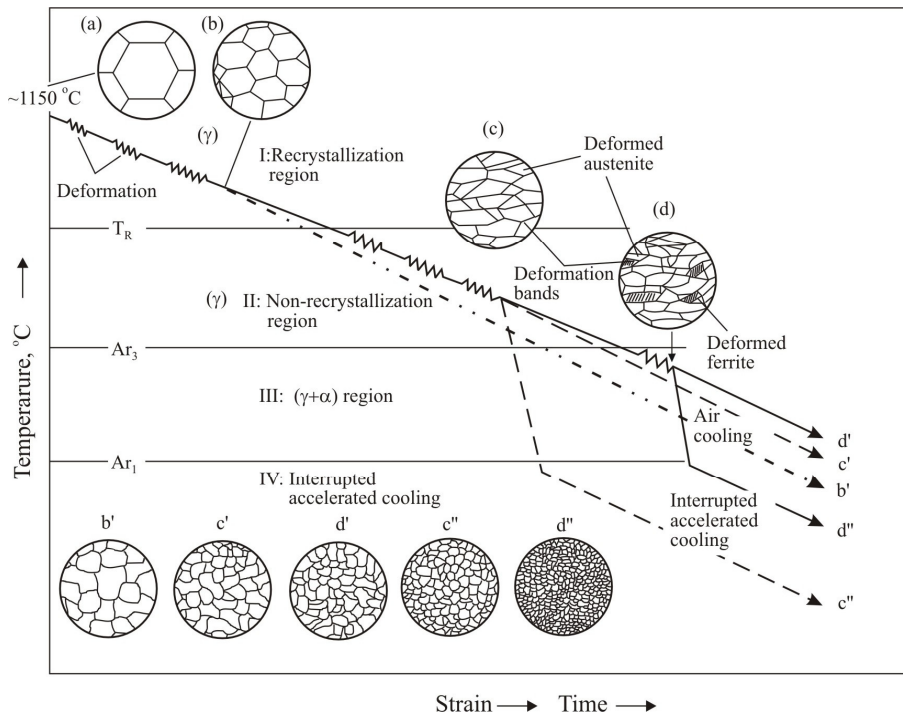


Fig. 10. Schematic diagram of the influence of accelerated cooling on the microstructure of low-carbon microalloyed steel products during controlled rolling [28]

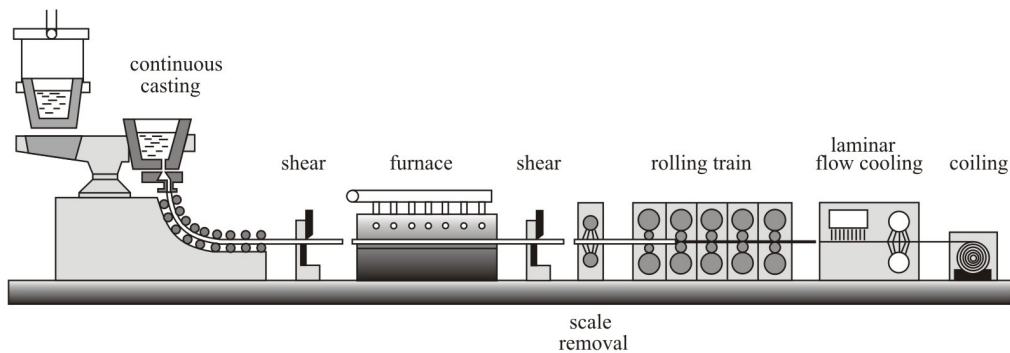


Fig. 11. Schematic diagram of an integrated production line of the hot-rolled plates in the CSP (Compact Strip Production) process

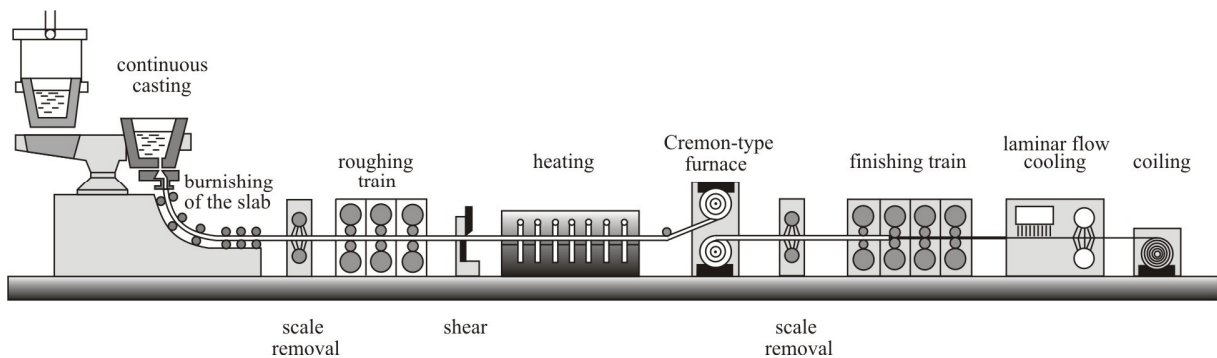


Fig. 12. Schematic diagram of an integrated production line of rolled plates in the ISP (In-line Strip Production) process

At the end of last century the ISP (Inline Strip Production) line was developed in ATA Averdi steel plant in Italy with production capacity of $5 \cdot 10^5$ tons/year. This line (Fig.12) consists of the continuous casting machine for concast slabs 60 mm thick with the burnishing mill stand for squashing the slab with the liquid core to 43 mm, roughing train, furnace for inductive heating of the strand, Cremon-type furnace with the strand coiler and decoiler, ensuring maintaining the correct feedstock temperature, mill scale breaker, finishing train, laminar flow cooling, and a coiler for the finished product. The steel sheets rolled by this line may be up to 1 mm thick. Other designs of the integrated manufacturing lines for sheets and selected long sections are presented in [26].

4. Mechanical properties of the selected steel products

The general use steel products manufactured with the controlled rolling from the weldable microalloyed steels containing up to 0.2% C, to 0.6% Si, to 1.7% Mn and microadditions Nb, Ti, V i N; they achieving - depending on thickness - YS_{min} from 255 to 460 MPa, UTS from 360 to 720 MPa, TEI from 24 to 17%, carbon equivalent CEV from 0.34 to 0.46, and energy of breaking the Charpy V test pieces at room and low temperatures, specified in Table 2. Weldability of these steels is characterised by the carbon equivalent value determined from the relationship:

$$CEV = C + \frac{Mn}{6} + \frac{Cr+Mo+V}{5} + \frac{Cu+Ni}{15}, \quad (8)$$

where:

C, Mn, Cr, ... - contents of these elements in mass fraction %

Steels with the value of $CEV \leq 0.45$ with thickness of the joined elements up to 25 mm and $CEV \leq 0.41$ with thickness of the joined elements from 25 to 37 mm have good weldability, without the susceptibility of the welded joints to cold cracks [29]. However, the limited concentration of S and modification of the nonmetallic inclusions eliminates susceptibility of the microalloyed steels to hot cracks.

The low-carbon COR-TEN type rust resilient steels have an important technical significance, containing apart of the Nb microaddition up to 1.4% Mn, $\leq 0.5\%$ Si, $\leq 0.7\%$ Cr, $\leq 0.5\%$ Cu, and sometimes also the increased concentration of P to 0.040%, achieving YS from 350 to 480 MPa, UTS from 480 to 580 MPa, TEI from 25 to 18%. Structures made from these steels are resistant to the atmospheric corrosion and do not require the periodical painting [30].

Oil and gas pipelines are made from steels containing up to 0.03% C, to 1.6% Mn, to 0.3% Si, ≤ 0.02 , and even 0.005% P, $\leq 0.003\%$ S, $\leq 0.05\%$ Nb, $\leq 0.15\%$ Ti, ≤ 60 ppm N and 0.001% B, and sometimes also low concentrations of Cu, Ni, and Cr, desulphurised carefully, degassed, and free from the nonmetallic inclusions in the ladle treatment process.

Sheets from these steels after the controlled rolling and accelerated cooling to the temperature lower than A_{r1} acquire the

ferritic bainite structure and $YS_{0.2}$ from 490 to 590 MPa, UTS from 480 to 670 MPa, TEI above 34%, $KV_{-40^\circ C} > 200J$, fracture appearance transition temperature FATT - determined for 50% portion of the brittle fracture area below $-100^\circ C$, $CEV \leq 0,40$ [31,32]. Weldability of these steels, apart from the carbon equivalent (7) is characterised by the heat affected zone cracking parameter Pcm, more useful in the extreme climatic conditions, determined from the relationship [33]:

$$P_{cm} = C + \frac{Si}{30} + \frac{Mn+Cr+Cu}{20} + \frac{Ni}{60} + \frac{Mo}{15} + \frac{V}{10}, \quad (9)$$

where:

C, Si, Mn, ... - mass fractions of these elements in the steel.

Pipelines made from welded tubes used for oil or gas transport under high pressure are often subjected to hydrogen damage because of the humid hydrogen sulphide contaminating these media. This process called the SSCC stress corrosion (sulfide stress corrosion cracking) is caused by the chemisorption dissociation of the humid H_2S on surface of tubes. The atomic hydrogen created in this reaction is absorbed by the surface layer of the pipes and penetrates inside the steel. The migrating H atoms are trapped in gaps in the neighbourhood of the nonmetallic inclusions and other defects, where they are recombined to the H_2 particle form. The inflow of the H atoms into these gaps and the exothermic recombination reaction of this element causes development of the H_2 gas bubbles with the growing pressure, which summing up with the constant or pulsating pressure of the transported medium leads to nucleation and development of cracks, succoured by segregation of the atomic hydrogen in stress concentration in front of their vertices. Therefore, steels used for these devices should have the fine-grained structure with the coherent or semi-coherent MX phases' reinforcing particles and the limited portion of the nonmetallic inclusions and impurities.

Also the welding process of these steels calls for using the low-hydrogen methods and auxiliary materials without humidity, crystalline water, and organic substances. Otherwise, the atomic hydrogen coming from dissociation of these substances penetrates from the joint to the heat affected zone austenite and results in the increase of the structural stresses connected with the martensitic transformation. This may be the cause for the hydrogen induced corrosion HIC of the welded joints.

Heat treated heavy plates with $YS_{0.2}$ from 550 to 960 MPa, UTS from 670 to 1130 MPa, TEI $> 15\%$, and $KV_{-40^\circ C} > 27J$ were initially made from the low-carbon alloy steel with CEV from 0.72 to 82. This causes the necessity for the local initial heating of the welded elements to the temperature exceeding even $200^\circ C$, which had an important effect on manufacturing costs of the welded constructions [25, 26].

Introducing the Nb, Ti, and B elements and limiting the P and S contents made it possible to limit the concentration of the alloying elements and carbon in steel. Lowering the carbon equivalent CEV connected with it calls for the local heating up of elements before welding to the correspondingly lower temperature and results in cutting the manufacturing costs of the welded constructions.

Initially, heavy plates from these steels were heat treated after their controlled rolling. Nowadays, rolling with the energy-saving thermo-mechanical method is used in the temperature range of precipitation of the MX phases of the microadditions introduced

Table 2.
Chemical composition and mechanical properties of metallurgical products from weldable fine-grained microalloyed steels

Steel grade	Maximum concentration ¹⁾ , %										Mechanical properties										
	C	Si	Mn	P	S ²⁾	Nb	V	Ti	Al _c ³⁾	N	YS _{min} MPa/CEV for products with a thickness, mm			UTS, MPa	TEI ⁴⁾ , %	KV, J at a temperature, °C					
											≤16	16÷40	40÷63			20 -10 -20 -30 -40 -50					
S275M S275ML	0.13	0.50	1.50	0.035	0.030	0.05	0.08	0.05	0.02	0.015	275	265	255	360÷510	24	55	43	40	-	-	-
0.030				0.025	0.34						0.34	0.35	63			51	47	40	31	27	
S355M S355ML	0.14	0.50	1.50	0.035	0.030	0.05	0.10	0.05	0.02	0.015	355	345	335	450÷610	22	55	43	40	-	-	-
0.030				0.025	0.39						0.39	0.40	63			51	47	40	31	27	
S420M S420ML	0.16	0.60	1.70	0.035	0.030	0.05	0.12	0.05	0.02	0.020	420	400	390	500÷660	19	55	43	40	-	-	-
0.030				0.025	0.43						0.45	-	63			51	47	40	31	27	
S460M ⁵⁾ S460ML ⁵⁾	0.16	0.60	1.70	0.035	0.030	0.05	0.12	0.05	0.02	0.025	460	440	430	530÷720	17	55	43	40	-	-	-
0.030				0.025	0.45						0.46	-	63			51	47	40	31	27	

¹⁾ moreover, steels may contain up to 0.20% Mo and up to 0.30% Ni, and in case of long products the S420 steel grade may contain up to 0.60%, and S460 grade - up to 0.70% Ni, total of Cr+Cu+Mo ≤ 0.60%; ²⁾ in case of delivery of products with thickness ≤ 16 mm for railways concentration of S_{max} up to 0.007% may be agreed; ³⁾ - with the sufficient concentration of nitrogen binding elements the Al_c contents is of no importance; ⁴⁾ - for length L₀ = 5.65 S₀^{0.5}, where S₀ – initial test piece transverse section; ⁵⁾ - CEV specified for these steels pertain to long products

into the steel. In this case pauses between the consecutive rolling passes should ensure the full or more than 50% recrystallisation of the plastically deformed austenite. Also the final products should be isothermally held before quenching at the appropriately selected temperature of the end of rolling by the time t_{0.5}, necessary for development of the 50% fraction of the recrystallised austenite. Due to the long time of the full recrystallisation and the t_{0.5} time of the alloy steels' austenite, the retention shield has to be used in breaks between the final rolling passes and after completing the plastic forming. However, quenching of steel sheets directly from the temperature of the end of rolling makes it possible to limit the costly heat treatment to tempering only.

Failing to fulfil the conditions related to austenite recrystallisation results in cumulation of the deformation in the successive rolling passes. This leads to increase of the inhomogeneity of strain and to development of the segregation shear bands in austenite in the rolling direction (Fig.13). The inhomogeneity of structure in the transverse section of steel sheets and concentration of the dispersive MX phases and the M₃₃C₆ carbides in the segregation bands (Fig.14) results in the significant deterioration of the plastic properties and crack resistance and also in the disadvantageous growth of the plastic anisotropy of the steel sheets [26, 34].

The micro-alloyed steels with the addition of 0.5 or 2% Cu feature also an important group, used for manufacturing the heavy plates with the increased corrosion resistance, designed for making the submarine hulls, pressure tanks for gas storing and transport, nuclear reactor jackets, and other devices [26, 31, 35-37]. Steel plates with concentration of 0.5% Cu after controlled rolling are heat treated, and from the steel with concentration 2% Cu are quenched and aged. Such plates may also be made by rolling using the thermo-mechanical treatment method. Due to the high value of the CEV equivalent from 0.62 to 1.05, these steels require using the low-hydrogen welding methods and properly selected conditions of this process.

Steel sheets for the automotive industry have to be suited, as regards their mechanical and technological properties, to the requirements of the automotive industry striving to produce the transport means with the mass fit properly to their complex service parameters. Reduction of the vehicle's weight is connected directly with the decreasing fuel consumption and limited emission of the harmful exhaust gases to the environment. Therefore, there is a need to manufacture the weldable heavy plates and sheets with high strength and technological deformability as well as strain hardening. An essential importance in this case have the micro-alloy steels making manufacturing possible of products with the fine-grained structure. Apart from steel sheets with the ferritic structure, the increasing technical importance is gained by sheets with the multi-phase structure, i.e., ferritic-martensitic of the DP (dual phase) type, ferritic-bainitic with the retained austenite of the TRIP (transformation induced plasticity) type - subjected to the martensitic transformation during technological forming or in case of collision, with the complex combination of various phases of the CP (complex phase) type, and martensitic of the TMS (transformation martensite steel) type, characteristic of the increasing strength while retaining the necessary plastic properties (Fig.15) and technological deformability by, e.g., press forming. Some of these sheets are used for car bodies and others for the particularly loaded car elements. Corrosion protection is also of the significant importance.

HSZ type sheets with the YP_{0.2} from 390 to 570 MPa, UTS from 450 to 660 MPa, total elongation TEI from 28 to 24% and the uniform one TEI_g from 20 to 14% are usually made from steels containing 0.05 or 0.07% C, 0.4 to 1.0% Mn, 0 or 0.4% Si, up to 0.12% V, 0.03% Nb and up to 0.020% N. They are used for chassis of buses and lorries, wheel disks, bumper reinforcements, suspension elements, and others [39-41].

Cold-rolled sheets are made from the low-carbon low-alloy steels. Deformability of these sheets for deep drawing is characteristic of the high value of the normal anisotropy

coefficient $\bar{\Gamma}$ and of the index exponent n of the real deformation in the relationship describing the yield stress of the steel in the uniform elongation range of the tensile test pieces [25, 26]. Rolled sheets with the high rolling reduction have the high value of the $\bar{\Gamma}$ coefficient, obtaining the strong $\{111\}$ weak $\{100\}$ textures after the recrystallisation annealing. The strong texture and $\{111\}$ are obtained also by rolling the sheets in the temperature range from slightly lower than A_{r3} of the steel to slightly higher from the ferrite recrystallisation temperature and their slow cooling [42]. The BH (bake hardening) type sheets are included in this group, reinforced by strain ageing during baking the lacquer or depositing the organic layers at the temperatures from 210 to 240°C, as well as the IF (interstitial free) sheets without the interstitial elements characteristic of the particularly high deformability during drawing [26, 43, 44]. Sheets from the siliconfree steels are corrosion protected by hot-dip zinc coating in the continuous line of the galvannealing type [26, 45].

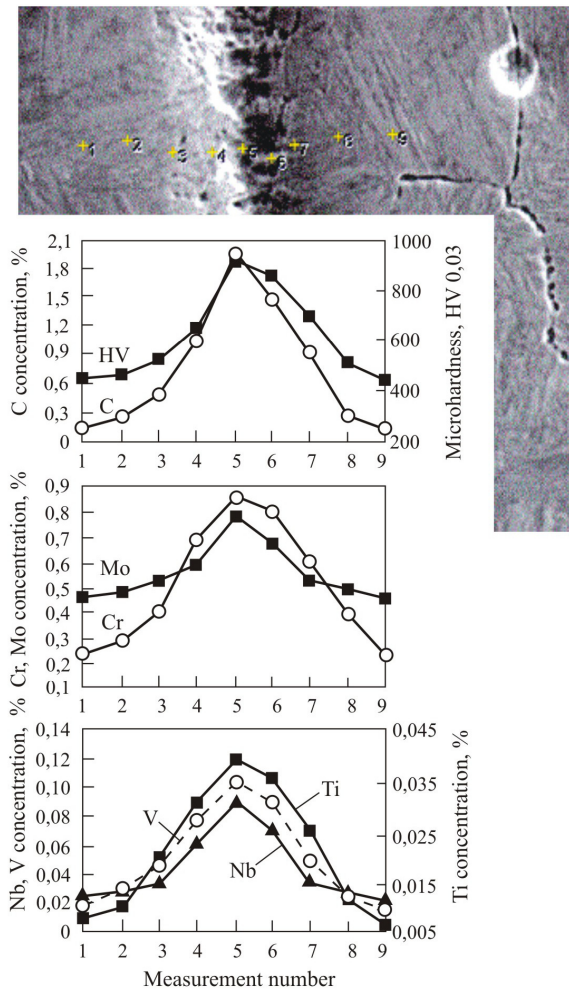


Fig. 13. Distribution of the C, Cr, Mo, Nb, Ti, and V content and microhardness on the cross-section of the segregation band and in the adjacent martensite zones of a steel matrix

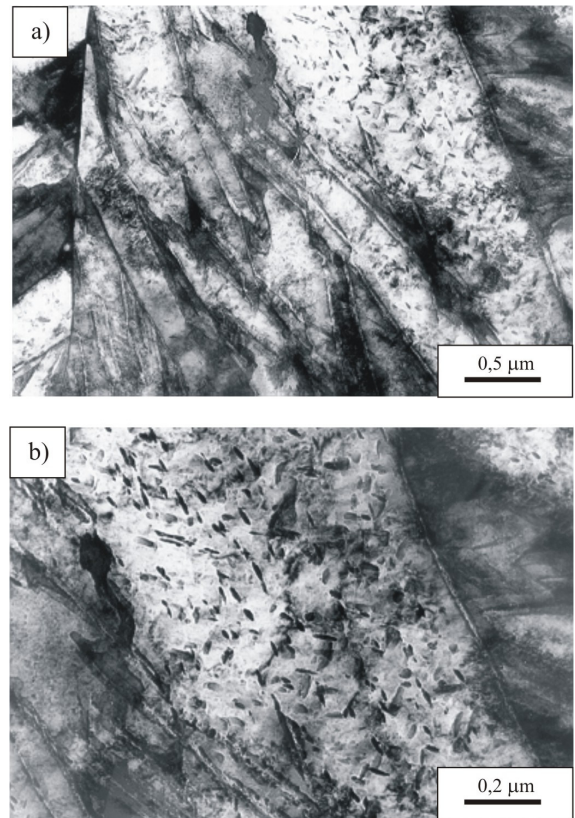


Fig. 14. Structure of the segregation band: a - martensite located inside the dynamically recrystallised austenite grains elongated in the rolling direction with an envelope of the $M_{23}(C,B)_6$ carbides on grain boundaries, b - dispersive precipitations of the MX type interstitial phases in the plate of martensite [34]

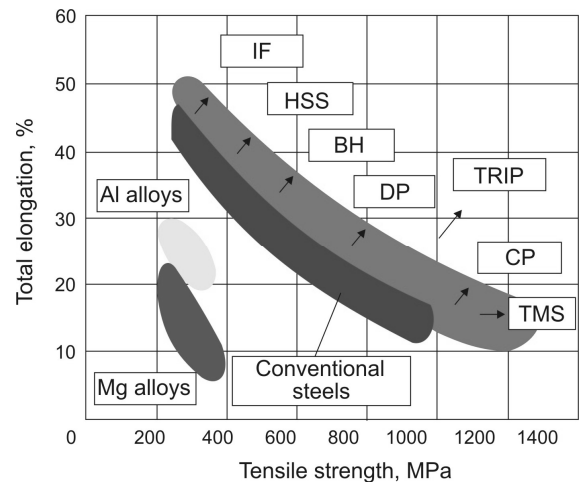


Fig. 15. Present and future steel and alloy grades applied for car body steel sheets production [38]: IF - Interstitial-Free steels, HSZ - High-Strength steels, BH - Bake-Hardening steels, DP - Dual Phase ferritic-martensite steels, TRIP - Transformation Induced Plasticity steels, CP - Complex Phase steels, TMS - Martensitic steels

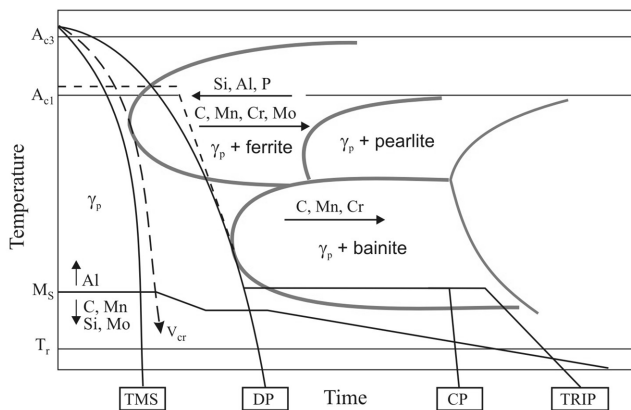


Fig. 16. Schematic diagram of the transformation of the supercooled austenite of steels used to produce sheets with a DP-, TRIP-, CP-, and TMS-structure rolled by means of thermo-mechanical treatment or underhardened or isothermally quenched from a temperature slightly higher than A_{c1} (dashed line)

The multi-phase structure of the DP and TRIP types in sheets from the low-carbon steel hot- or cold-rolled may be obtained either by the underhardening from the temperature slightly higher than A_{c1} of the steel or by isothermal underhardening at the temperature slightly higher than M_s of the austenite enriched with carbon, in relation to the concentration of this element in steel, usually at the temperature from 350 to 450°C. In the last case, steels with concentrations of 0.2 to 0.5% C, up to 1.5% Mn and up to 1.7% Si + Al, are useful for that, being characteristic of the $\gamma \rightarrow \alpha$ transformation shifted to short times, and with the pearlitic and bainitic transformation in the opposite direction (Fig.16). The Si and Al elements segregate in front of the bainitic transformation; therefore, the originating bainite contains the slight portion of Fe_3C while the non-transformed austenite is enriched with carbon and acquires stability at room temperature in the amount depending on the chemical composition of steel and temperature of the isothermal transformation.

Technical interest is attracted by production of these sheets using the thermo-mechanical treatment method. Sheets 2 mm thick of the TRIP and DP types made with this method from steel containing 0.2% C, 1.4% Mn and 0.5% Si and micro-additions Nb and Ti demonstrate respectively $Y_{P_{0.2}}$ above 510 and 500 MPa, UTS above 780 and 880 MPa, elongation TEI_g above 19 and 17%, and the product $UTS \cdot TEI_g$ characterising their deformability for drawing 15000 and 14000 MPa % [46]. However, the TRIP and DP type sheets made from the same steel using the traditional heat treatment have significantly lower mechanical and plastic properties. Increasing carbon concentration to 0.5% and Si to 1.5% as well as the isothermal quenching from the temperature slightly higher than A_{c3} of the steel give the possibility for fabricating the TRIP type elements with the bainitic structure with the about 25% portion of the retained austenite, acquiring $Y_{P_{0.2}} > 780$ MPa, UTS > 1100 MPa, $TEI = TEI_g = 34\%$, and $Y_{P_{0.2}}/UTS = 0.69$, characteristic with absorbing the big amount of energy during plastic deformation at high rates.

Similar heat- and thermo-mechanical treatment conditions are used for the CP type sheets. In this case the isothermal bainitic transformation time should be adjusted to enrich the retained austenite with carbon to the concentration ensuring M_s a little above room temperature and fast cooling of the product (Fig.16).

Drop forged elements for the automotive industry, mining-, road-making machinery and others are also made from the micro-alloy steels with the controlled forging or thermo-mechanical treatment. In this case the charge heating conditions should not cause the total dissolving of the MX phases in the solid solution, leading to the excessive austenite grain growth, as the high deformation rate and short breaks between the successive forging stages do not provide the convenient conditions for the recrystallisation course and refinement of grains of this phase. Forgings made from the steel with the ferritic-pearlitic structure acquire $Y_{P_{0.2}}$ from 450 to 650 MPa, UTS from 750 to 900 MPa and $KV > 27J$. The more advantageous mechanical properties, including $Y_{P_{0.2}} > 700$ MPa, UTS > 800 MPa and $KV > 150J$, acquire forgings with the micro-addition of boron fabricated with the thermo-mechanical method. The relatively low hardness after the high-temperature tempering does not pose any difficulties for machining of the forged elements [45, 46].

5. Summary

The related information show that the micro-alloy steels are the new generation of the weldable constructional steels - non-alloyed and alloyed ones, making manufacturing possible of steel products and machine elements with high strength and crack resistance using the energy-saving technologies. As important technical and economical achievement features the mass-production of goods with the stable service properties in the integrated production lines with the controlled form of the metallurgical process flow and further processing in conditions fit to the types and concentrations of the micro-additions and alloying elements introduced into the steel. This gives the possibility to limit the number of production facilities with the considerable reorganisation of metallurgical plants. Investment expenditures made are returned within a short time due to manufacturing of goods necessary for building the modern constructions, as well as machines and devices with the rationally suited mass and extended service life. This indicates also to the possibility of reducing production of the as-cast steel, as well as of energy and raw materials needed for manufacturing this material.

References

- [1] C.F. Tipper: Report on Conference on Brittle Fracture in Steel Plates, Mat. Conf. BISRA, Cambridge, England 1945.
- [2] J.A. Reinbolt, W.J. Harris: Trans. ASM, 43, 1951, p. 1175.
- [3] A.M. Sage, F. Copley: Iron Steel Inst., 195, 1960, p. 422.
- [4] W.D. Biggs: Brittle Fracture of Steel, McDonalds and Evans, London 1960.
- [5] B.L. Averbach (et al. eds.): Fracture, IMT Prees, Cambridge Massachusetts, 1963.
- [6] D.G. Drukner, J.J. Gilman (eds.): Fracture of Solid, J. Wiley - Interscience Publ., New York - London, 1962.

- [7] C. McMahon: *Micromechanisms of Cleavage Fracture in Polycrystalline Iron*, IMT Press, Cambridge Massachusetts, 1963.
- [8] C.J. Osborne: *Fracture*, Butterworth, London, 1965.
- [9] F.A. McClintock, A.S. Argon: *Mechanical Behaviour of Materials*, Addison - Wesley Publ., Massachusetts, 1966.
- [10] A. Maciejny: *Brittleness of metals*, Wyd. Śląsk, Katowice, 1973. (in Polish)
- [11] S. Butnicki: *Weldability and brittleness of steel*, WNT, Warszawa, 1975. (in Polish)
- [12] E.O. Hall: *Proc. Roy. Soc.*, 64B, 1951, p. 747.
- [13] N.J. Petch: *Journ. Iron Steel Inst.*, 173, 1953, s. 25; 174, 1953, p. 28.
- [14] A.N. Stroh: *Phil. Mag.*, 46, 1955, s. 968; *Adv. Phys.*, 6, 1957, p. 418.
- [15] A.H. Cottrell, N.J. Petch: *Fracture*, Averbach B.L. (et al. eds.), Swampscott Conference, New York – London, 1959.
- [16] T. Gilman: *The Physical Metallurgy of Microalloyed Steel*, Cambridge Univ. Press, Cambridge 1957.
- [17] M. Korchynsky (ed.): *Microalloying'75*, Union Carbide Corp., New York, 1977.
- [18] A.J. De Ardo, G.A. Ratz, J.P. Wray (eds.): *Thermomechanical Processing of Microalloyed Austenite*, Met. Soc. AIME, Warrendale PA, 1982.
- [19] *HSLA – Technology and Applications*, Conf. Proc. ASM, Metals Park, Ohio, 1984.
- [20] K.A. Taylor (et al. eds.): *Physical Metallurgy of Direct Quenched Steels*, Proc. Symp. TSM and ASM, Chicago, Ilionis, 1992.
- [21] *Microalloying'95*, Proc. Int. Conf., Iron and Steel Soc., Pittsburgh PA, 1995.
- [22] K. Hulka: *Microalloyed Constructional and Engineering Steel*, Niobium Product Comp., GMBH, 1990.
- [23] M. Korchynsky: *8-th PDT Conf. Proc.*, 1988, p. 92.
- [24] W. Müschenborn et al.: *Proc. Int. Conf. Microalloying'95*, Iron and Steel Soc., Pittsburgh PA, 1995, p. 35.
- [25] J. Adamczyk: *Engineering of steel products*, Wyd. Politechniki Śląskiej, Gliwice, 2000. (in Polish)
- [26] J. Adamczyk: *Engineering of metal products*, Wyd. Politechniki Śląskiej, Gliwice, 2004. (in Polish)
- [27] C.M. Sellars: *Proc. Hot – Working and Forming Processes*, C.M. Sellars, G.J. Davies (eds.), London 1979, p. 3.
- [28] Tanaka T.: *Proc. Int. Conf. Microalloying'95*, Iron and Steel Soc., Pittsburg PA, 1995, p. 165.
- [29] J. Mikula, L. Wojnar: *Employment of analytical methods in steel weldability assessment*, FOTOBIT, Kraków, 1996. (in Polish)
- [30] *Dura Plate 50, 60, 65 and 70 Atmospheric Corrosion Resistant High Strength Low Alloy Steel*, Republic Steel, 1982.
- [31] P.J.P. Bordignon: *Sider Latinomer*, 315, 1986, p. 75.
- [32] J.G. Williams et al.: *Proc. Int. Conf. Microalloying'95*, Iron and Steel Soc., Pittsburgh, 1995, p. 117.
- [33] Y. Ito, K. Bessyo: *A Prediction of Welding Procedure to Avoid Heat Affected Zone Cracking*, IIW Doc., IX-631-69.
- [34] J. Adamczyk, M. Opiela: *Inżynieria Materiałowa* 6, 2002, p. 717; *Journ. Materials Processing Technology*, 157-158, 2004, p. 456; *Hutnik – Wiadomości Hutnicze*, 3, 2005, p. 162.
- [35] T.W. Montemarano et al.: *Journ. Ship Production*, 2, 1986, p. 145.
- [36] S.J. Micolac, M.G. Vassilalos: *Journ. Ship Production*, 2, 1986, p. 331.
- [37] S.J. Manganello; *Proc. Int. Conf. Processing, Microstructure and Properties of Microalloyed and Other Modern High Strength Low Alloy Steels*, De Ardo A.J. (ed.), Iron and Steel Soc., Pittsburgh PA, 1991, p. 289.
- [38] *Stahl für den Automobilbau*, *Stahlmarkt*, 12, 1999, p. 38.
- [39] M. Korchynsky, S. Zajac: *Proc. Int. Conf. AMME*, Gliwice-Wisła, 1996, p. 1.
- [40] A. Catterjee, S. Chandra: *Steel World*, 2, 1997, p. 45.
- [41] W. Bald et al.: *Innovative Technologies for Strip Production*, SMS, 1999; *Proc. World Steel Conf.*, London, 1999.
- [42] M.R. Barnet, J.J. Jonas: *ISIJ Int.*, 7, 1977, p. 706.
- [43] D.A. Jones, R.S. Daniel: *IF Steel Proceedings*, Pittsburgh, 2000.
- [44] W. Bleck et al.: *Thyssen Techn. Berichte*, 1, 1991, p. 43.
- [45] W. Warnecke et al.: *Thyssen Techn. Berichte*, 1, 1991, p. 75.
- [46] J. Adamczyk, A. Grajcar: *Hutnik – Wiadomości Hutnicze*, 3, 2005, p. 170.