

# Influence of plastic deformation on CCTdiagrams of low-carbon and mediumcarbon TRIP-steels

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# Manufacturing and processing

# <u>ABSTRACT</u>

**Purpose:** The aim of the paper is to investigate the influence of plastic deformation and cooling conditions on a structure and a shape of CCT-diagrams of new-developed TRIP-aided microalloyed steels.

**Design/methodology/approach:** The diagrams of undeformed and plastically-deformed supercooled austenite transformations for low-carbon and medium-carbon microalloyed steels were determined. A part of the specimens were austenitized at a temperature of 1100°C, then slowly cooled to 900°C and next cooled to ambient temperature with a various rate from 1 to 300°C/s. To investigate the influence of plastic deformation on a shape of CCT (Continuous Cooling Transformations) diagrams, another part of the specimens were 50% deformed at 900°C and cooled to ambient temperature with a rate from 88 to 1°C/s. The DIL805A/D dilatometer, with a LVDT-type measuring head, was used to carry out dilatometric tests.

**Findings:** It was found that a shape of CCT diagrams of elaborated steels predisposes them for multiphase sheets manufacturing. The new-developed steels possess ferritic and bainitic bays put forward to short times and pearlitic regions put aside. However, cooling the steel with a constant rate from austenitizing temperature doesn't lead to obtaining proper participation of ferrite. Plastic deformation of steel has a profitable influence on the shape of supercooled austenite curves. The region of  $\gamma \rightarrow \alpha$  transformation is translated to the left at simultaneous raise of start temperature of austenite into ferrite transformation resulting in definitely higher ferrite fraction. Moreover, significant refinement of microstructure in a whole range of cooling rate was also obtained.

**Research limitations/implications:** To increase the ferrite fraction, modification of the cooling after hotworking finishing should be applied. In the fist stage, steel should be rapidly cooled in order to enter the range of  $\gamma \rightarrow \alpha$  transformation and successively slowly cooled in a range of  $\gamma \rightarrow \alpha$  transformation.

**Practical implications:** The obtained CCT diagrams of supercooled plastically-deformed austenite transformations can be useful in a determination of cooling in the thermo-mechanical processing for TRIP-type steel sheets.

**Originality/value:** The diagrams of the plastically-deformed supercooled austenite for TRIP-type microalloyed steels were obtained.

Keywords: Metallic alloys; TRIP-steel; CCT-diagram; Plastic deformation; Supercooled austenite

# **<u>1. Introduction</u>**

In the last twenty years a great attention has been focused on multiphase steels, composed of ferritic matrix and bainitic islands with retained austenite. Sheets made of these steels undergo transformation of retained austenite into martensite, during sheet-metal forming. This transformation essentially contributes to the increase of mechanical properties of products and delays thinning of pressed element, what favours application of high deformations during technological cold forming [1-5]. Thanks to many studies, a new group of steels called TRIP (TRansformation Induced Plasticity) was elaborated, containing 0.15 to 0.5%C, 1-3%Mn, 0.3-2%Si, 0-1.5%Al, and sometimes also up to 0.2%P [6-9].

Influence of chemical composition on thermal and mechanical stability of retained austenite, deciding in large scale about mechanical properties of this group of steels, was investigated in detail [6-8]. Optimal conditions of heat treatment after cold rolling, resulting in obtaining from 5 to 15% of retained austenite, were also examined [2, 7, 8]. Problems present during pressing [10], galvanizing [11] and weldability of this steel group [12] were also the subject matter of the research.

For the present, one of the challenges is to increase mechanical properties of TRIP-type steels without worsening their ductility and formability. Their measure is a product of UTS (Ultimate Tensile Stress) and UEI (Uniform Elongation). These properties are especially important in the aspect of further section decrease of individual structural components of produced cars. Increase of mechanical properties contributes also with improvement of passengers' safety in the result of application of steels absorbing higher amounts of energy, especially in the zones of controlled degree of deformation, where mechanical properties should change in a gradient way. In this case, tests determining mechanical characteristics of steels elaborated in the conditions of dynamic loads have a crucial meaning [13]. One of the key solutions for increasing discussed mechanical properties of TRIPtype steels is the use of microadditions technology, well controlled in case of conventional HSLA-type steels with ferriticpearlitic, bainitic and tempered martensite microstructure. Beneficial influence of Nb, Ti and V microadditions made to the steel boils to effective taking advantage of precipitation strengthening and grain refinement [3, 5].

Particularly efficient use of microadditions is present in case of sheets production using the method of controlled rolling or controlled recrystallization rolling, where dispersive particles of MX-type interstitial phases precipitating on dislocations limit the grain growth of recrystallized austenite in the intervals between successive roll passes [5, 14]. Certain limitation is minimal thickness of hot-rolled sheets, which in modern processing lines reaches 1.7 mm [5]. Moreover, lower accuracy of surface predisposes these sheets for elements of internal structure of a car. However, they have a wide range of possibilities of potential application for miscellaneous parts reinforcing doors, hood, cross and longitudinal beams, reinforcements of A and B pillars [15]. Application of hot-rolled sheets for parts of chassis, such as lower balance lever - from which high stretch-flangeability is required is especially important [16]. It was found that the lower is the strength difference between ferritic matrix and forming martensite the higher is the stretch-flangeability [11]. Strengthening of ferritic matrix through dispersive precipitations of carbonitrides and strain hardening coming from rolling finish below the temperature of austenite recrystallization are the cause of intense strengthening of  $\alpha$  phase. Solution in this aspect is also substitution of ferritic matrix with bainite [9-16].

Obtaining fine-grained ferritic-bainitic microstructure with retained austenite is strictly connected to the microstructure of primary austenite formed during hot-working. Indispensable condition for microstructure refinement of steel is to adjust the conditions of hot-working to chemical composition of steel allowing controlled course of recrystallization in the intervals between successive sequences of deformation. These problems were well examined for conventional C-Mn and microalloyed steels as well as IF (Interstitial Free) and DP (Dual Phase) type steels. Subjects of the research were power - force parameters of hot-rolling and physical and mathematical modelling of microstructure development in successive cycles of plastic deformation [17-24]. In case of TRIP-type steels equally important is to provide proper cooling conditions allowing obtaining desired participation of individual phases with the adequate morphology. Despite large number of elaborated TRIPtype steels, not many publications referring to this problem can be found in the literature [25]. Cooling conditions of sheets after rolling finishing have to be adjusted to the diagrams of supercooled austenite transformations dependent on chemical composition of steel. High possibilities of shaping complex multiphase microstructure of steel are present in case of  $\gamma \rightarrow \alpha$  and  $\gamma \rightarrow$  bainite transformations put forward to short times, at simultaneous delay of pearlitic transformation start. It should be taken into account that translation of  $\gamma \rightarrow \alpha$  transformation to the left occurs together with increased concentration of Si, Al and P in steel, while C and Mn influence in the opposite direction. Silicon, aluminium and niobium also increase Ac<sub>3</sub> temperature of steel. In addition, silicon displaces pearlitic transformation to the left, however C, Mn, Al, Cr, Mo, Ni, Nb and B influence in the opposite direction [15, 23]. In the aspect of obtaining carbide-free bainite, Al, Si as well as microaddition of Nb have a particular meaning [1, 3, 11]. It has to be also taken into consideration - the increase of martensite start temperature caused by Al - which continuous decreasing to room temperature during successive stages of heat treatment is necessary for stabilization of adequate fraction of retained austenite. Special practical meaning for selection of specific course of cooling have diagrams of supercooled plastically deformed austenite transformations, giving consideration to the influence of plastic deformation performed in the last cycle of hot-working [3, 25, 26].

## 2. Experimental procedure

Chemical composition of elaborated steels was optimized in respect of obtaining optimal participation of retained austenite in the conditions of hot-working and controlled cooling. Steels will be used for working out the conditions of thermo-mechanical processing allowing to obtain ferritic-bainitic microstructures with retained austenite. Additionally, the conditions necessary for obtaining bainitic microstructure with retained austenite of medium-carbon steel will be elaborated. Chemical composition of worked out steels is shown in Table 1. Steels were melted in Balzers VSG-50 vacuum induction furnace employing modification of non-metallic inclusions with Ce and La. Liquid metal was cast from the top into ingot moulds with 25 kg capacity in argon shield. Ingots, after solidification, cutting off the top and base, were subjected to hot-forging into flat bars with 20x200 mm section.

#### Table 1.

Chemical composition of the investigated steels

Mass contents, (%)							
С	Mn	Si	Al	Р	S	Nb	Ti
0.24	1.55	0.87	0.40	0.010	0.004	0.034	0.023
0.43	1.45	0.98	1.00	0.010	0.004	0.033	0.010

Critical steel temperatures,  $Ac_3$  and  $Ac_1$  as well as diagrams of supercooled, non-deformed and plastically deformed austenite transformations were assigned using DIL805A/D dilatometer with LVDT gauge head. Tubular samples,  $\emptyset$ 4x10mm, with 3mm diameter hole were used for CCT (Continuous Cooling Transformations) diagrams determination, while for CCT diagrams after plastic deformation,  $\emptyset$ 4x7 mm cylindrical specimens were applied. Thermal stabilization of samples consisted in their heating with a rate of 10°C/s to the temperature of 650°C, in which they were held for 10min and then cooled to ambient temperature with a rate of 0.5°C/s. After thermal stabilization specimens were heated with a rate of 3°C/s to austenitizing temperature 1100°C and successively austenitized for 300s. Subsequently, samples were cooled with a rate of 5°C/s to the temperature of 900°C being the start of controlled cooling or plastic deformation. Deformation of cylindrical samples was conducted with the reduction of 50% and 1s<sup>-1</sup> strain rate. Vickers hardness of samples cooled in a range from 600 to 1°C/s was measured at 100 N load. Microstructure observations were performed at 400x magnification, after etching the samples in Nital.

# 3. Results and discussion

The diagram of supercooled austenite transformations of steel containing 0.24%C together with microstructures of samples cooled from the temperature of 900°C with a rate in a range from  $100^{\circ}$ C/s to 2°C/s are presented in Fig. 1a-g.



Fig. 1. Diagram of supercooled austenite transformations of steel containing 0.24%C - (a), with structures obtained after cooling from 900°C with a rate: (b) 100°C/s, (c) 20°C/s, (d) 10°C/s, (e) 7°C/s, (f) 4°C/s, (g) 2°C/s



Fig. 2. Diagram of supercooled plastically-deformed austenite transformations of steel containing 0.24%C - (a), with structures obtained after cooling from deformation temperature of 900°C with a rate: (b) 94°C/s, (c) 20°C/s, (d) 10°C/s, (e) 7°C/s, (f) 4°C/s, (g) 2°C/s

High  $Ac_3 = 932^{\circ}C$ , confirms the influence of aluminium, silicon as well as Nb and Ti on raise of this temperature. The difference of 200°C between  $Ac_3$  and  $Ac_1$  temperatures creates high possibilities to control ferrite and austenite participation during intercritical annealing of cold-rolled sheets in conventional production method of multiphase microstructure sheets.  $M_s$  temperature of the steel is equal 360°C and it's relatively low despite increased concentration of Al in the steel. It's beneficial to begin bainitic transformation in comparatively short time and start pearlitic transformation only after 80s, whereas it's less profitable to start  $\gamma \rightarrow \alpha$  transformation only after 20s. Translation of bainitic transformation to the left results in attaining martensitic-bainitic microstructures in a wide range of cooling rates, from approximately 100 to 20°C/s (Fig. 1a-c).

Decrease of the cooling rate to 10°C/s results in entering into a range of  $\gamma \rightarrow \alpha$  transformation (Fig. 1a). However, participation of acicular ferrite is minimal (Fig. 1d). The situation doesn't change much along with further decrease of the cooling rate to 7°C/s (Fig. 1e). Increase of  $\alpha$  phase participation up to approximately 15% occurs together with a decrease of the cooling rate to 4°C/s (Fig. 1f). It's the minimal cooling rate not causing entering in a range of pearlitic transformation. Slight portion of ferrite present in samples cooled in a range of the cooling rate from 10 to 4°C/s is a sequence of short time for realization of  $\gamma \rightarrow \alpha$  transformation. It is maximally equal 30s. After cooling the steel with the rate of 2°C/s pearlite appears in microstructure and ferrite begins to take a form of equiaxial grains (Fig. 1g).

Plastic deformation of steel essentially changes the shape of supercooled plastically deformed austenite transformations diagram (Fig. 2a). First of all, enlargement of the ferritic bay can be noted. In particular, increase of the temperature of austenite into ferrite transformation occurs apart from a rate of cooling the samples from temperature of plastic deformation. Moreover, decrease of bainitic transformation start temperature and translation of ferritic bay to shorter times can be observed. Displacement of pearlitic transformation in a direction of shorter times is also significant. Nevertheless, thanks to acceleration of ferritic transformation, time that can be used for realization of  $\gamma \rightarrow \alpha$  transformation is still equal 70s. In practice, these factors create considerably higher chances of influencing the participation and morphology of ferrite after hot-rolling finish. Ferrite is present in the microstructure of steel deformed in 900°C already after cooling with highest cooling rate applied (Fig. 2b).

Decrease of cooling rate to  $20^{\circ}$ C/s results in obtaining approximately 15% of fine-grained ferrite (Fig. 2c), while for non-deformed samples the same participation of  $\alpha$  phase was achieved after decreasing the cooling rate to 4°C/s. Increase of



Fig. 3. Diagram of supercooled austenite transformations of steel containing 0.43%C - (a), with structures obtained after cooling from 900°C with a rate: (b) 100°C/s, (c) 20°C/s, (d) 10°C/s, (e) 7°C/s, (f) 4°C/s, (g) 2°C/s

ferrite participation to around 40% occurs together with further decrease of the cooling rate to 10 and 7°C/s (Fig. 2d,e). Application of cooling rate being equal 4°C/s leads to the initiation of pearlitic transformation (Fig. 2f). The portion of pearlite increases along with a decrease of the cooling rate to 2°C/s (Fig. 2g). Apart from the increase of ferrite fraction, meaningful refinement of structure in a whole range of cooling rate was obtained. It is a result of considerable increase of population of areas convenient for  $\alpha$  phase nucleation, which are not only the boundaries of austenite grains but also deformation bands formed in plastically deformed austenite [1,3,5].

The CCT diagram for medium-carbon steel together with structures of samples cooled with the rate from 100 to  $2^{\circ}$ C/s are presented in Fig. 3a-g. In comparison with low-carbon steel, medium-carbon steel has a similar Ac<sub>3</sub> temperature and considerably lower Ac<sub>1</sub> temperature which is equal 689°C. It's probably caused by increased concentration of Al in the steel. Decreasing the temperature of austenite into ferrite transformation finish to around 530°C, but first of all, earlier initiation of ferritic transformation after approx. 6s is also beneficial (Fig. 3a). This impact should be ascribed once more to the high concentration of Al. Although pearlitic transformation

starts slightly faster than in the case of low-carbon steel, maximal time for realization of  $\gamma \rightarrow \alpha$  transformation is equal about 65s. Higher concentration of C in the steel causes decrease of M<sub>s</sub> temperature to 300°C. These factors allow shaping multiphase microstructure of the steel during cooling from austenitizing temperature.

Change of the cooling rate in a range from 100 to 7°C/s from austenitizing temperature doesn't influence much microstructures of steel presented in Fig. 3b-e. Its microstructure is martensiticbainitic with slight participation of acicular ferrite. Lack of clear change of microstructure is also reflected by minor variations of hardness in a range from 648 to 619HV. Advantageous selection of chemical composition in the aspect of stabilization of retained austenite to room temperature is confirmed by high participation of this phase, located mainly between laths of martensite and bainite. Decrease of the cooling rate to  $4^{\circ}$ C/s leads to appearance of ferrite in a granular form along previous grains of primary austenite (Fig. 3f). However, pearlite is the disadvantageous constituent of microstructure. Further decrease of the cooling rate to  $2^{\circ}$ C/s results in the increase of pearlite participation, at negligible increase of ferrite fraction (Fig. 3g).



Fig. 4. Diagram of supercooled plastically-deformed austenite transformations of steel containing 0.43%C - (a), with structures obtained after cooling from deformation temperature of 900°C with a rate: (b) 88°C/s, (c) 20°C/s, (d) 15°C/s, (e) 10°C/s, (f) 4°C/s, (g) 2°C/s

Plastic deformation of samples made of steel containing 0.43%C also changes significantly the shape of supercooled plastically deformed austenite transformations diagram (Fig. 4a). Considerable raise of start temperature of austenite to ferrite transformation at simultaneous decrease of finish temperature of this transformation, especially in a range of cooling rate from 20 to 57°C/s, should be listed among advantageous factors. A positive result of this phenomenon is a high possibility to control the temperature of transition through  $\gamma \rightarrow \alpha$  transformation during manufacturing process of multiphase sheets. For instance, the difference between start and finish temperatures of ferritic transformation for cooling rate in a range from 15 to 30°C/s is equal 200°C. It has also an essential meaning for assurance of higher temperature tolerances in relation to those planned during cooling the sheets after rolling is finished. In the aspect of production of sheets with bainitic microstructure with retained austenite, big region of bainitic bay is also beneficial, allowing controlling, to a high degree, the temperature of isothermal holding. In the industrial conditions, this stage complies with coiling of hot-rolled sheets in coils.

Cooling the steel from deformation temperature of 900°C with a rate of 88°C/s results in obtaining martensitic-bainitic microstructure with small participation of acicular ferrite (Fig. 4b). Decrease of the cooling rate to 20°C/s causes an increase of ferrite fraction in a form of fine equiaxial granules arranged in the martensitic-bainitic matrix (Fig. 4c). The minimal cooling rate of steel that doesn't cause entering in a range of pearlitic transformation is equal 15°C/s (Fig. 4a), for which the portion of finegrained ferrite increased up to around 30% (Fig. 4d). Great participation of ferrite in microstructure of samples cooled with cooling rates mentioned above results in a decrease of hardness of the steel when compared to plastically non-deformed samples in the temperature of 900°C. Plastic deformation, similarly as in case of low-carbon steel, positively influences refinement of microstructure of the steel, however causing an acceleration of pearlitic transformation by approximately 40 s. Taking into account that in the case of low-carbon steel the time for initiation of pearlitic transformation was shortened only by 15s, significant acceleration of pearlitic transformation for medium-carbon steel should be ascribed to higher concentration of carbon in this steel. The sample cooled with a rate of 10°C/s contains higher fraction of ferrite as well as

first traces of pearlite (Fig. 4e). Decrease of the cooling rate to  $4^{\circ}C/s$  results in obtaining significant participation of pearlite, however, at still meaningful portion of bainite (Fig. 4f). Whereas decrease of the cooling rate to  $2^{\circ}C/s$  leads to obtaining fine-grained ferritic-pearlitic microstructure with small fraction of bainite (Fig. 4g).

# **4.**Conclusions

The shape of curves of supercooled austenite transformations diagram of elaborated steels predisposes them for multiphase sheets manufacturing. Both steels have high Ac3 temperature and a great difference between Ac3 and Ac1. It creates high possibilities to control ferrite and austenite participation during intercritical annealing of cold-rolled sheets. These steels possess ferritic and bainitic bays put forward to short times and pearlitic regions put aside. For this reason, they have ferritic-bainitic-martensitic microstructures in a wide range of the cooling rate. Disadvantageous pearlite appears only after cooling with a rate of 3°C/s and 4°C/s – for low- and medium-carbon steel, respectively. The essential microstructure constituent of both steels is presence of retained austenite, located most often between laths of bainite and martensite. Despite advantageous shape of ferritic bay, participation of ferrite is very slight; it occurs mainly as acicular ferrite. Grains of equiaxial ferrite appear only at cooling rates which also favour formation of pearlite - unwanted for TRIP-type steels.

Plastic deformation of steel in the temperature of 900°C has a definitely profitable influence on the shape of supercooled austenite curves. The region of  $\gamma \rightarrow \alpha$  transformation is translated to the left at simultaneous raise of start temperature of austenite into ferrite transformation. For identical cooling conditions, ferrite participation is definitely higher and for cooling with the lowest rate, not allowing initiation of pearlitic transformation, it's equal around 30%. In the result of plastic deformation, meaningful refinement of microstructure in a whole range of cooling rate was also obtained. Moreover, the effect of plastic strain is a slight acceleration of pearlitic transformation for steel containing 0.24%C and even more meaningful for medium-carbon steel. Despite this, regions of  $\gamma \rightarrow \alpha$ transformation are big enough to permit to control precisely the course of austenite to fine-grained ferrite transformation, both in respect of its participation and morphology. Further increase of ferrite fraction can be achieved by modification of the cooling method after hot-working finish. In the fist stage, steel should be rapidly cooled in order to enter the range of  $\gamma \rightarrow \alpha$  transformation and successively slowly cooled in this range making maximum use of the time for diffusive transformation of austenite into ferrite. It will be the subject of further research.

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