Archives of Materials Science and Engineering

Volume 32
Issue 1
July 2008
Pages 13-16

International Scientific Journal published monthly by the World Academy of Materials and Manufacturing Engineering

Diagrams of supercooled austenite transformations of low-carbon and medium-carbon TRIP-steels

A. Grajcar*, M. Opiela
Division of Constructional and Special Materials, Institute of Engineering Materials and Biomaterials, Silesian University of Technology, ul. Konarskiego 18a, 44-100 Gliwice, Poland
* Corresponding author: E-mail address: adam.grajcar@polsl.pl

Received 20.03.2008; published in revised form 01.07.2008

ABSTRACT

Purpose: The aim of the paper is to determine the influence of cooling conditions on a structure and a shape of CCT-diagrams of TRIP-aided steels.

Design/methodology/approach: The diagrams of undeformed supercooled austenite transformations for low-carbon and medium-carbon steels were determined. The specimens were austenitized at a temperature of 1100°C and cooled from a temperature of 900°C with a rate in a range from 1 to 300°Cs^{-1}. The dilatometric tests were carried out by the use of the DIL805A/D dilatometer with a LVDT-type measuring head.

Findings: It was found that obtained CCT-diagrams of low-carbon and medium-carbon steels are favourable for manufacturing TRIP-type steels with multiphase structures. The steels are characterized by large ferritic and bainitic fields and a right-displaced pearlitic range. However, a ferrite fraction obtained after cooling with an optimum rate from a temperature of 900°C is low. Increasing the fraction of the α phase requires two-stage cooling after austenitizing.

Research limitations/implications: To obtain the optimum ferrite fraction, it is necessary to modify a cooling course in a range of $\gamma \rightarrow \alpha$ transformation. It should result in an effective utilization of the time for the transformation of austenite into the fine-grained ferrite.

Practical implications: The obtained diagrams of supercooled austenite transformations can be useful in a determination of a cooling course from a finishing rolling temperature for sheets with a multiphase structure.

Originality/value: The diagrams of the undeformed supercooled austenite for the low-carbon and medium-carbon steels containing Nb and Ti microadditions were obtained.

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

MATERIALS

1. Introduction

The research performed in recent decade in order to elaborate new groups of steels with profitable combination of high strength, ductility and formability have led to a development of multiphase steels with TRIP effect [1-3]. Conducted investigations generally concerned elaboration of optimal chemical composition, particularly in the aspect of carbon, silicon and aluminium concentration, having decisive influence on stabilization of adequate fraction of retained austenite [4, 5]. The second direction of research refers to elaboration of optimal conditions of heat treatment, allowing to obtain retained austenite with proper morphology and mechanical stability in the conditions of further cold plastic deformation [6, 7]. Investigations of strain-induced transformation of retained austenite into martensite were conducted in conditions of static and dynamic deformation. In case of sheets production from TRIP-type steels applying thermo-mechanical processing method, it's necessary to know the behaviour of this group of steels in the conditions of hot plastic deformation.

© Copyright by International OCSCO World Press. All rights reserved. 2008

Obtaining fine-grained microstructure of products before their cooling from the temperature of hot-working finishing requires knowledge of hot-working conditions influence on work-hardening and thermally activated processes removing the effects of hardening. Determination of power – force parameters of a process necessary for elaboration of adequate technologies of metallurgical processing is also essential. This kind of examinations is rarely performed for TRIP-type steels. So far, a great attention was devoted to behaviour of conventional C-Mn steels as well as microalloyed steels during hot-working [8, 9]. Recently, hot-working technologies of IF (Interstitial Free) and DP-type (Dual Phase) steels have been elaborated [10, 11]. These problems are a subject of physical and mathematical modelling and also enclose the influence of plastic deformation parameters on microstructure development before beginning of cooling from a temperature of finishing rolling [12, 13].

In case of multiphase steels, the key problem is to obtain a proper fraction of retained austenite with a specified size, morphology and mechanical stability. In [14, 15] works it was found out that grain refinement of austenite microstructure as well as steel deformation in a range of austenite and ferrite stability has a profitable impact on retained austenite stability. Therefore it's purposeful to elaborate chemical composition of C-Mn-Si-Al steels containing microadditions with high chemical affinity to carbon and nitrogen. In properly selected conditions of hot-working they allow achieving fine-grained microstructure of austenite, inherited during further cooling of the steel. Equally important part of production of TRIP-type steels is to control temperature – time conditions of several-stage cooling. It's necessary to adjust the conditions of cooling to the diagram of supercooled austenite transformations. It guarantees acquiring adequate fraction of ferrite, type of bainite, and therefore proper thermal stability of γ phase connected to a decrease of M_s temperature of retained austenite below room temperature. A very limited number of CCT diagrams can be found in the literature [16], despite a wide range of chemical composition of TRIP-type steels produced.

2. Experimental procedure

The investigations were carried out on new-developed TRIP-type steels with a chemical composition given in Table 1. The melts were dove in Balzers VSG-50 vacuum induction furnace. Liquid metal was cast into moulds in argon atmosphere. Obtained 25 kg ingots were subjected to open die forging to 220 mm wide and 20 mm thick flats, from which ϕ4x7 mm samples with a hole of 3 mm were made.

Dilatometric analyses were performed with the help of DIL805A/D dilatometer equipped with LVDT-type gauge head. Before starting the experiment samples were subjected to thermal stabilization, i.e. they were heated to the temperature of 650°C with a rate of 10°C/s, then held for 10min and cooled to ambient temperature with a rate of 30°C/min. Specimens were austenitized in temperature of 1100°C for 30s and cooled to temperature of the beginning of cooling 900°C with a rate of 5°C/s. Samples were cooled with a rate in a range from 600 to 1°C/s. Vickers hardness was measured with 100 N load applied. Specimens were etched in nital, and photos were taken at 400x magnification.

Table 1.
Chemical composition of the investigated steel

Mass contents in percentage (%)							
C	Mn	Si	Al	P	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

3. Results and discussion

The diagram of supercooled austenite transformations of elaborated low-carbon steel is presented in Fig. 1. From its analysis it arises, that the steel possesses rather high A_{c3} temperature, being equal 932°C. It's a result of high aluminium concentration in the steel and Nb and Ti microadditions influence. Considerable difference between A_{c1} and A_{c3} temperatures creates high possibilities of precise control of ferrite to austenite proportion during intercritical annealing of sheets previously subjected to cold-rolling. The temperature of martensitic transformation start is equal 360°C what is typical for steels with 0.24% of carbon concentration. Cooling the steel with a rate of approximately 100°C/s results in obtaining martensitic microstructure with hardness of around 490 HV. Bainitic bay put forward in a direction of short times guarantees achieving bainitic-martensitic microstructures in a wide range of cooling rates, from about 100 to 20°C. It's interesting that the hardness of steel cooled with a rate of 20°C/s is only slightly lower when comparing to the sample cooled from the austenitizing temperature with a rate being five times higher. The microstructure of the steel cooled with a rate of 20°C/s is shown in Fig. 2. Cooling with a lower rate results in a presence of ferrite in microstructure of the steel (Figs. 3-5). It's profitable to start pearlitic transformation only after approximately 70s. It indicates the effective influence of Al and Si. For this reason the time from the beginning of ferritic transformation to the beginning of pearlitic transformation is equal to around 60s. However steel cooling from the temperature of 900°C with constant rate of 10°C/s causes obtaining nearly homogeneous microstructure of bainite (Fig. 3).

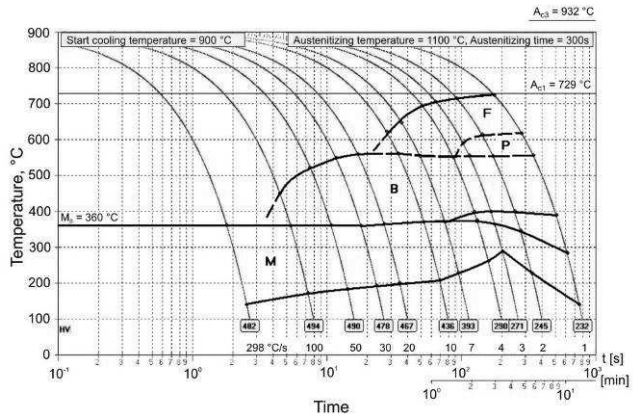

Fig. 1. Diagram of supercooled austenite transformations of elaborated steel containing 0.24%C (CCT-diagram)

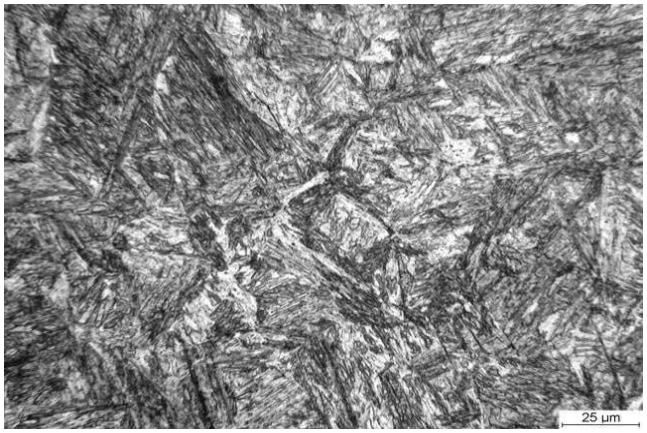

Fig. 2. Bainitic-martensitic microstructure of the steel containing 0.24%C cooled with a rate of 20°Cs^{-1} from 900°C

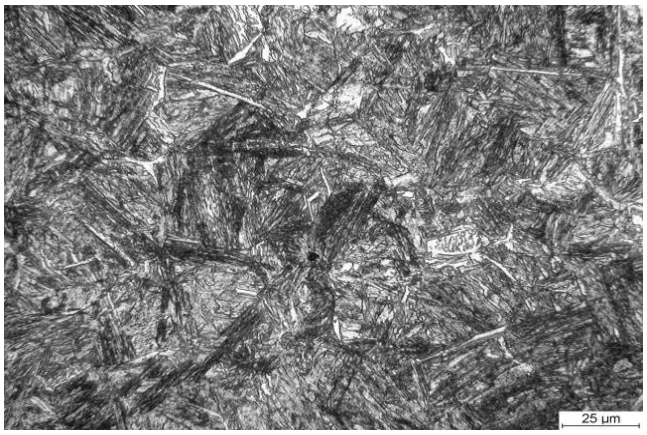

Fig. 3. Bainitic microstructure of the 0.24%C steel containing small fractions of ferrite, martensite and retained austenite cooled with a rate of 10°Cs^{-1} from 900°C

Fig. 4. Bainitic-ferritic microstructure of the 0.24%C steel containing retained austenite cooled with a rate of 7°Cs^{-1} from 900°C

Fig. 5. Bainitic-ferritic microstructure of the 0.24%C steel containing retained austenite cooled with a rate of 4°Cs^{-1} from 900°C

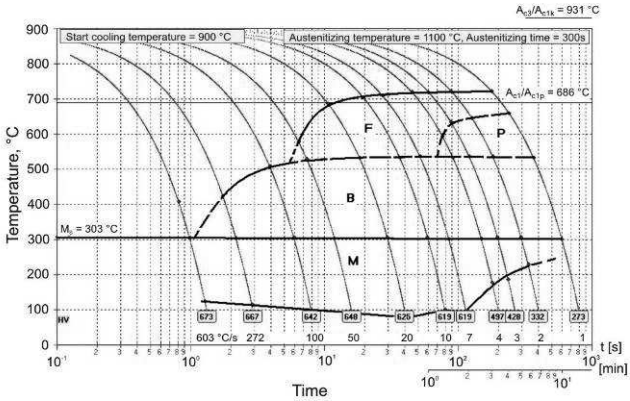

Fig. 6. Diagram of supercooled austenite transformations of elaborated steel containing 0.43%C (CCT-diagram)

Also a decrease of a cooling rate to 7°C/s doesn't lead to achieving proper participation of ferrite (Fig. 4). Phase α has a lath character and is arranged only between lathes of bainite. Even cooling with lowest rate not causing entering into a range of pearlitic transformation doesn't result in obtaining proper portion of ferrite, being equal around 15% (Fig. 5). It's caused by too short time for γ→α transformation and for instance for cooling rate of 4°C/s is equivalent only 30s. Optimal use of time for austenite into ferrite transformation would require applying two-stage course of cooling after austenitizing. In the first stage steel would have to be cooled with a rate of approximately 15°C/s to the temperature of around 650°C and then a very slow cooling in retention shield for possibly long time being equal approximately 60s should be applied – however not allowing pearlitic transformation to begin. Further decrease of a cooling rate below around 3°C/s results admittedly in an increase of ferrite participation up to around 30%, however along with unfavourable portion of pearlite. On the other hand, a profitable phenomenon is occurrence of small particles of retained austenite in the microstructure, which can be observed, for instance, in a presence of martensite or at interfacial grain boundaries of bainite and ferrite.

Increase of carbon and aluminium concentration in 0.43C-1.5Mn-1Si-1Al-Nb-Ti steel causes essential changes in the diagram of supercooled austenite (Fig. 6). A_{c3} temperature is just as high as in case of low-carbon steel and is equal 931°C. A_{c1} temperature decreased to 689°C what can be ascribed to the impact of high Al concentration. Considerable difference between A_{c1} and A_{c3} creates a possibility of precise controlling α and γ phase participation during intercritical annealing being the first stage of several-stage heat treatment of TRIP steels produced in a conventional way. Increase of Al concentration causes significant displacement of γ→α transformation beginning to shorter times with simultaneous increase of temperature of the beginning of supercooled austenite transformation into ferrite. In spite of considerably higher concentration of carbon, pearlitic transformation starts only after approximately 70s, what creates convenient conditions for obtaining proper participation of α phase after the end of hot rolling. Obvious effect of increased carbon concentration is a decrease of temperature of martensitic transformation beginning to 300°C. Hardness of steel containing 0.43% C is resolutely higher than the hardness of steel containing 0.24%C - in a whole range of cooling rate.

4. Conclusions

Elaborated steels have a profitable shape of supercooled austenite transformations diagrams for obtaining TRIP-type multiphase microstructure. Considerable difference between A_{c1} and A_{c3} temperatures creates high possibilities to control ferrite to austenite proportion during intercritical annealing of steel after cold rolling. Moreover, both steels are characterized by ferritic transformation put forward to short times and pearlitic bay moved to the right. It's mainly connected to beneficial influence of increased concentration of Al in steels. However, cooling the steel with a constant rate from austenitizing temperature doesn't lead to obtaining proper participation of ferrite. It's connected with too short time for realization of γ→α transformation. It's predicted that increase of α phase fraction can be achieved through application of rapid cooling of steel to the temperature of the beginning of γ→α transformation and slow cooling of the steel for around 50s in a range of this transformation – not allowing for initiation of pearlitic transformation.

Acknowledgements

Scientific work was financed from the science funds in a period of 2006-2008 in the framework of project No. 3 T08A 080 30.

References

[1] I.B. Timokhina, P.D. Hodgson, E.V. Pereloma, Effect of alloying elements on the microstructure-property relationship in thermomechanically processed C-Mn-Si TRIP steels, Steel Research 73 (2002) 274-286.

[2] A.K. Lis, B. Gajda, Modelling of the DP and TRIP microstructure in the CMnAlSi automotive steel, Journal of Achievements in Materials and Manufacturing Engineering 15 (2006) 127-134.

[3] J. Adamczyk, A. Grajcar, Structure and mechanical properties of DP-type and TRIP-type sheets obtained after the thermomechanical processing, Journal of Materials Processing Technology 162-163 (2005) 267-274.

[4] A. Grajcar, Effect of hot-working in the γ+α range on a retained austenite fraction in TRIP-aided steel, Journal of Achievements in Materials and Manufacturing Engineering 22/2 (2007) 79-82.

[5] B. Gajda, A.K. Lis, Thermal processing of CMnAlSi steel at (α+γ) temperature range, Journal of Achievements in Materials and Manufacturing Engineering 18 (2006) 355-358.

[6] S. Wen, L. Lin, B. De Cooman, P. Wollants, Y. Chun-Xia, Thermal stability of retained austenite in TRIP steel after different treatments, Journal of Iron and Steel Research 15 (2006) 61-64.

[7] A. Grajcar, Determination of the stability of retained austenite in TRIP-aided bainitic steel, Journal of Achievements in Materials and Manufacturing Engineering 20 (2007) 111-114.

[8] J. Majta, R. Kuziak, M. Pietrzyk, Modelling of the influence of thermomechanical processing of Nb-microalloyed steel on the resulting mechanical properties, Journal of Materials Processing Technology 80-81 (1998) 524-530.

[9] M. Opiela, Thermo-mechanical treatment of the C-Mn steel with Nb, Ti, V and B microadditions, Archives of Materials Science and Engineering 28 (2007) 377-380.

[10] E. Hadasik, R. Kuziak, R. Kawalla, M. Adamczyk, M. Pietrzyk, Rheological model for simulation of hot rolling of new generation steel strip for automotive applications, Steel Research 77 (2006) 927-933.

[11] R. Kawalla, G. Goldhahn, W. Jungnickel, Laboratory rolling condition and its effect on rolling parameters on material properties of hot rolled products, Proceedings of the 2nd International Conference "Thermomechanical Processing of Steel" TMP'2004, Liege, 2004, 17-24.

[12] T. Siwecki, Physical metallurgy and modeling of thermomechanically processed microalloyed steels, Materials Engineering 3 (1998) 163-170.

[13] P. Uranga, B. Lopez, J-M. Rodriguez-Ibabe, Microstructural modelling of Nb microalloyed steel during thin slab direct rolling processing, Steel Research 78 (2007) 199-209.

[14] B. Gajda, A.K. Lis, Intercritical annealing with isothermal holding of TRIP CMnAlSi steel, Journal of Achievements in Materials and Manufacturing Engineering 20 (2007) 439-442.

[15] O. Muransky, P. Hornak, P. Lukas, J. Zrnik, P. Sittner, Investigation of retained austenite stability in Mn-Si TRIP steel in tensile deformation condition, Journal of Achievements in Materials and Manufacturing Engineering 14 (2006) 26-30.

[16] M. Zhang, L. Li, R.Y. Fu, D. Krizan, B.C. De Cooman, Continuous cooling transformation diagrams and properties of micro-alloyed TRIP steels, Materials Science and Engineering A 438-440 (2006) 296-299