CONTINUOUS COOLING DIAGRAM OF NEW-DEVELOPED HIGH-STRENGTH STEEL MICROALLOYED WITH Nb, Ti, V AND B

Ph.D. Opieła M.¹, Ph.D. Zalecki W.², Ass.Prof. Grajcar A.¹
Institute of Engineering Materials and Biomaterials – Silesian University of Technology, Poland¹, Institute for Ferrous Metallurgy, Poland²

Abstract: The aim of the paper is to determine the influence of cooling conditions on a structure and a shape of CCT-diagram of new-developed HSLA steel with Nb, Ti, V and B, assigned for production of forged machine parts using the method of thermo-mechanical treatment. The dilatometric tests were carried out by the use of the DIL 805A/D dilatometer with a LVDT-type measuring head. The specimens were austenitized at a temperature of 985°C and cooled to ambient temperature at various rates from 234°C to 1°C/min. Dilatometric research revealed that the steel is characterized with  \( A_1 = 843°C \), \( A_3 = 707°C \) and a relatively low M temperature equal 370°C. The obtained CCT diagrams of supercooled austenite transformations can be useful in determination of cooling conditions of the thermo-mechanical processing for high strength forged machine parts from microalloyed steels.

Keywords: MICROALLOYED STEEL, CCT-DIAGRAM, SUPERCOOLED AUSTENITE

1. Introduction

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

Economic considerations determine that the majority of forgings for automotive industry, mining, agricultural and other machines are currently produced of ferritic-pearlitic microalloyed steels. Steel designated as 49MnV53 containing 0.44–0.54%C, up to 0.6%Si, 0.6–1.0%Mn, 0.045–0.065%S and 0.08–0.13%V, characterized with YS > 450 MPa, UTS from 780 to 900 MPa and DVM impact energy of specimens ranging from 15 to 30 J, was the first grade of micro-alloyed steel used for an engine crankshaft in Thyssen Edelstahlwerke [9]. Such high mechanical properties of forged parts can be achieved by appropriate selection of forging conditions, i.e. temperature of charge heating and plastic deformation, since the distribution of strains and strain rate during production of complex shape die forgings is difficult to be adjusted. The conditions of charge heating for forging should not lead to total dissolution of interstitial phases of microadditions introduced into steel in a solid solution for it causes disadvantageous grain growth. Deformation at high rate and short duration intervals for moving the produced part from one die impression to another do not create convenient conditions for the course of static recrystallization, allowing refinement of austenite grains. Indeed, \( \gamma \rightarrow \alpha \) transformation of both thick- and fine-grained plastically deformed austenite, begins on grain boundaries, twin boundaries and deformation bands, in case of coarse-grained \( \gamma \) phase it doesn’t assure sufficiently fine-grained microstructure and expected mechanical properties of forged parts. Forgings produced under such conditions, free-air cooled from the of plastic working finish temperature, admittedly obtain high strength as a result of strong precipitation hardening, but also low crack resistance. An effective way to increase ductility and strength of ferritic-pearlitic steel is to obtain a microstructure consisting of ultrafine excess ferrite and finest areas of pearlite limited with narrow angular boundaries, which are individual colonies or areas enclosing several neighbouring colonies. This can be realized through transformation of austenite with finest grains and decrease of ferritic and pearlitic transformation temperature. The studies on the increase of toughness of micro-alloyed ferritic-pearlitic steels have led to development of grades with decreased concentration of carbon. An example of such grade is 27MnSiV6 steel containing 0.25–0.30%C, 1.30–1.60%Mn, 0.5–0.8%Si, 0.030–0.050%S and 0.08–0.13%V. This steel is characterized with YS > 500 MPa, UTS from 800 to 950 MPa and DVM impact energy ranging from 40 to 60 J [10].

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

Knowledge of supercooled austenite transformations is necessary for proper design of conditions of thermo-mechanical treatment and controlled cooling of forgings from the forging finish temperature. However, classical CCT diagrams have limited usefulness for elaboration of conditions of products cooling from hot-working finish temperature. Diagrams of transformations of supercooled plastically deformed austenite have significant technical usability. For example, studies of influence of plastic deformation on the course of supercooled austenite transformation curves were performed in [16] for steel containing 0.17%C, 1.37%Mn, 0.26%Si, 0.24%Cr, 0.48%Mn and microadditions of Nb, V, Ti and B in the amount of 0.025%, 0.019%V, 0.004% and 0.002%, respectively. Conducted experiments indicated that plastic deformation of austenite prior to transformation caused considerable acceleration of diffusive transformations, i.e. ferritic and pearlitic transformation and leads to shorter duration of bainitic transformation as well as to slight decrease of $M_s$ temperature for investigated steel. Similar issues have been studied in [17-28]. The aim of the paper is to investigate the influence of cooling conditions on a structure and a shape of CCT-diagrams Nb-Ti-V microalloyed steel.

2. Material and experimental procedure

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

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

Evaluation of the influence of hot plastic deformation on phase transformations of supercooled austenite of investigated steel applying continuous cooling of samples was done using dilatometric method. The experiment was performed in the Institute of Ferrous Metallurgy (IFM) in Gliwice, using DIL 805/A/D dilatometer, manufactured by Bähr Thermoanalyse GmbH, equipped with LVDT type measuring head with theoretical resolution of ±0.057 mm. Heating of specimens in dilatometer was realized with the induction method using a generator at frequency of 250 kHz. Both, heating in isothermal holding of samples at assigned temperature were carried out in 5-10⁻³ mbar vacuum, created by rotary and turbomolecular pump. Temporary temperature deviations from assigned value did not exceed ±1.0°C. Temperature measurement was done using S PtRh10-Pt type thermoelement with diameter of wires equal 0.1 mm. Both thermoelement ends were welded onto samples in the middle of their length.

Experiments and analysis of results were performed using the technique consisting in putting a tangent against a dilatation curve in the vicinity of the start and finish of phase transformation. In case of inseparable transformations (occurring one after another) numerical differentiation of dilatation curves was used for the analysis. In case of weak ferritic and pearlitic transformations signals, the IFM developed method based on linear transformation of analyzed section of dilatation curve was applied in order to determine the phase transformation start and finish temperatures. Basing on performed examinations, critical points of steel ($A_{1}, A_{3}$) and $M_s$ were determined as well as ranges of phase transformations of supercooled austenite. Investigation of phase transformations was performed using $44x3x7$ mm tubular specimens. Prior to the experiment, all samples were subjected to thermal stabilization, i.e. they were heated to the temperature of 650°C at the rate of 10°C/s, then held for 600s at this temperature and successively cooled to ambient temperature with the rate of 30°C/min. In case of determination of phase transformations of supercooled austenite, specimens were heated at the rate of 10°C/s up to the temperature of 885°C, being the initiation of controlled cooling. Samples were austenized at this temperature for 600 s and then cooled to ambient temperature at diversified rate, i.e. 234°C/s, 99°C/s, 50°C/s, 20°C/s, 10°C/s, 4°C/s, 2°C/s, 1°C/s, 30°C/min, 15°C/min, 6°C/min, 3°C/min, 1°C/min. In order to identify phases occurrence of products of supercooled austenite transformations, after dilatometric studies, samples were subjected to metallographic analysis using NEOPHOT 2 light microscope with digital image recording, at magnification of 400x and 800x. HV10 hardness of samples was studied using Vickers method applying the load of 98N, implementing Swiss Max 300 universal testing machine.

3. Results and discussion

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

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

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

Acknowledgements

Scientific work was financed from the science funds of the Polish Ministry of Science and Higher Education in a period of 2010-2013 in the framework of project No. N N508 585239.

References


