ARCHIVES OF CIVIL AND MECHANICAL ENGINEERING 12 (2012) 334–341

Available online at www.sciencedirect.com

SciVerse ScienceDirect

journal homepage: www.elsevier.com/locate/acme

Original Research Article

Third generation of AHSS with increased fraction of retained austenite for the automotive industry

A. Grajcar[a,*], *R. Kuziak*[b], *W. Zalecki*[b]

[a]*Silesian University of Technology, Faculty of Mechanical Engineering, Institute of Engineering Materials and Biomaterials, 18a Konarskiego Street, 44-100 Gliwice, Poland*

[b]*Institute for Ferrous Metallurgy, 12-14 Karola Miarki Street, 44-100 Gliwice, Poland*

ARTICLE INFO

Article history:
Received 30 May 2012
Accepted 17 June 2012
Available online 23 June 2012

Keywords:
TRIP-aided steel
AHSS
DCCT diagram
Stress–strain curves
Retained austenite

ABSTRACT

Third generation of advanced high-strength steels for the automotive industry contains a high volume fraction of fine-grained ferrite, carbide-free bainite, martensite and retained austenite. The level of strength and ductility is highly dependent on the fraction and mechanical stability of austenitic phase. One of the methods to obtain an increased fraction of γ phase is trough its chemical stabilization by Mn. Two 0.17C–3Mn–1.5Al–0.2-Si–0.2Mo steels with and without Nb microaddition were developed in the present study. The steels were subjected to the thermomechanical processing designed on the basis of the DCCT diagram (deformation – continuous cooling transformation). The paper presents the results of the multi-stage compression tests and multiphase microstructures obtained as a result of the controlled multi-stage cooling. It was found that the hot workability of a new generation of AHSS is very challenging due to high values of flow stresses required. However, the thermomechanical processing enables to obtain very fine-grained bainite-based microstructures with a fraction of retained austenite up to 20%. The highest fraction of fine grains and interlath austenite was obtained for the temperature range between 400 and 450 °C. The effect of Nb results in higher flow stresses and better distribution of all the microstructural constituents.

1. Introduction

Third generation of advanced high strength steels (AHSSs) for the automotive industry is a further step in development of modern high strength–ductility balance steel sheets. They utilize complex interaction of solid solution hardening, precipitation hardening, microalloying, grain refining, strain aging and TRansformation Induced Plasticity (TRIP) or TWinning Induced Plasticity (TWIP) effects. Third generation AHSS combine the advantages of multiphase structures (characteristic for 1st generation AHSS) [1–6] and austenitic phase, especially suitable to enhance different hardening mechanisms, strain-induced martensitic transformation and mechanical twinning (characteristic for 2nd generation AHSS) [7–11]. In spite of excellent mechanical properties of high-Mn austenitic steels belonging to the 2nd generation of AHSS their application will be probably limited only for most challenging autobody elements, for example, specially geometrically-designed

*Corresponding author.
E-mail address: adam.grajcar@polsl.pl (A. Grajcar).

1644-9665/$ - see front matter © 2012 Politechnika Wrocławska. Published by Elsevier Urban & Partner Sp. z o.o. All rights reserved.
http://dx.doi.org/10.1016/j.acme.2012.06.011

sections absorbing energy during crash events [12]. The reasons are various technological problems related to poor casting, hot-working above 1150 °C, corrosion resistance, Mn segregation and especially the high cost due to Mn (between 20 and 30 wt%), Al and Si alloying concept [7–11].

The main idea of 3rd generation of AHSS is to obtain the mechanical properties regime between 1st and 2nd generation of AHSS at cost slightly higher compared to 1st generation multiphase steels (DP, TRIP, CP) [13–15]. New microstructure concepts consist in increasing volume fraction of hard constituents and retained austenite. It is related to replacing polygonal ferrite by acicular or bainitic ferrite, non-carbide bainite, martensite and stabilization of austenitic phase in different ways, i.e., chemically or mechanically. Replacing the polygonal ferrite matrix by non-carbide bainite/martensite improves also the stretch-flangeability of multiphase steel sheets, which is one of the most important technological parameter characterizing their practical application for the auto-body structure [13,16]. One of the chemical strategy concepts to obtain a bainitic matrix containing a high volume fraction of metastable retained austenite is Mn alloying but to a much lesser extent than in high-Mn steels of the 2nd generation AHSS. Manganese is a main austenite stabilizer and its content reaches up to 12 wt% in recently investigated C–Mn–Al–Si steels [13–15]. Increased hardenability of steel due to Mn alloying leads to a considerable decrease in the ferrite fraction as a result of shifting the $\gamma \rightarrow \alpha$ transformation region to the right side on CCT diagrams [17,18]. A heat treatment for steels with increased Mn content is similar to that used for DP and TRIP steels, i.e., intercritical annealing after cold rolling or intercritical annealing and isothermal holding at a bainitic transformation range. The thermomechanical processing, microalloying with Nb, Ti, V, reverse martensite transformation and quenching and partitioning processing are the examples of ideas to obtained fine-grained complex microstructures with a high fraction of austenitic phase of optimal stability for strain-induced martensitic transformation during drawing, stretching, bending, etc. [7,13–15].

In our previous works [19,20] the stress–strain curves and softening kinetics of austenite under conditions of hot-compression for 3Mn–1.5Al and 5Mn–Al steels were investigated. Elaboration of the thermomechanical rolling requires also the knowledge of hot-working behavior of steels during multi-stage deformation. Additionally, especially important for hot-rolled multiphase steels are CCT (Continuous Cooling Transformation) diagrams, which are necessary for proper design of multi-stage cooling after finishing rolling. Despite many DP and TRIP steels investigated, very few CCT diagrams can be found in the literature [17,18,21–24]. Silicon and aluminum increase A_{c3} temperature of steel. In addition, Si shifts pearlitic transformation to the left, however C, Mn, Al, Mo, Cr, Ni and Nb act in the opposite direction. The effect of Nb usually manifests by slight increase of the $\gamma \rightarrow \alpha$ transformation temperature and shortening time to initiate the transformation [17,18].

2. Experimental procedure

The paper addresses the thermomechanical processing of new-developed high-Mn high-Al TRIP steels with and without Nb microaddition. Special attention was paid to the effect of Nb on the hot-working behavior and multiphase microstructure formed at various cooling conditions of the thermomechanical treatment. The chemical compositions of the investigated steels are given in Table 1.

The chemical composition strategy was designed with the focus on the maximizing of retained austenite volume fraction (increased Mn content) and obtaining carbide-free bainite by low-Si high-Al concept [2,3,23]. Special attention was paid to the effect of Nb on the hot deformation resistance and final multiphase microstructures after the thermomechanical processing. The steels were produced by vacuum induction melting in the Balzers VSG-50 furnace. Liquid metal was cast in the Ar atmosphere into a cast iron mould. Ingots with a mass of about 25 kg were forged at temperature range from 1200 to 900 °C to a thickness of 22 mm. Then, cylindrical samples ∅10 × 12 mm for hot compression tests and ∅4 × 7 mm specimens for dilatometric analyses were machined.

Parameters of the thermomechanical processing are shown in Fig. 1. The experiments were carried out using the DSI Gleeble 3800 thermomechanical simulator. The specimens were inserted in a vacuum chamber, where they were resistance-heated to a temperature of 1200 °C. After austenitizing for 30 s the specimens were cooled to a temperature of first deformation. The thermomechanical processing consisted of four deformation steps (1150, 1050, 950, 850 °C) and multi-stage cooling according to Fig. 1. The logarithmic strain value was equal to 0.25 at the strain rate of $10\,s^{-1}$ for each deformation step. After the final deformation at 850 °C the specimens were cooled to the temperature of 700 °C. The essential step of the thermomechanical treatment consisted in the various time of slow cooling in a temperature range 700–650 °C. The time was equal to 10, 25 or 50 s depending on the cooling rate applied (Fig. 1). Next, the samples were cooled at the rate of 40 °C/s (similar to the laminar cooling facilities in industrial conditions) to the holding temperature in the bainite transformation range (T_B). The isothermal holding temperature of 450, 400 or 350 °C was the second variable parameter of the thermomechanical processing. The holding time (t_B) was constant, 300 s. Finally, the specimens were colled with a rate of 0.5 °C/s to room temperature. The whole cooling route was designed on the basis of the DCCT diagram.

Table 1 – Chemical composition of the investigated steels, wt%.

Designation	C	Mn	Al	Si	Mo	S	P	Nb
3Mn–1.5Al–0.2Si–0.2Mo	0.17	3.30	1.70	0.22	0.23	0.014	0.010	–
3Mn–1.5Al–0.2Si–0.2Mo–Nb	0.17	3.10	1.60	0.22	0.22	0.005	0.008	0.04