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Structure–mechanical property relationship in low carbon microalloyed steel plate processed using controlled rolling and two-stage continuous cooling

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ABSTRACT

Controlled rolling followed by two-stage continuous cooling was carried out in-house to study the microstructure and mechanical properties of a low carbon microalloyed steel plate of medium thickness gauge. The objective of the study was to develop a process of obtaining excellent mechanical properties (strength, toughness, and ductility) in a microalloyed steel. The process is industrially viable because it does not require high degree of cooling and large reduction during thermo-mechanical processing. The study demonstrates that it is possible to obtain yield strength, tensile strength, elongation, and percentage reduction in area 585 MPa, 680 MPa, 29.5%, and 55%, respectively, with total deformation reduction of 50% and small first stage cooling rate. The total impact energy at −20 °C was 140 J. The microstructure consists of polygonal ferrite and acicular ferrite. The strain-induced 20–30 nm precipitates act as nucleation sites for acicular ferrite. The acicular ferrite contributes to high strength and good toughness, while polygonal ferrite provides excellent elongation.

1. Introduction

There is a continued effort to develop high strength low alloy (HSLA) steels with high strength–high toughness combination. To obtain good toughness and weldability, carbon content is reduced. The decrease in strength due to lower carbon content is compensated by the addition of Si and Mn. A further increase in strength is acquired through precipitation hardening and refinement of grain size by microalloying with Nb, V, and Ti, individually or in combination [1–3]. The nanoscale microalloyed carbonitrides form during long time holding at 450–650 °C and carbides significantly enhance the yield strength by preventing movement of dislocations. The grain refinement is realized, when stable TiN precipitates pin the prior austenite grain boundary during the reheating process and Nb atoms/NbN precipitates retard the recrystallization of deformed austenite [4,5]. Compared to the Nb–Ti microalloyed steels, V–N steels exhibit grain refinement through intragranular nucleation of ferrite on VN precipitates partly due to low lattice mismatch of VN with ferrite. The addition of N in V microalloyed steel stimulates the precipitation of V carbonitrides and increases their volume fraction. The V–N steels with high S content are used in forging and long products, and the strength and toughness are simultaneously improved by nucleation of acicular ferrite on MnS +VN complex inclusions, where insoluble MnS core provides heterogeneous sites for nucleation of VN [6,7].

A microstructure composing of fine interwoven ferrite laths or plates is generally defined as acicular ferrite. This fine interlocking structure is a preferred microstructure in low carbon HSLA steels because of superior toughness [8]. However, the nucleation of acicular ferrite requires following conditions: they include (1) availability of inclusions for intragranular nucleation; (2) relatively large prior austenite grain size to decrease the probability for nucleation of intergranular phase; (3) relatively small deformation reduction; and (4) adequate cooling rate to keep austenite stable up to bainite transformation [9–12]. The inclusions that are effective in nucleation site of acicular ferrite are Ti_2O_3, MnS+ V(C, N), and MnS+CuS [13–15]. Recently, it was demonstrated that [16,17], in the absence of sulfide inclusions, acicular ferrite is capable of nucleating on V(C, N) precipitates in V–N microalloying S-lean steel. To obtain relatively large-sized V(C, N) precipitates, holding and/or deformation at 900–950 °C were applied [17,18]. Moreover, the utilization of intragranular nucleation of ferrite on VN precipitates leads to a significant decrease in the grain size, by ~50%, while the polygonal ferrite forms instead of acicular ferrite

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during subsequent low continuous cooling rate of 3.5 °C/s. It was proposed that [19,20] the first growth of a layer of inert allotriomorphic ferrite at the austenite grain surface had the effect of suppressing the formation of bainitic sheaves and led to a transition from intergranular nucleation of bainite to intragranular nucleation of acicular ferrite. A theoretical analysis indicated that bainitic transformation is prevented at allotriomorphic ferrite/austenite boundaries by the carbon concentration field present in the austenite at the allotriomorphic ferrite/austenite interface. An approach for obtaining a high volume fraction of acicular ferrite was developed [21] by heat treatment of medium-carbon forging steel with two stage continuous cooling, resulting in an increase in the industrial production and application. However, the realization of acicular ferrite formation in hot rolled V–N microalloyed S-lean steel has not been reported to the best of our understanding. The acicular ferrite microstructure has the potential of providing high strength and excellent toughness [11].

It is also stated that compared to the hot strip mill, the interpass time of the reversing mills was significantly longer [22]. Moreover, the steel plates processed in reversing mills are medium thick or thick gauge, and the deformation reduction and cooling rate are inadequate [23,24]. The objective of the present study is to eliminate the disadvantage of reversing mills to a favorable situation such that acicular ferrite is nucleated, and the excellent comprehensive mechanical properties (strength, toughness, and ductility) in low carbon V–N microalloyed steel are obtained via controlled rolling and two stage continuous cooling, and the desirable microstructure obtained. The evolution of the V(C, N) precipitation was studied through thermodynamic analysis and combined with the TEM observation. The V–N steel processed using the above outlined approach has distinct advantages of high strength, excellent impact toughness, outstanding elongation, low cost, and is industrially applicable without the need for high degree of cooling and higher deformation reduction.

2. Experimental

2.1. Materials and thermo-mechanical processing

The experimental steel was melted in vacuum induction furnace and cast as 150 Kg ingot. The chemical composition of the steel in weight % was 0.058 C, 0.15 Si, 1.8 Mn, 0.002 S, 0.03 Al, 0.12–0.18 V, and 0.015–0.02 N, and balance Fe. The equivalent carbon content (C_{eq}) and the welding crack susceptibility index (P_{cm}) were 0.42 and 0.17, respectively, calculated using Eqs. (1) and (2) [25,26].

$$C_{eq} = C + \frac{Mn + Si}{6} + \frac{Ni + Cu}{15} + \frac{Cr + Mo + V}{5} \quad (1)$$

$$P_{cm} = C + \frac{Si}{30} + \frac{Mn + Cu + Cr}{20} + \frac{Ni}{60} + \frac{Mo}{15} + \frac{V}{10} + 5B \quad (2)$$

The 40 mm thick slab was heated to 1200 °C for 2 h to dissolve the microalloying elements, and then air-cooled to 950 °C. The slab was rolled to steel plate of 20 mm thickness, after three passes with interpass time of 30 s on Φ 450 mm trial rolling mill. The end temperature of finish rolling was controlled at 850 °C. After holding for 10 s, the plate was subjected to two stage continuous cooling. In the first stage, the plate was water-cooled at a rate of 16 °C/s to 540 °C, and subsequently the temperature was increased to 565 °C (self-tempering). In the second stage, the plate was air-cooled to room temperature at the cooling rate of ~0.5 °C/s. To observe the microstructural evolution and precipitation morphology in the deformed austenite, another steel plate was directly water-quenched to room temperature using an ultra-fast cooling (UFC) system after finishing rolling and holding. A schematic diagram of thermo-mechanical processing schedule is presented in Fig. 1.

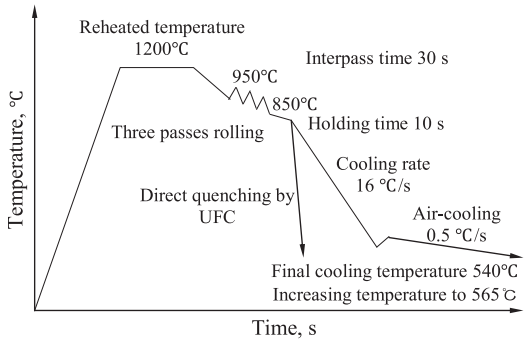

Fig. 1. Schematic diagram of thermo-mechanical processing.

2.2. Mechanical properties

Tensile samples of dimensions 10 mm diameter and 50 mm gauge length were machined from the plates parallel to the rolling direction. The tensile tests were conducted at room temperature using a cross-head speed of 3 mm/min using a Shimadzu AG-X universal testing machine. Charpy v-notch impact tests were performed at −20 °C using standard samples (dimensions: $10 \times 10 \times 55$ mm^3) with a v-notch parallel to the rolling direction using Instron Dynatup 9200 series instrumented drop weight impact tester, consistent with ASTM E23 specification [27]. The samples were cooled to −5 °C below the temperature of −20 °C to take into consideration the rise in temperature during transfer of sample to the Charpy v-notch impact tester.

2.3. Microstructure and thermodynamic analysis

The specimens for microstructural studies were polished using standard metallographic procedure and etched with a 4 vol% nital solution and observed using a Leica DMIRM optical microscope (OM) and Zeiss Ultra 55 scanning electron microscope (SEM). The chemical composition of the precipitates was determined by energy-dispersive X-ray spectroscopy (EDX). For electron backscattered diffraction (EBSD), the sample was electrolytically polished in a solution of perchloric acid and ethyl alcohol. The fracture surface of impact specimen was studied by a FEI Quanta 600 SEM. Transmission electron microscopy (TEM) studies were conducted using 3 mm diameter thin foils, electropolished using a solution of 8% perchloric acid and alcohol, and examined by FEI Tecnai G^2 F20 TEM at an accelerating voltage of 200 kV.

The theoretical calculations concerning evolution of various second phases with temperature, such as V(C, N), cementite, AlN, and MnS, and the composition of V(C, N) were studied using Thermocalc combined with TCFE6 database for thermodynamic calculation in equilibrium.

3. Results and discussion

3.1. Thermodynamic analysis of second phase

Fig. 2 summarizes the volume fraction of second phase precipitates as a function of temperature (V(C, N), cementite, AlN, and MnS) for the experimental steel and weight percent of element in V(C, N) precipitates calculated by Thermocalc. The start and end temperatures of ferrite to austenite transformation are 654 °C (A_{e1}) and 840 °C (A_{e3}), respectively. The blue line represents the volume fraction of V(C, N) precipitates as a function of temperature.