Ferrite channel effect on ductility and strain hardenability of ultra high strength dual phase steel

Ravi Kumar B.[a],[*], Nand Kumar Patel[b], Krishnendu Mukherjee[a], Mahesh Walunj[a], Gopi Kishor Mandal[a], Venugopalan T.[c]

[a] CSIR-National Metallurgical Laboratory, Jamshedpur 831007, India
[b] O.P Jindal University, Raigarh 496001, India
[c] Tata Steel Limited, Jamshedpur 831001, India

ARTICLE INFO

ABSTRACT

This study describes an effect of controlled austenite decomposition on microstructure evolution in dual phase steel. Steel sheets austenitized at various annealing temperatures were rapidly cooled to the inter-critical annealing temperature of 800 ºC for the isothermal decomposition of austenite and then ultra fast cooled to room temperature. The scanning electron microscope analysis of evolving microstructure revealed ferrite nucleation and growth along prior austenite grain boundaries leading to ferrite network/channel formation around martensite. The extent of ferrite channel formation showed a strong dependence on the degree of undercooling in the inter-critical annealing temperature regime. Uniaxial tensile deformation of processed steel sheets showed extensive local inter-lath martensite damage activity. Extension/propagation of these local micro cracks to neighboring martensite grains was found to be arrested by ferrite channels. This assisted in delaying the onset of global damage which could lead to necking and fracture. The results demonstrated an alternate possible way of inducing ductility and strain hardenability in ultra high strength dual phase steels.

1. Introduction

Intense research endeavours are ongoing to understand the effect of martensite and ferrite on mechanical properties of dual phase (DP) steels. Of the two, martensite being hard phase is understood to impart strength while ferrite being softer phase carries the most of the strain and thus provides ductility [1,2]. Further, it was realized that a trade-off exists between strength and ductility. At high strength levels ductility is compromised considerably and to circumvent it several efforts were made to vary the distribution and morphology of the two phases [3–5]. Volume fraction of martensite and grain size of ferrite is reported to affect the deformation mechanism and hence have strong influence on the strengthening behavior of DP steels [6–8]. In general, widely dispersed fine martensite particles provide a good combination of strength and ductility. Upto medium strength levels of 800 MPa, a reasonably good combination of strength and ductility is achievable for DP steels. Beyond this, in the high and ultra high strength regime, it is required to increase the volume fraction of the hard martensite phase. It causes a shift in the damage mechanism for these steels leading to large decrease in uniform elongation, upto a value of as low as 5%. Different strategies are adopted to introduce ductility and strain

hardening at ultra high strength levels of DP steels. Early research activities were mostly limited to process parameters related to inter-critical deformation or inter-critical annealing strategies for micro-structure control. However, recent trends are aimed at producing ultrafine-grained microstructure [9]. Of late, alloy additions (e.g. Mn, Si, Al, P, Nb, Cr, Mo or increasing carbon content) are being tried to improve the properties [10–13]. The present study is an attempt to understand the effect of controlled austenite decomposition at the inter-critical annealing temperature on the evolution of ferrite distribution and morphology. The role of ferrite morphology in imparting ductility in the ultra high strength regime, (i.e. above 1100 MPa DP steel) was looked into in case of conventional lean chemistry.

2. Experimental

2.1. Materials and processing

An industrially 67% cold rolled sheet of 0.83 mm thickness having a lean chemistry as shown in Table 1 was used in this study. Sheets of dimensions 100 mm×40 mm×0.83 mm were subject to rapid heating and stepped cooling annealing cycles in HDPS (Hot Dip Process

Table 1
Chemical composition of steel in wt%.

C	Mn	P	S	Si	Al	Cr	N
0.091	1.86	0.014	0.007	0.47	0.038	0.017	0.0035

Table 2
Peak annealing temperatures, inter-critical annealing temperature/austenite decomposition temperature and degree of undercooling below 800 °C for three different annealing processes.

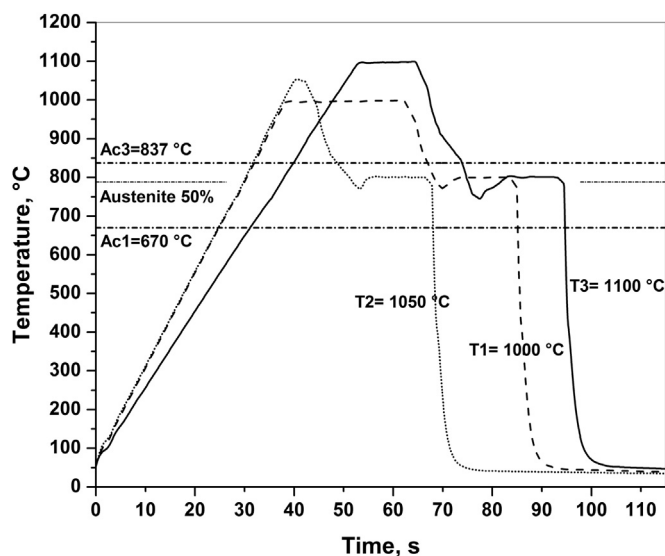| Annealing parameters	Peak Annealing temperature, °C	Undercooling below 800 °C, °C	Inter-critical Annealing temperature, °C
T1	1000	25	800
T2	1050	32	800
T3	1100	50	800

Fig. 1. Time-temperature profile of steel sheet annealed at three different PATs and superimposed with critical temperatures (Ac1 and Ac3) and 50% austenite formation temperature of 790 °C.

Simulator of ICS IWATANI SURTEC). The samples were heated to different peak annealing temperatures (PAT) of T1-1000 °C; T2-1050 °C; and T3-1100 °C, followed by a stepped cooling wherein they were held for short times at a selected inter-critical annealing temperature of 800 °C. Infra red (IR) heating technique was used for annealing process which was typically carried out under N_2−10 vol% H_2 gas atmosphere at a low dew point of −15 °C. At the first stage, a heating rate of 25 °C/s was used to reach the PAT. Subsequently the sheets were rapidly cooled to an inter-critical annealing temperature of 800 °C at the cooling rate of 30 °C/s by purging gas mixture. Different degrees of undercooling below 800 °C were introduced while rapidly cooling from PATs to 800 °C as per given in Table 2. A controlled isothermal decomposition of austenite was performed for 10 s at 800 °C. Finally the samples were ultra fast cooled (cooling rate ~150 °C/s) to room temperature by purging gas mixture. Annealing temperature-time data was acquired by a spot welded K type thermocouple at the surface of the steel sample. Hence, the temperature profile indicated in this study is the actual temperature experienced by the steel sheet. Time-temperature profiles for three steel samples T1, T2 and T3 are shown in Fig. 1. Range of inter-critical annealing temperature (between Ae1 & Ae3 of 670 °C & 837 °C, respectively) and the equilibrium temperature that will provide 50% austenite+50% ferrite (of 790 °C) for steel chemistry used in this study were obtained

using commercial software Thermo-Calc 3.0.1. As, in this study a microstructure consisting of 50% martensite and 50% ferrite is targeted, the inter-critical annealing temperature is chosen to be slightly (i.e., 10 °C) above the equilibrium 50% austenite+50% ferrite temperature. This slightly elevated temperature was used keeping in mind that austenite may not fully transformed to martensite under experimental conditions. In this study undercooling is defined as the cooling from the intercritical annealing temperature of 800 °C. These equilibrium temperatures determined by ThermoCalc were superimposed over time-temperature profile of annealing cycle in Fig. 1. Details of various annealing parameters are given in Table 2. The duration of the entire annealing process was very short, for about 2 min.

2.2. Microstructure characterisation

Small specimens were extracted from the annealed steel sheets for microstructure analysis. Standard metallographic techniques were used to reveal the microstructures. FEI quanta NOVA NANOSEM 430 field emission Scanning electron microscope (FEG-SEM) was used to record the microstructures. Image analysis technique was used for quantifying microstructure constituents. In this study the reported data are an average of measurements made on ten microstructure frames. Electron backscattering diffraction (EBSD) studies were performed by using an EDAX/TSL system equipped with a high-speed Digiview camera for the diffraction pattern acquisition. For EBSD studies specimens were prepared using the standard metallographic techniques, which were followed by final polishing with Silica gel. A step size of 0.1 μm was used for EBSD data acquisition.

2.3. Mechanical property evaluation

Tensile curves were determined as per the ASTM standard E-8M under displacement control at a strain rate of 1.3×10^{-4} s^{-1} using an Instron 8862 system of 100 kN capacity. Micro-hardness measurements were conducted in Vicker's scale using 5−50 gf load with a dwell time of 15 s to determine the hardness of individual microstructure constituents, i.e. of ferrite and martensite. In the case of smaller interacting volumes smaller loads were used in this study to determine the hardness of a single ferrite/martensite particle, without the interference from any other particle.

3. Results and discussion

3.1. Microstructure evolution

From the SEM quantitative microstructure analysis it was found that 53.6%, 54.7% and 56.3% of martensite formed at PAT's of 1000, 1050 and 1100 °C, respectively. The variation was noted to be small owing to identical inter-critical isothermal annealing or austenite decomposition temperature of 800 °C. These results agree well with the ThermoCalc predicted volume fraction of austenite (58%) at 800 °C. The slight lower fraction of martensite obtained in the experiments than the equilibrium austenite fraction, may be due to the fact that the short isothermal holding time did not allowed equilibrium partitioning of carbon from ferrite to austenite. Further, a very small increase in martensite fraction was observed in the case of T3 wherein larger undercooling below 800 °C was introduced (Table 2). Representative microstructures at various PATs depicting various ferrite morphologies are shown in Fig. 2. Interestingly, a correlation in the nature of ferrite distribution and its morphological distribution around martensite were noticed with increasing PAT. At T1, a random spatial ferrite and martensite distribution was seen. With increasing PAT to T2 and T3, the randomness of ferrite distribution was observed to disappear as shown in Fig. 2(b). At T2 and T3, martensite appeared to be surrounded by ferrite, giving an impression of core-shell type structure. This GB allotriomorphic ferrite network formation was more