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A study of the softening mechanisms of laser-welded DP1000 steel butt joints

Jinfeng Wang [a,b], Lijun Yang [a,c,*], Mingsheng Sun [a], Tong Liu [a], Huan Li [a]

[a] School of Materials Science and Engineering, Tianjin University, Tianjin 300072, PR China
[b] School of Materials Science and Engineering, Hubei University of Automotive Technology, Shiyan, Hubei 442002, PR China
[c] State Key Laboratory of Advanced Welding and Joining, Harbin Institute of Technology, Haerbin, Heilongjiang 150001, PR China

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ABSTRACT

Joint softening occurs in the laser-welded joints of many types of dual-phase steels that are used in automobile structures and weakens the joints' mechanical properties. This paper presents a study of the softening mechanisms in laser-welded joints of DP1000 steel. A Gleeble-3500 thermal simulator was used to simulate the welding thermal cycles of all of the sub-zones of the heat-affected zone (HAZ) of laser-welded specimens of DP1000 steel. The hardness and microstructure of the specimens were then tested and analysed. The results indicate that the softening resulted from two changes in the microstructure. The first change was the transformation of the pre-existing harder martensite into softer tempered martensite and the precipitation of carbides when the DP1000 steel was heated to the tempering temperature during the welding process. The other change was the decrease in the percentage of martensite, while the percentage of softer phases increased because the DP1000 steel was heated to the inter-critical temperature. At this temperature, the ferrite transformed from the martensite and some of the ferrite in the original microstructure were transformed into austenite. During the subsequent cooling process, the austenite was transformed into polygonal ferrite, bainite and martensite-austenite phases, therefore, the percentage of softer phases increased.

1. Introduction

DP1000 steel consists of a soft ferritic matrix that contains islands of hard martensite. The soft ferrite phase is generally continuous, which gives this steel excellent ductility. When DP1000 steel deforms, the strain is concentrated in the lower-strength ferrite phase that surrounds the islands of martensite, which produces the unique high work-hardening rate of this steel. The work-hardening rate and the excellent elongation give dual-phase (DP) steel a much higher ultimate tensile strength (TS), a higher initial work-hardening rate, and a lower yielding-to-tensile strength ratio (YS/TS) than high strength low alloy (HSLA) steels with similar yield strength (YS). As a result, dual-phase steels are widely used in the automotive industry to improve crash performance and reduce weight [1].

A laser beam with a high and concentrated energy density can produce a weld seam with a large depth-to-width ratio, reduce the width of the heat-affected zone (HAZ) of welded joints, cause minimal post-welding deformation in the workpieces, and avoid deteriorating the properties of the welded joints; consequently, laser welding is widely applied in the automobile manufacturing industry [2–3]. However, the inhomogeneity of the microstructure and properties of welded joints of high-strength steel has hindered its application in auto manufacturing. The softening of the HAZ of a welded joint is the most significant problem [4–7].

Many researchers have studied the softening of laser-welded dual-phase steels. Several studies have examined welded dual-phase steels with strength grades that range from 450 to 980 MPa, the results showed that HAZ softening occurs in all of these steels and that the degree of softening and the size of the soft zone increase with increasing strength grade of the steel [8–10]. Additional studies have shown that the degree of HAZ softening in a dual-phase steel is also closely related to the chemical components of the steel and that the resistance to softening increases with increasing alloy content [7,11]. In addition, the heat that is required for HAZ softening in dual-phase steel decrease with increasing martensite carbon content in the steel. The existence of carbide-forming elements, such as Cr and Mo, in a dual-phase steel can increase the resistance to softening. Furthermore, the degree of softening in the welding HAZ increases with increasing heat input [12].

The consensus is that the higher the strength grade of high-strength steel is, the more severe the softening phenomenon is. Softening of the HAZ decreases the strength, plasticity, formability, and fatigue performance of the welded joint [13–16], which in turn affect its service performance. Due to the global environmental and energy crisis, the demand for lightweight automobiles has increased. One of the main approaches to producing lightweight automobiles is to use advanced high-strength steels with higher strength grades and better formability in

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[*] Corresponding author at: 92 Weijin Road, Nankai District, Tianjin, PR China.
E-mail address: yljabc@tju.edu.cn (L. Yang).

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large quantities. DP1000 steel meets these requirements. Researchers have conducted many studies on the softening phenomenon of dual-phase steels but relatively few on the softening mechanisms of DP1000 steel. Our research group has discovered that detectable softening occurs in the HAZ of a laser-welded joint during the laser welding process of DP1000 steel [17]. In addition, different heat inputs have different effects on the degree of softening of the welded joint. To elucidate the softening mechanism of laser-welded DP1000 steel, this study thermally simulated the microstructures in different sub-zones of the HAZ with a Gleeble-3500 thermal simulator using curves of welding thermal cycles that were obtained from a combination of experimental and numerical simulations using ANSYS software. This study analysed the characteristics of the microstructures and the hardness distributions in the sub-zones of the HAZ in welded joints. Additionally, this study characterised the softening mechanisms of the HAZ of laser-welded DP1000 steel joints using an optical microscope (OM), a scanning electron microscope (SEM), a transmission electron microscope (TEM), and a hardness metre.

2. Experimental materials and methods

2.1. Experimental materials

A 1.5-mm-thick continuous cold-rolled DP1000 steel sheet was used as the experimental material. Table 1 lists the main chemical components (wt.%) and the mechanical properties in the parallel rolling directions.

2.2. Experimental method

The welding thermal simulation specimens were cut from the 1.5-mm-thick steel sheet in the rolling direction. The steel sheet was relatively thin, and the heating and cooling rates of the laser welding were relatively fast. Therefore, to ensure that they would be subject to suitable heating and cooling rates, the specimens were machined into a dumbbell shape. Fig. 1 shows the shape and dimensions of these specimens. The welding thermal simulation experiment was conducted with a Gleeble-3500 thermal simulator.

The softening phenomenon in the HAZ during the actual welding process was investigated. Different values of the peak temperature (T_{max}) were set based on the Ac_1 temperature (686 °C) and the Ac_3 temperature (893 °C) of DP1000 steel, which were determined using a combined differential scanning calorimetry-thermogravimetric analysis system and the sub-zones of the welding HAZ that were identified in previous reports [12,18]. At room temperature, DP1000 steel consists of ferrite and martensite, the volume fraction of martensite is approximately 57%. The martensite in the original microstructure of DP1000 steel will change during the laser welding process and cause softening of the HAZ. Therefore, welding thermal cycles were selected for the thermal simulation at temperatures from 150 °C to 1300 °C. Table 2 lists the sub-zones of the welding HAZ and the T_{max} values, and Fig. 2 shows the curves of the welding thermal cycles.

Different T_{max} values were selected for the numerical thermal simulations using ANSYS software to study the characteristics of the microstructures and the hardness distributions in different sub-zones of the HAZ (coarse-grain HAZ, fine-grain HAZ, inter-critical HAZ, and sub-critical HAZ). The central parts of the thermal simulation specimens were cut off and used as the metallographic specimens (which doubled as the hardness specimens) and the TEM specimens. The metallographic specimens were corroded in a 4% (volume fraction) nitric acid solution in alcohol for 4–6 s. The microstructures of the metallographic specimens under different welding thermal cycle conditions were observed using a Neophot32 OM and a JSM6360LV SEM. The fine microstructures of the precipitated phases were observed using a JEM-2100 TEM. The hardness tests were conducted using an MHV-2000 Vickers hardness metre with a digital display (load: 200 g, dwell time: 15 s, spacing between subsequent indentations: 0.5 mm) in the cross sections of the etched specimens. Twenty-five points on each specimen were tested using the array that is shown in Fig. 3. The average grain size was measured with the OM by taking at least 20 measurements according to GB/T6394-2002 (metal-methods for estimating the average grain size).

3. Experimental results and discussion

3.1. Hardness distributions of sub-zones

Fig. 4 shows the average hardness values of the base metal and the thermally specimens that were obtained under different T_{max} values, and Table 3 shows the relationship between T_{max} and the corresponding hardness range of the specimens. The average hardness of the base metal is 319 HV, and the hardness ranges from 312 HV to 330 HV. When T_{max} is between 150 °C and 250 °C, the hardness distribution is not significantly different from the hardness of the base metal, which has an average hardness of 317 HV and 318 HV and a range of hardness from 301 HV to 321 HV. The hardness of the specimen ranges from 312 HV to 330 HV at 150 °C and from 292 HV to 320 HV at 250 °C. Therefore, the hardness of DP1000 steel changes only slightly when T_{max} is below 250 °C, which may be slightly different from the conventional low-temperature tempering of martensite (theoretically, low-temperature tempering of martensite occurs upon heating to 80 °C).

Table 1
Main chemical components and mechanical properties of DP1000 steel.

Main chemical components (wt.%)									Mechanical properties			
C	Si	Mn	Cu	Al	Cr	Ni	Nb	V	Tensile strength	Yield strength	Yield-to-tensile strength ratio	Elongation A%
0.141	0.49	1.47	0.02	0.041	0.03	0.04	0.016	0.01	1034 MPa	680 MPa	0.67	11

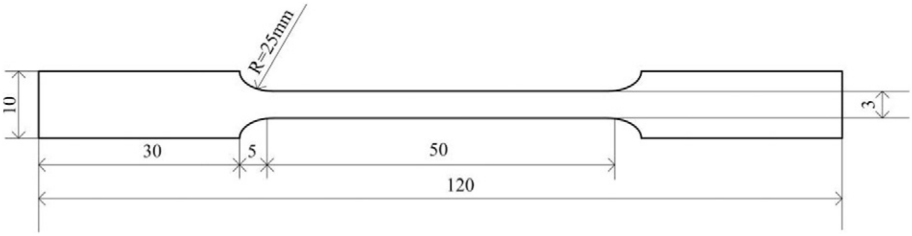

Fig. 1. Schematic diagram of the thermal simulation specimen.