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Influence of SC-HAZ microstructure on the mechanical behavior of Si-TRIP steel welds

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ABSTRACT

Transformation induced plasticity (TRIP) steel provides enormous potential for auto-body construction in the automotive sector, owing to its enhanced mechanical behavior. In this work, Si-alloyed TRIP steel is joined by employing laser beam welding (LBW) and by utilizing two arc welding processes: gas tungsten arc welding (GTAW) and gas metal arc welding (GMAW) in order to assess the effect of the net heat input on the microstructure, the uniaxial tensile properties and the forming response. Results indicate that in spite of the Si content in TRIP steel; precipitation and growth of carbides (tempering) are observed in both: the martensite islands and the retained austenite phase, thus leading to the measurable softening at the sub-critical heat affected zone (SC-HAZ) of the arc welded samples. Although the failure location was predominantly found at the sub-critical heat affected zone of the GMAW samples, the maximum stress and elongation was basically controlled by the total extension of the weldment including fusion zone and heat affected zone. While the limiting dome height upon tension-tension (T-T) and tension-compression (T-C) depended primarily on the fusion zone hardness, weld width and geometry of the sample; the fracture location was outside the weld for T-C, whereas the fracture initiated at the weld in T-T samples. LBW specimens showed optimum forming performance.

1. Introduction

Transformation induced plasticity (TRIP) steel belongs to the family of advanced high strength steels (AHSSs) and is composed of a ferrite (α) matrix along with a fraction of retained austenite (γ or RA), dispersed islands of martensite (α´) and some additions of bainite (B) [1]. The relatively high content of Si (i.e. Si-TRIP alloying) and C favors the presence of retained austenite, which in turn promotes an increased hardening rate due to martensite transformation at elevated strains. Such an exceptional structure-property connection is taken into consideration for body-in-white designs where significant stretch forming is required. Furthermore, Si-TRIP steel allows auto makers to meet government regulations concerning both fuel efficiency and passenger safety requirements; thus, making it suitable particularly for crash-worthy behavior in automotive structural components as for example: B-pillar, engine cradle, front and rear rails, among others [1].

In industrial practice, TRIP steel is mainly joined by resistance spot welding (RSW) in the course of assembly [1–8], by laser beam welding (LBW) mostly used in production of tailor-welded blanks [9–12], and by gas metal arc welding (GMAW) in which strength and rigidity is required [13–17]. High fusion zone (FZ) hardness is commonly achieved because of the fast cooling rates encountered by the water-cooling electrodes if employing RSW; however, the joint is prone to fail at the nugget or FZ, in the form of interfacial or partial-interfacial failure mode upon lap-shear tensile testing [2]. The above mentioned failure modes in resistance spot welding of TRIP steel ca n be overcome by employing post-weld heat treatment via a secondary electrode pulse or in-process tempering [3]. On the other hand, penetration, porosity, microstructure, and hardness can be controlled when using LBW in order for the weldment to obtain acceptable mechanical response [9]. For example, a reduction in fusion zone hardness has been obtained when employing twin-spot laser welding [18]. The extent of improvement (once the controlling parameters are set up) has been observed for tailor-welded blanks of dissimilar combinations of advanced high strength steel or mild steel with TRIP steel; for example, excellent fatigue life has been obtained when pairing TRIP steel to mild steel by LBW [10]. Furthermore, the possibility of further controlling the FZ hardness in order to minimize brittleness without losing strength can be

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efficiently accomplished by utilizing GMAW; which is attributed to a wide range of filler metals and welding parameter variations available upon this welding process. For instance, the metal inert gas (MIG)-Brazing method with filler metal $ERCuAl_8$ was successfully employed for joining TRIP800 steel thus resulting in comparable strength, lower heat input, reduced zinc evaporation and decreased spatter of the filler metal [13]. Moreover, FZ failures with diminished ultimate tensile strength (i.e. 73% of the weakest alloy) have encountered when attempting to join transformation induced plasticity steel to Twinning-Induced Plasticity steel with an ER307Si filler metal by GMAW process [14].

However, TRIP steel has been hardly studied in the way of employing various welding processes particularly for rating the effect of welding heat input on the resultant fusion zone and/or heat affected zone (HAZ) microstructure, neither to systematically correlate microstructure to its mechanical behavior. For instance, high Si-content TRIP800 steel was laser welded and arc welded by means of metal active gas and tungsten inert gas processes, in the last two employing alloyed filler metal [19]. Although the tensile results indicated that the failure in laser welded sample was located at the base metal (BM) with enhanced ultimate tensile strength (UTS) and an acceptable elongation, the failure switched to the HAZ-BM interface for the arc welded specimens with diminished UTS and clearly reduced ductility [19]. However, in the later report no major details were provided in terms of microstructure development in both fusion zone and heat affected zone. In particular, the effect of softening (tempering) at the SC-HAZ on the tensile properties and the forming characteristics of laser and arc welded TRIP steel.

It is worthy to mention here that another member of the family of AHSS called dual phase (DP) steel is prone to softening at the sub-critical heat affected zone due to tempering of the martensite islands, which has been studied extensively [20–25]. Contrary to this view, scarce communications have been released regarding to softening at the sub-critical heat affected zone of TRIP steels in terms of laser or arc welding processes [26,27]. Zhao et. al. [26] indicated that the reason for softening was the depletion of retained austenite associated with a difference in cooling rates, and to the existence of a large fraction of ferrite but no major discussion was provided. Xia et al. [27] also reported HAZ-softening in diode laser welded TRIP800 steel but no further consideration was given.

This work aims on systematically analyze the influence of fusion zone and sub-critical heat affected zone microstructure on the mechanical response (tensile) and the forming behavior of Si-TRIP steel subjected to different net heat input such as those that involve laser and arc welding.

2. Experimental procedure

2.1. Material

Cold rolled TRIP steel sheet with nominal chemical composition (listed in Table 1), 800 MPa of UTS and thickness of 1.0 mm was used in this study. It is worthy to mention that this type of TRIP steel is also called "Si-TRIP" or "Si-alloyed TRIP" as per the relatively high content of Si [27].

The carbon equivalent as calculated by Yurioka´s equation [28] is $CE_Y = 0.527$. Whereas the corresponding calculated transition temperatures called critical temperatures: Ac_1 and Ac_3 are about 747 °C and 915 °C, respectively. Additionally, the martensite start temperature (M_s) is 380 °C as reported elsewhere [3]. The initial microstructure or base metal microstructure of Si-TRIP steel illustrated in Fig. 1a is composed of a ferrite (α) matrix along with dispersed islands of martensite (α´), retained austenite (γ or RA) with volume fractions of 12–14% and 10–12%, respectively, as per results obtained from metallographic techniques and X–ray diffraction (XRD) analysis (Fig. 1b). In addition, the presence of bainite (B) is also observed in Fig. 1a.

Table 1
Chemical composition of the base metal (wt%).

C	Mn	Si	Cu	Ni	Al	Cr	Mo	P
.18	1.63	1.61	0.02	0.016	0.03	0.02	0.01	0.01

2.2. Welding procedure and heat input

Monolithic Si-TRIP steel sheets were butt-welded by employing three different welding processes: fiber laser welding (LBW), gas tungsten arc welding (GTAW) and gas metal arc welding (GMAW). It has been previously established that heat input increases (from low to high) for the above mentioned welding processes as follows: LBW→GTAW→GMAW [29,30]. Whilst the laser welding process promotes lower heat input; the two arc welding processes represent higher heat input conditions.

LBW samples were made by an *IPG Photonics Ytterbium Doped Fiber YLS-6000* machine operated at 4 kW with a welding speed of 12 m/min, a beam spot size of 600 µm and a focal length of 200 mm. No shielding gas was used during the welding process. GTAW sample was carried out by means of a semiautomatic *Miller Syncrowave™ 250* power source operated at 17.6 kVA / 8.6 kW, employing a pure tungsten type electrode of 1.6 mm in diameter, travel and work angle of 70° and 90°, respectively, and by using direct current electrode negative. GMAW blanks were obtained in an *Infra™ MM 300-E* machine operated at 10.1 kVA / 7.7 kW with non-pulsed short circuit metal transfer mode, using an ER70S-6 solid welding wire of 0.9 mm in diameter, with a travel and work angle of 90°, and by using direct current electrode positive. All welds were made transverse to the rolling direction with zero gap, a series of trials were carried out in order to obtain appropriate welding parameters, all samples were free of any induced visible defects (i.e. lack of fusion, burn-through, porosity, spatter, etc.). Before the welding operation all sheets were fixed in copper-alloy jigs to achieve a heat sink condition in order to minimize further distortion. The optimized parameters for both arc welding processes are provided in Table 2.

Calculations made for the heat input were obtained according to the methodology suggested by Xia et al. [21] through the following equation:

$$\frac{Q_{net}}{vd} = \frac{\rho c (r_{Ac_1} - r_m)(2\pi e)^{1/2}}{\left(\frac{1}{T_{Ac_1} - T_0} - \frac{1}{T_m - T_0}\right)} \tag{1}$$

The net heat input (Q_{net}/vd) as indicated in Eq. (1) representing the thickness normalized net absorbed energy per unit weld length. In this case, ρ is the density of steel, c is the specific heat capacity of steel, r_{Ac1} and r_m are the isotherms positions corresponding to the T_{Ac1} (Ac_1 temperature) and T_m (melting temperature), T_0 is the initial room temperature. The corresponding values of Q_{net}/vd are listed in Table 3.

A physical meaning for the heat input can be properly approached by employing a term called time constant (τ), representing the time to heat a point at the SC-HAZ basically at the lower line of critical temperature Ac_1; hence, high heat input processes require more time to reach the temperature [22].

By assuming that the temperature history during welding is of parabolic shape [25]; and utilizing the method of Xia et al. [21],τ is calculated by means of:

$$\tau = \frac{\left(\frac{Q_{net}}{vd}\right)^2}{4\pi e \rho c_p \lambda (T_{Ac_1} - T_0)^2} \tag{2}$$

where λ is the thermal conductivity.