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Nitrogen loss and effects on microstructure in multipass TIG welding of a super duplex stainless steel

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ABSTRACT

Nitrogen loss is an important phenomenon in welding of super duplex stainless steels. In this study, a super duplex stainless steel was autogenously TIG-welded with one to four bead-on-plate passes with low or high heat inputs using pure argon shielding gas. The goal was to monitor nitrogen content and microstructure for each weld pass. Nitrogen content, measured by wavelength dispersive X-ray spectrometry, was after four passes reduced from 0.28 wt% in the base metal to 0.17 wt% and 0.10 wt% in low and high heat input samples, respectively. Nitrogen loss resulted in a more ferritic structure with larger grains and nitride precipitates. The ferrite grain width markedly increased with increasing number of passes and heat input. Ferrite content increased from 55% in base metal to 75% at low and 79% at high heat inputs after four passes. An increasing amount of nitrides were seen with increasing number of weld passes. An equation was suggested for calculation of the final nitrogen content of the weld metal as functions of initial nitrogen content and arc energy. Acceptable ferrite contents were seen for one or two passes. The recommendation is to use nitrogen in shielding gas and proper filler metals.

1. Introduction

Duplex stainless steel (DSS) attracts a lot of attention and are increasingly used in different industries like petrochemical, oil and gas [1,2]. They, as economical alloys, have become a suitable alternative to austenitic stainless steels due to their special combination of good mechanical properties and corrosion resistance. DSS has a microstructure consisting of austenite and ferrite, typically around 50% of each phase in the base material but typically a variation of 35 to 60% in ferrite content results in acceptable properties [3–5].

Duplex stainless steels generally demonstrate good weldability, but the formation of excessive ferrite contents, nitrides, and intermetallic phases in the weld metal and heat affected zone (HAZ) can occur [6]. Recommendations [7], accordingly, have been established for welding DSS, for example to use limited heat inputs, filler metals with promoted nickel content, nitrogen containing shielding and backing gases in TIG welding. The above mentioned measures aim at controlling the ferrite content in the recommended range as well as avoiding the formation of nitrides and intermetallic phases [8–10].

Nitrogen, as one of the strongest austenite forming alloying elements, is an economical and efficient substitute for nickel [11,12]. It also contributes to high strength and corrosion resistance. If an excessively high content of ferrite forms in the weld metal due to loss of nitrogen during welding it will impair mechanical and corrosion performance of DSS and in particular high nitrogen super duplex stainless steel (SDSS) grades [13]. Du Toit et al. [14,15] monitored and modelled the absorption and desorption of nitrogen within the molten weld metal of stainless steels. They performed welding on various base metals with shielding gases with different nitrogen contents. It was concluded that the initial contents of nitrogen and surface active elements in the base metal were key factors in controlling the nitrogen loss. Hertzman et al. [16] studied nitrogen pick up of 2507-type SDSS and determined the effect of arc length and nitrogen content of the shielding gas. It was revealed that the nitrogen content of shielding gas and the volume of the weld pool affected the final nitrogen content. Karlsson et al. [13] performed low heat input autogenously laser welding on 2507 SDSS, which resulted in highly ferritic microstructures with nitride precipitates due to nitrogen loss and high cooling rates.

Cooling rate, in addition to the nitrogen content, dictates the balance between austenite and ferrite [17]. Higher cooling rates result in a higher ferrite content because austenite formation is restricted due to insufficient time for diffusion [18]. The formation of nitrides is also facilitated, since less nitrogen atoms can diffuse into the austenite. Supersaturation of nitrogen and precipitation of nitrides, as a consequence, occurs in ferrite with decreasing temperature [19]. Excessive heating, on the other hand, resulting in lower cooling rate can cause the formation of intermetallic compounds like sigma and chi phases [20].

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Table 1
Chemical composition (wt%) of super duplex plate material according to material certificate.

C	Si	Mn	P	S	Cr	Ni	Mo	N	Cu
0.02	0.29	0.79	0.023	0.001	24.79	6.89	3.79	0.277	0.15

Computational thermodynamics have been identified as an important tool to predict the microstructure and properties of welds in DSS and several investigations have confirmed thermodynamical calculations to predict characteristics like the reformation of austenite [21], microstructure coarsening [22] and formation of secondary phases [23] well. In addition, Tan et al. [24] pointed out that pitting resistance equivalents (PREN) calculated using Thermo-Calc software was in good agreement with measured critical pitting temperatures (CPT) of a DSS annealed at different temperatures.

A considerable number of researchers have addressed the effect of nitrogen loss on microstructure and properties of duplex weld metals [16,25–27]. However, the accumulative effects of multipass welding on nitrogen loss for different thermal cycles remain largely unknown. The objective of the present approach, hence, is to understand the mechanisms and effects of nitrogen loss in autogenous multipass TIG welding of a 2507-type SDSS at different heat inputs using a combined experimental and theoretical approach. A method, using robotic welding combined with thermal cycle analysis, was developed to produce multipass welds with well controlled thermal cycles. Nitrogen loss and microstructural features like ferrite content and grain size were evaluated and a model to predict nitrogen content was suggested.

2. Experimental

2.1. Material

Type 2507 SDSS sheets with 6 mm thickness (EN 1.4410 or UNS S32750) were used as base material. Actual chemical composition of the alloy is given in Table 1. A schematic drawing of a test plate with the weld at the center is presented in Fig. 1.

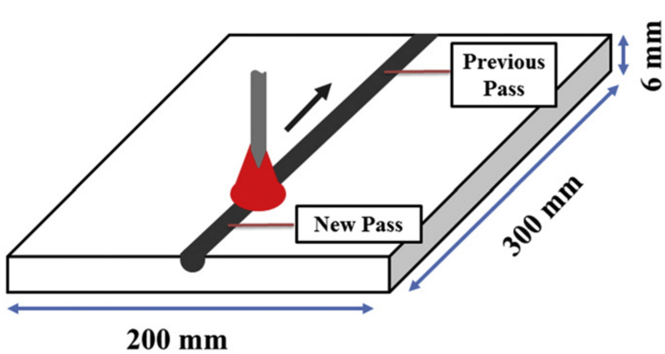

Fig. 1. A schematic drawing of the plate and autogeneous TIG welding. The same material volume was melted, and remelted, with one, two, three or four bead-on-plate weld passes.

Table 2
The designation and description of weld samples.

Designation	Description	Designation	Description
L1	LHI* - 1 weld pass	H1	HHI** - 1 weld pass
L2	LHI - 2 weld passes	H2	HHI - 2 weld passes
L3	LHI - 3 weld passes	H3	HHI - 3 weld passes
L4	LHI - 4 weld passes	H4	HHI - 4 weld passes

*Low heat input; **High heat input.

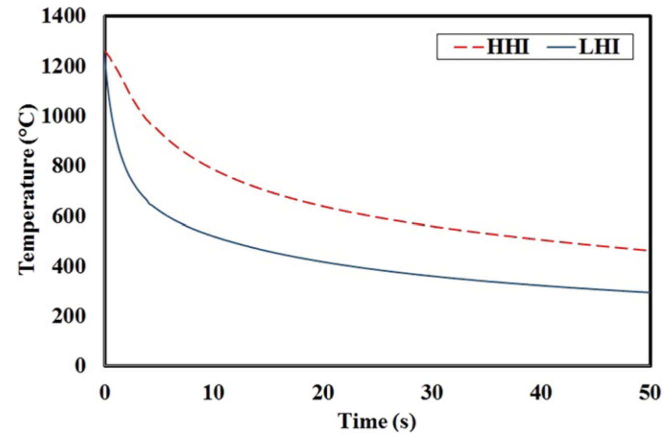

Fig. 2. Cooling curve of weld metal for low heat input and high heat input weld samples showing the more rapid cooling for the lower heat input (LHI) and the smaller difference in cooling rate with decreasing temperature.

2.2. Welding and arc efficiency determination

Robotic welding was conducted to guarantee a stable and repeatable welding procedure. A TIG COMMANDER AC/DC 400 power source was employed. Two different arc energies (0.47 kJ/mm and 1.08 kJ/mm) were chosen for autogenous TIG welding using pure argon as shielding gas. The average currents, voltages and welding speeds of these two different arc energies were 145.1 A, 13.0 V, and 4 mm/s for the low arc energy and 145.5 A, 14.7 V and 2 mm/s for the high arc energy, respectively. To increase the nitrogen loss of the weld step by step, at each arc energy, the same material volume was melted, and remelted, with one, two, three or four bead-on-plate weld passes (Fig. 1). Thereby, eight samples, as detailed in Table 2, were prepared to measure the nitrogen loss and to characterize microstructure at each arc energy for the different numbers of weld passes.

The heat inputs for the low and high arc energies were determined to be 0.37 kJ/mm, for the low heat input (LHI) and 0.87 kJ/mm, for the high heat input (HHI), using a calorimetric arc efficiency measurement method. The details of the method and equipment are presented in Ref. [28].

2.3. Thermal cycle measurements

A K-type thermocouple was introduced into the liquid weld pool at the mid length of the 300 mm long plate immediately after the arc passed the location while the weld pool was still liquid. It was impossible to attach the thermocouple before welding as it would be destroyed by the arc. The thermocouples were connected to a 16-Channel Thermocouple Input Module to record the thermal cycle and the temperature was recorded every 0.1 s.

2.4. Evaluation of microstructure and chemical composition

Cross sections of the samples were ground and polished using standard procedures. Macroetching was done by electrolytic etching using 20% NaOH with an applied voltage of 4 V for around 6 s. An Olympus SZX stereo microscope was used to record macrographs. Microetching was done using a 10% NaOH electrolytic etchant with an applied voltage

Table 3
Cooling time for LHI and HHI welded samples for two different temperature ranges.

Thermocouple	$\Delta t_{1200/800}$ (s)	$\Delta t_{800/500}$ (s)
LHI	1.9	10.9
HHI	7.5	24.3