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Influence of multipass pulsed gas metal arc welding on corrosion behaviour of a duplex stainless steel

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ABSTRACT

This work focuses the effect of secondary austenite (γ_2) on corrosion resistance of welded joints. The welded joints were obtained by pulsed gas metal arc welding (GMAW-P) with three different heat inputs. Each joint and their respective sub-regions through thickness were characterised by optical microscopy and scanning electron microscopy. The results were correlated with double loop electrochemical potentiodynamic reactivation (DL-EPR) and sulphide stress corrosion tests. The results suggest a good agreement between the DL-EPR and the four point bending tests. It was also verified that the simple presence of γ_2 does not necessarily gives rise to loss of corrosion resistance.

1. Introduction

Duplex stainless steels are metallic alloys based on the Fe–Cr–Ni system, in which chemistry and thermo-mechanical processing confers a dual phase microstructure with similar amounts of ferrite (δ) and austenite (γ), typically in a proportion 50–50%. These types of alloys are low carbon (less than 0.04%) and contain 20–30% Cr, 0–5%Mo, 1–8%Ni and 0.1–0.3%N, as well as some additions of Cu and W. Such highly alloyed grade steels show very good pitting and sulphide stress corrosion (SSC) resistance in low pH and sour environments even in presence of chlorides. The resistance to pitting corrosion of the base material can be optimised by selecting a proper annealing temperature. In this condition, alloying elements concentrations give rise to similar pitting resistance equivalent number (PRE_N, empirical parameter that defines the localised corrosion resistance of alloys, in general, PRE_N = wt.%Cr + 3.3 wt.%Mo + 20 wt.%N) for both δ and γ phases [1]. However, during welding a specific heat input is created and kinetics aspects play an important role on the local properties, which may result in changes in chemical composition, weight percentage and morphology of δ and γ phases [2,3]. Therefore, during heat treatment or welding, attention should be drawn to ensure that the PRE_N in each phase is maintained. Actually, a redistribution of alloying elements can occur as a result of formation of nitrides or by excessive precipitation of γ_2. All kinetic factors that change the chemical composition of phases may lead to local decrease of pitting resistance and, consequently, local corrosion attack can occur [4].

From a totally δ matrix, γ tends to precipitate first on the δ/δ boundary and then grow to the ferrite grain with widmanstätten morphology. This phase is called primary austenite, since it is formed just after the complete ferritic solidification ($L \rightarrow \delta + \gamma_1$). However, in case of rapid cooling, this austenite might be partially replaced by a supersaturated ferrite, enriched with nickel, manganese and mainly nitrogen. Under this supersaturated condition, after subsequent thermal process, such as multipass welding, the decomposition of ferrite phase can occur, resulting in the formation of a new austenite phase ($\delta \rightarrow \delta + \gamma_2$). This γ_2 arises at temperatures where a dual phase microstructure ($\delta + \gamma_1$) is already established [5–9]. The γ_2 phase can be also formed by eutectoid decomposition of ferrite in sigma and austenite ($\delta \rightarrow \sigma + \gamma_2$). Another mechanism giving rise to γ_2 is associated with Cr_2N. In fact, when this precipitate is formed, generally in the δ/δ boundary, the neighbouring region is chromium depleted and thus γ_2 might appear. If the heat cycle gives rise to Cr_2N precipitation inside the δ grain, during its dissolution δ grain becomes supersaturated in nitrogen and then γ_2 might be produced [4,10–14]. In general, the rapid cooling affects directly the diffusivity and solubility of alloying elements in the phases during δ and γ transformations. This may lead to a considerable imbalance between phases, changes in the morphologies and detrimental effects on individual

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Table 1
Chemical composition and pitting resistance equivalent number (PRE_N) of the material.

Material UNS	wt.%										PRE_N
	C	Cr	Ni	Mo	N	Cu	Mn	S	W	Fe	
S32750	0.031	25.38	6.72	3.80	0.21	0.136	1.176	0.0005	0.020	Bal	≈41.4

Table 2
Specification of etchant used, objective and procedure by etch used.

Etching	Objective	Procedure	References
NaOH (20%)	General microstructure observation	Electrolytic: 3 Volts by 10–20 s [20,21]	[4,7,10,14,24,25]
KOH (40%)	Highlights the σ phase	Electrolytic: 3 Volts by 5–15 s [20,21]	[4,7,10,14,24,25]
Oxalic Acid (10%) + NaOH (40%)	Highlights the presence of Cr_2N or CrN	Electrolytic in two stages: 1°–3 Volts by 5–15 s (Highlights Cr_2N or CrN) and 2°–3 Volts by 10–15 s (Highlights the localisation of Cr_2N or CrN)	[4,12]
HNO_3 (40%)	Highlighting the different types of γ in the weld metal	Electrolytic in two stages. 1°: 1–1.2 Volts by 120 s (highlighting the grain boundaries) and 2°: 0.7 Volts by 420 s (differentiate the types of γ)	[4]

PRE_N [15,16]. As examples, Pak and Karlsson [17] showed that multipass welding may produce intragranular γ_2 in 23Cr–3Mo–9Ni weld metal with an adverse effect on pitting corrosion resistance. The authors suggested that this is caused by low concentration of nitrogen in this phase. A chromium depleted γ_2 phase was also observed after an isothermal heat treatment in super-duplex stainless steel 25Cr–4Mo–7Ni base metal [18,19]. The formation of such phase occurred by nucleation at ferrite/austenite phase boundaries and grows into the ferrite in the temperature range 700–900 °C.

The present work focuses on the effect of γ_2 on the corrosion resistance of welded joints. The welded joints were obtained by GMAW-P with three different heat inputs. Each joint and their respective sub-regions were characterised by optical microscopy, scanning electron microscopy (SEM) and correlated with electrochemical tests.

2. Experimental

The material used was a super-duplex stainless steel UNS32750 pipe with outside diameter of 152.40 mm and wall thickness of 19.05 mm. The chemical composition (%wt) is presented in Table 1.

All welded joints were prepared using a half V configuration with filler metal of 1.2 mm diameter (specification 25. 9.4 N L) and Ar (70%) + He (30%) as shielding gas. The welding process applied was GMAW-P with maximum interpass of 150 °C and three different heat inputs: 0.8 kJ/mm, 1.26 kJ/mm and 2.0 kJ/mm. The numbers of welding passes to fulfil the joint in each heat input were 20, 16 and 12 for 0.8 kJ/mm, 1.26 kJ/mm and 2.0 kJ/mm, respectively.

3. Microscopy

Optical microscopy was used for identification of intermetallic phases and/or microconstituents as well as quantification of the δ/γ balance. Four different chemical etchants were used for correct microstructural characterisation, including HNO_3, which have presented good results on γ_2 identification [4]. In Table 2 the etchant used, the objective and the methodology applied are presented. The SEM equipped with an energy dispersive X-ray (EDX) and electron backscattered diffraction (EBSD) system integrated unit system was used for further analysis before and after corrosion testing.

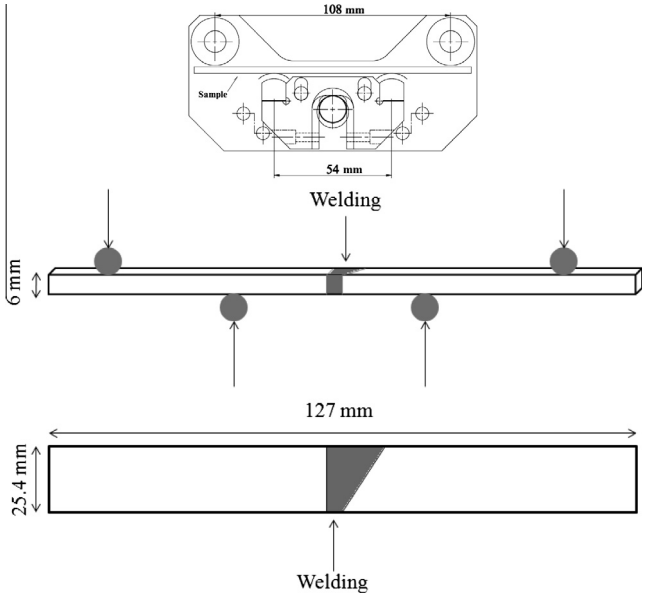

Fig. 1. Position of welded joint in the four points bend tests.

4. Corrosion testing

DL-EPR tests, based on the ISO 12732:2006 standard [20], were performed in a three-electrode cell configuration, using saturated calomel as reference and a platinum grid as counter-electrode. A portable IVIUM potentiostat was used for applied a scan rate potential of 0.56 mV s^{-1} that ranged from open circuit potential (E_{ocp}) until E_{ocp} + 700 mV_{sce} in the anodic direction and then back to the E_{ocp}, as suggested by ISO 12732. The electrolyte was a 3 M HCl fresh prepared solution. The quantitative evaluation of secondary phase through local depleted zones was performed using the relation of charges (Q_r/Q_a). The charges Q_a and Q_r were extracted from the current density (mA/cm^2) versus potential (V/SCE) curves