



# Isostatic diffusion bonding and post-solution treatment between Cr<sub>22</sub>Ni<sub>5</sub>Mo<sub>3</sub>MnSi and Cr<sub>30</sub>Ni<sub>7</sub>Mo<sub>3</sub>MnSi duplex stainless steels

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## ABSTRACT

In order to satisfy the requirement of different corrosion resistances in the same application, superplastic diffusion bonding at 1100 °C between expensive Cr<sub>30</sub>Ni<sub>7</sub>Mo<sub>3</sub>MnSi (SAE 2906, having higher corrosion resistance) and cheap Cr<sub>22</sub>Ni<sub>5</sub>Mo<sub>3</sub>MnSi (SAE 2205, having lower corrosion resistance) duplex stainless steels was systematically studied using a Gleeble-1500 thermo-mechanical simulator. The joint shear strength rapidly increased with an increase in holding time and pressure and then slowly increased when the holding time and pressure reached 5 min and 10 MPa, respectively. This increased strength was due to the void shrinkage and grain boundary migration across bonding interface. Post-solution at 1100 °C for 10 min significantly increased the joint shear strength up to 808 MPa, which was higher than 740 MPa of the base steel. It was ascribed to the dissolution of  $\sigma$  phase and further void shrinkage by volume diffusion. However, post-solution at 1050 °C deteriorated the bonding strength due to  $\sigma$  phase precipitation, while post-solution above 1100 °C also slowly reduced the bonding strength predominantly because of a decrease in the austenite fraction.

## 1. Introduction

Diffusion bonding, which is an important solid-state welding process, joins two faying surfaces by holding them at an elevated temperature (0.5–0.8 melting point of the materials) for a certain time under a selected pressure [1,2]. Whereas, the fusion welding, such as gas tungsten arc welding, is a conventional welding process involving the fusion zone (molten material) and heat affected zone. It is well known that, due to a high heat input, the microstructures in the fusion zone and heat affected zone are difficult to avoid the grain growth and the precipitation such as intermetallics and carbides, leading to the significant deterioration of mechanical properties and corrosion resistance [3]. In contrast, due to the absence of fusion zone and heat affected zone, the diffusion bonding, through matter diffusion across the interface, can achieve a strong and defect-free joint even having indistinguishable microstructure from base materials, resulting in the excellent mechanical properties [4,5]. Therefore, diffusion bonding has been widely applied in the welding of similar/dissimilar alloys. For example, a honeycomb structure of Ti–6Al–4V alloy was successfully manufactured by superplastic forming and diffusion bonding at 930 °C for 60 min under a pressure of 0.6 MPa [6]. The superplastic diffusion

bonding in similar magnesium AZ31 achieved a sound joint by holding at 400 °C for 120 min under a pressure of 3 MPa [7]. Except for similar diffusion bonding [6–8], dissimilar diffusion bonding has also attracted much attention [9–12]. Low temperature (650–800 °C) diffusion bonding between Ti–6Al–4V alloy and Cr<sub>22</sub>Ni<sub>5</sub>Mo<sub>3</sub>MnSi (SAE 2205) duplex stainless steel was successfully performed without any discontinuity along the interface [9]. In addition, a near  $\alpha$ -phase titanium alloy (Ti–4Al–2V) and austenitic stainless steel (0Cr18Ni9Ti) were successfully diffusion bonded at a pressure of 5 MPa for only 120 s with the assistance of phase transformation superplasticity induced by a cycling heating and cooling between 800 and 890 °C [13].

Duplex stainless steels simultaneously consist of austenite and  $\delta$ -ferrite, leading to a stronger strength than austenitic stainless steels and a better ductility than ferritic stainless steels [14]. Despite a lower nickel content in the duplex stainless steels, the pitting corrosion and weldability are equal or even better than austenitic stainless steels on the dependence of austenite-ferrite constituents [15]. These excellent properties attract much attention such as in the marine and nuclear industries. However, their applications are limited because the investigation on diffusion bonding between similar/dissimilar duplex stainless steels is insufficient. Zhang et al. systematically studied similar

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diffusion bonding of martensitic stainless steel (1Cr11Ni2W2MoV) with respect to holding temperature [16], holding time [17], external pressure [18] and surface roughness [19]. The diffusion bonding of ferritic stainless steels (Cr11Ni2MnSi) was also investigated by Sharma et al. [20,21] using impulse pressure, which accelerated bonding process by grain refinement along the bonding interface. However, the study and analysis on diffusion bonding of duplex stainless steels are few. Ridley et al. [22] and Islam et al. [23] just reported similar diffusion bonding of Cr25Ni7Mo3MnCu and Cr22Ni5Mo3MnCu duplex stainless steels, respectively. Yeh et al. [24] improved the welding properties of a duplex stainless steel (Cr23Ni6MoMnCu) by superplastic diffusion bonding. Noticeably, based on the literature review, only Komizo et al. [25] briefly studied the superplastic diffusion bonding between Cr18Ni4Mn3CuSi, SAE 2205 and Cr25Ni7Mo3MnCu duplex stainless steels.

Although our group has reported improved bonding properties between SAE 2205, Cr25Ni7Mo4MnSi and Cr30Ni7Mo3MnSi (SAE 2906) duplex stainless steels using phase transformation superplastic diffusion bonding [26,27], the present study is the first time to systematically investigate the effect of holding time and pressure on the isostatic diffusion bonding between SAE 2205 and SAE 2906. Furthermore, the properties of bonding interface were improved by the following post-solution treatment. The sound joint shear strength evaluated using lap shear test exceeds that of the base material (SAE 2205). It is noticed that this study offers useful information on the designation and production of submarine pipelines using expensive SAE 2906 and cheap SAE 2205 as outer and inner parts, respectively.

## 2. Experimental details

Commercially used duplex stainless steels were received in hot rolled condition, whose chemical compositions are listed in Table 1. According to the contents of Cr and Ni, they were referred to SAE 2205 and SAE 2906. Firstly, the samples were solution treated at 1350 °C for 40 min using resistance furnace (SRJX-8-13A) having an accuracy of  $\pm 5$  °C, followed by water quenching. Secondly, the samples were cold rolled by a reduction of 80%. Fig. 1 shows typically elongated microstructures after cold rolling. A larger fraction of austenite was observed in SAE 2906 because of higher contents of Cr and Ni.

The samples of  $15 \times 10 \text{ mm}^2$  for diffusion bonding were cut along the rolling direction using electric discharge machine. They were heated to 1100 °C at a rate of 20 °C/s using a Gleeble-1500 thermo-mechanical simulator in the Ar protective atmosphere, followed by adding the setting pressure (2, 5, 7, 10 and 20 MPa) in 5 s and holding at this pressure for different times (1, 5, 7, and 10 min). Post-solution treatment was performed between 1050 and 1350 °C for 3 ~ 20 min using the resistance furnace. For scanning electron microscopy (SEM) characterization, the samples were cut perpendicular to the bonding interface along the rolling direction. They were polished in a standard way and etched in a solution of 40 g NaOH and 100 ml water at a voltage of 6 V.  $\phi$  3 mm discs for transmission electron microscopy (TEM) characterization were machined from the center of the bonding interface. They were mechanically polished down to  $\sim 40 \mu\text{m}$  and finally twin-jet electropolished in a solution of 10 ml HClO<sub>4</sub> and 90 ml CH<sub>3</sub>CH<sub>2</sub>OH. These samples were characterized using Tecnai G2 F30 TEM operating at 300 kV.

The joint shear strength of the bonding interface was measured using lap shear specimen, as illustrated in Refs. [26,27], which ensured

**Table 1**

The chemical compositions of studied steels (wt. %).

	Cr	Ni	Mo	Mn	Si	C	S	P	N	Cu	V	Fe
SAE 2205	22.05	5.37	3.22	1.10	0.42	0.017	0.006	0.024	0.15	0.043	0.039	67.56
SAE 2906	30.06	7.26	2.86	0.53	0.25	0.0064	–	–	0.47	0.52	–	58.04

the fracture along the bonding interface. The test was carried out using WDW-50E universal testing machine at a constant speed of 10 mm/min. Two specimens for each condition were used for diffusion bonding and lap shear test at least.

## 3. Results

### 3.1. Isostatic diffusion bonding between SAE 2205 and SAE 2906

#### 3.1.1. Effect of holding time on the diffusion bonding

Fig. 2 shows the microstructures after diffusion bonding by holding at a pressure of 10 MPa for different times. Because of different microstructural constituents between SAE 2205 and SAE 2906 (Fig. 2), it was pretty hard to reveal their microstructures at the same etching condition. For SAE 2205, the austenite fraction after diffusion bonding became larger because of ferrite-to-austenite transformation (c.f. Figs. 1(a) and 2(a, b)). With an increase in the holding time from 1 to 10 min, the microstructure became coarser and more equiaxed (c.f. Fig. 2(a) and (b)). In addition, the TEM characterization shows a low dislocation density in equiaxed microstructure, and an increased grain size with an increase in the holding time (Fig. 3), indicating the recovery and recrystallization. For SAE 2906, in comparison with cold rolled microstructure (Fig. 1(b)), the recovery and recrystallization also occurred, resulting in a coarser microstructure (Fig. 2(c–f)). Similar to SAE 2205, increasing holding time led to more equiaxed microstructure.

Fig. 2(c–f) shows the bonding interfaces between SAE 2205 (upper unetched part) and SAE 2906 after holding for different times. There was a clear straight line dispersed with many voids after 1 min holding (Fig. 2(c)), indicating the position of bonding interface. These voids located at phase boundaries as revealed in Fig. 4(a). After 3 min holding, the bonding interface was decorated by a discontinue line dispersed with some voids (Fig. 2(d)). However, most of these voids were not along the phase boundaries any more but were in the interior of ferrite or austenite (Fig. 4(b)). After holding for 5 and 10 min, the bonding interface was clean and almost did not contain any voids (Fig. 2(e, f)). By TEM characterization, Fig. 4(c) shows very tiny voids, indicating that most of voids were closed during diffusion bonding. In addition, the precipitation of  $\sigma$  phase was observed, for all holding times, in the SAE 2205 adjacent to the bonding interface. An example was shown in Fig. 3(c).

Fig. 5 shows the shear strength of the joint significantly increased from 535 to 650 MPa with increasing the holding time from 1 to 5 min. It corresponded to a great improvement of the bonding interface (c.f. Fig. 2(c) and (e)). When the holding time continuously increased from 5 to 10 min, the joint shear strength slightly increased from 650 to 660 MPa because of comparable bonding interfaces (c.f. Fig. 2(e) and (f)). Thus, holding time of 5 min was chosen for the following experiments. The fracture surface after holding for 1 min exhibited inhomogeneous dimples where some of them were pretty shallow (Fig. 6(a)), indicating unbonded areas. With an increase in holding time up to 10 min (Fig. 6(b–d)), the dimples became more and more homogeneous. In addition, the  $\sigma$  phase precipitation acting as void nucleation sites were observed inside the dimples, which was identified by energy dispersed spectrum (Fig. 6(e)).

#### 3.1.2. Effect of holding pressure on the diffusion bonding

Fig. 7 shows the evolution of microstructure with holding pressure

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