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Isostatic diffusion bonding and post-solution treatment between Cr22Ni5Mo3MnSi and Cr30Ni7Mo3MnSi duplex stainless steels

Xueping Ren^a, Shuxia Li^b, Zhiping Xiong^{c,d,*}

^a School of Materials Science and Engineering, University of Science and Technology Beijing, Beijing 10083, PR China

^b Beijing Aeronautical Manufacturing Technology Research Institute, Beijing 10024, PR China

^c School of Materials Science and Engineering, Beijing Institute of Technology, Beijing 10038, PR China

^d Institute of Mechanics, Materials and Civil Engineering, Université catholique de Louvain, Louvain-la-Neuve 1348, Belgium

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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 Cr30Ni7Mo3MnSi (SAE 2906, having higher corrosion resistance) and cheap Cr22Ni5Mo3MnSi (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 σ phase and further void shrinkage by volume diffusion. However, post-solution at 1050 °C deteriorated the bonding strength due to σ 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-4 V 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 Cr22Ni5Mo3MnSi (SAE 2205) duplex stainless steel was successfully performed without any discontinuity along the interface [9]. In addition, a near α -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 δ 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

* Correspondence author.

E-mail address: zhiping.xiong@uclouvain.be (Z. Xiong).

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Received 27 February 2018; Received in revised form 3 June 2018; Accepted 8 June 2018 Available online 13 June 2018 1526-6125/ © 2018 The Society of Manufacturing Engineers. Published by Elsevier Ltd. All rights reserved. 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 postsolution 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. \varPhi 3 mm discs for transmission electron microscopy (TEM) characterization were machined from the center of the bonding interface. They were mechanically polished down to $\sim 40\,\mu m$ and finally twin-jet electropolished in a solution of 10 ml HClO₄ and 90 ml CH₃CH₂OH. 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

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 σ 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 σ 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

Table 1The chemical compositions of studied steels (wt. %)

The chemical compositions of studied steels (wt. %).												
	Cr	Ni	Мо	Mn	Si	С	S	Р	Ν	Cu	v	Fe
SAE 2205 SAE 2906	22.05 30.06	5.37 7.26	3.22 2.86	1.10 0.53	0.42 0.25	0.017 0.0064	0.006 -	0.024 -	0.15 0.47	0.043 0.52	0.039 -	67.56 58.04



Fig. 1. Microstructures of (a) SAE 22050 and (b) SAE 2906 after cold rolling at a reduction of 80%. δ is ferrite and γ is austenite. Cold rolling was horizontal.



Fig. 2. Microstructures adjacent to the bonding interfaces of (a, b) SAE 2205 and (c–f) SAE 2906 by holding at a pressure of 10 MPa for (a, c) 1 min, (d), 3 min, (b, e) 5 min and (f) 10 min. δ is ferrite and γ is austenite.

at a fixed holding time of 5 min. The microstructures of SAE 2205 were not etched (upper part) because of a large difference in the microstructure with SAE 2906. An example of SAE 2205 after bonding at a pressure of 10 MPa was shown in Fig. 2(b). The microstructures of SAE 2906 and SAE 2205 almost did not change with an increase in the pressure. In terms of bonding interface, a clear straight line including many voids can be seen after holding at a pressure of 2 MPa (Fig. 7(a)). Holding at a pressure of 5 and 7 MPa led to a discontinuous line



Fig. 3. Microstructures of SAE 2205 adjacent to the bonding interfaces after holding at a pressure of 10 MPa for (a) 1 min, (b, c) 5 min and (d) 7 min.

dispersed with fewer voids (Fig. 7(b, c)). After holding at a pressure of 10 MPa, almost no voids can be seen (Fig. 7(d)). However, by TEM characterization, Fig. 8(c) indicates that some tiny voids were observed even after holding at 20 MPa. With an increase in the holding pressure from 2 to 20 MPa, the size of voids decreased (Fig. 8(a–c)). All observed voids were not along grain boundaries but distributed in the interior of ferrite or austenite. In addition, σ phase precipitation in SAE 2205 adjacent to the bonding interface was detected for all pressures, as an example shown in Fig. 8(d).

Fig. 9 shows a significant increase in the joint shear strength from 517 to 650 MPa with an increase in the pressure from 2 to 10 MPa. It corresponded to a significant decreased number of voids as shown in Fig. 7. When the pressure reached 20 MPa, the joint shear strength only increased by 7 MPa. It related to that almost no voids were observed by SEM after holding at 10 MPa (Fig. 7(d)) and very small voids became closed by volume diffusion (Fig. 8(b, c)). Therefore, the holding pressure of 10 MPa was selected in the following experiments.

3.2. An improvement of the bonding interface by post-solution treatment

According to the above investigation on the effect of holding time and pressure, in the following experiments, the diffusion bonding was processed by holding at a pressure of 10 MPa for 5 min, which achieved a joint shear strength of 650 MPa. As the precipitation of σ phase (Figs. 3(a) and 8(d)) is harmful to the joint shear strength, the postsolution treatment is studied.



Fig. 5. The evolution of the joint shear strength with holding time at a pressure of 10 MPa. The standard deviation is within 10 MPa.

3.2.1. The dependence of the joint shear strength on solution time at 1350 $^\circ\mathrm{C}$

When post-solution treatment was at 1350 °C, the microstructure became more equiaxed, homogeneous and coarser with increasing solution time from 0 to 20 min (Fig. 10). Importantly, the voids along the bonding interface were almost not observed after solution treatments (Fig. 10(b-d)). The distribution of Cr and Fe across the bonding



Fig. 4. The diffusion bonding interfaces after holding at a pressure of 10 MPa for (a) 1 min, (b) 3 min and (c) 5 min. The interface in (a) was along the voids and the interfaces in (b, c) were adjacent to the voids.



Fig. 6. The fracture surfaces adjacent to SAE 2205 after holding at 10 MPa for (a) 1 min, (b) 3 min, (c) 5 min and (d) 10 min; (e) a representative spectrum for sigma phase precipitation.

interface was analyzed using energy dispersive spectroscopy. With an increase in solution time from 0 to 10 min (Fig. 11(a, c, e, g)), the concentration gradient of Cr and Fe became smaller, indicating their partitioning between SAE 2205 and SAE 2906. However, with further increasing solution time up to 20 min, the concentration gradient almost did not change, which was similar to that after 10 min holding (c.f. Fig. 11(g) and (j)). It indicated that holding for 10 min achieved full partitioning of Cr and Fe between SAE 2205 and SAE 2906.

Fig. 11(b, d, f, h, k) show the ductile fracture in all conditions. The σ phase was observed inside the dimples only when the solution time was between 0 and 7 min (Fig. 11(b, d, f)). In addition, with an increase in solution time from 0 to 7 min, the number and size of σ phase were reduced (Fig. 11(b, d, f)). When the holding time reached 10 min and above, there was almost no observation of the σ phase (Fig. 11(h, k)).

Fig. 12 illustrates the evolution of the joint shear strength with solution time. When the solution time increased from 0 to 10 min, the shear strength sharply increased from 650 to 750 MPa, which corresponded to the reduced size and amount of σ phase and increased partitioning of Cr (Fig. 11). When the solution time continuously increased up to 20 min, the shear strength slowly increased only up to 760 MPa because of almost full partitioning by holding for 10 min. Thus, 10 min for solution treatment was selected in the following experiments.

3.2.2. The dependence of the joint shear strength on solution temperature

The evolution of microstructure, bonding interface and fracture surface with solution temperature was shown in Fig. 13. With an increase in solution temperature from 1050 to 1350 °C, the amount of ferrite increased and the microstructure became coarser (Fig. 13(a, c, f) and Fig. 10(c)). When the solution temperature was 1050 °C, below the bonding temperature of 1100 °C, many voids along the bonding interface were still observed in Fig. 13(a). In addition, lots of σ phase were detected inside the dimples as shown in Fig. 13(b). When the solution temperature, no voids were observed along the bonding interface (such as examples in Fig. 13(c, f)). Besides, σ phase was not detected in ductile fracture surfaces (Fig. 13(d, g)).

When the solution temperature was the same to the bonding temperature of 1100 °C, the distribution of Cr and Fe was similar to that without solution treatment (c.f. Figs. 13(e) and 11(a)). It indicated that the partitioning of Cr and Fe was almost completed at 1100 °C during bonding process. With an increase in solution temperature from 1100 to 1350 °C, the concentration gradient became smaller, indicating that the partitioning of Cr and Fe was enhanced (Figs. 13(e, h) and 11(h)).

Fig. 14 shows the influence of solution temperature on the joint shear strength. When the temperature was 1050 °C, the shear strength (620 MPa) was even smaller than that before solution treatment



Fig. 7. Microstructures adjacent to the bonding interfaces after holding for 5 min under a pressure of (a) 2 MPa, (b) 5 MPa, (c) 7 MPa and (d) 10 MPa. δ is ferrite and γ is austenite.



Fig. 8. (a–c) The diffusion bonding interfaces after holding for 5 min at a pressure of (a) 2 MPa, (b) 10 MPa and (c) 20 MPa; (d) an example of σ phase precipitation in SAE 2205 at a pressure of 20 MPa. The interfaces in (a–c) were adjacent to the voids.

(650 MPa). It was ascribed to the formation of σ phase (Fig. 13(b)) as void nucleation sites during deformation (Fig. 13(b)). Increasing the temperature up to 1100 °C, the shear strength was significantly increased up to 808 MPa predominantly due to the dissolution of σ phase (c.f. Figs. 6(c) and 13(d)). However, when the temperature continuously increased from 1100 to 1350 °C, the shear strength slowly decreased predominantly due to a decrease in the fraction of hard

austenite (Fig. 13). In addition, the coarse microstructure was also harmful to the strength.



Fig. 9. The evolution of the joint shear strength with pressure at a fixed holding time of 5 min. The standard deviation is within 10 MPa.

4. Discussion

4.1. The analysis of diffusion bonding process

The diffusion bonding temperature was selected as 1100 °C due to the presence of superplasticity and the absence of σ phase precipitation in both studied steels [28–30]. It is well known that superplasticity is beneficial for diffusion bonding [31,32] and σ phase is harmful to the strength [33]. The external pressure (Fig. 7) and holding time (Fig. 2) predominantly affects the diffusion bonding process. When the pressure was applied, the asperities arising from surface roughness were first contacted. This limited contact area enlarged local stress and, in turn, plastic deformation occurred, leading to a planar array of voids [18]. This early contact process was largely accelerated by the localized superplastic deformation [32]. When the pressure was small, the voids were detected along the bonding interface and the size of voids was large (Figs. 7(a, b)). Increasing the pressure from 2 to 10 MPa helped reduce the void size and boosted void shrinkage, resulting in no void detection along the bonding interface by SEM (Fig. 7) and a great increase in the joint shear strength from 517 to 650 MPa. Similar phenomenon was observed in Refs. [22–25,34]. Larger pressure also can increase local deformation and indirectly accelerate atom diffusion [18], leading to a decreased void size (Fig. 8). With an increase in the pressure up to 20 MPa, the joint shear strength only slightly increased by 7 MPa. It is because the pressure of 10 MPa was enough to reach a tight initial interface contact. Islam et al. also reported a significant increase in the joint shear strength when the pressure increased from 3 to 10 MPa and a slight increase when the pressure continuously increased up to 12 MPa in the similar diffusion bonding of Cr22Ni5-Mo2Mn2SiN duplex stainless steels [23].

During holding at a certain pressure, the creep/superplasticity induced and accelerated grain boundary migration [18]. This was consistent with the voids along grain boundaries when the holding time was only 1 min (Fig. 4(a)) and the voids being in the interior of ferrite and/or austenite when the holding time increased up to 3 and 5 min (Fig. 4(b, c)). This grain boundary migration was beneficial to healing the bonding interface and reducing voids along the interface [17]. Furthermore, with an increasing in the holding time, mass transfer by volume diffusion can fill the voids by transportation atoms from adjacent areas (c.f. Fig. 4(b) and (c)). With a decrease in the amount (Fig. 2(c-e)) and the size (Fig. 4) of voids, the joint shear strength significantly increased from 535 to 650 MPa (Fig. 5) and the corresponding fracture surface became more ductile with more homogenous dimples (Fig. 6). Similar phenomenon was reported in Refs. [17,23,25,35]. For instance, Zhang et al. reported a decrease in the void size and an increase in the joint shear strength with an increase in the holding time from 1 to 90 min in a similar diffusion bonding of the 1Cr11Ni2W2MoV martensitic stainless steel [36]. When the holding time continuously increased from 5 to 10 min, the joint shear strength



Fig. 10. Microstructures adjacent to the bonding interfaces (a) before and (b–d) after post-solution heated at 1350 °C for (b) 5 min, (c) 10 min and (d) 20 min. δ is ferrite and γ is austenite.



Fig. 11. (a, c, e, g, j) The distribution of iron and chromium across the bonding interface and (b, d, f, h, k) fracture surfaces (a, b) before and (c–k) after post-solution at 1350 °C for (c, d) 3 min, (e, f) 7 min, (g, h) 10 min and (j, k) 20 min.



Fig. 12. The evolution of the joint shear strength with solution time by holding at 1350 $^\circ$ C. The standard deviation is within 10 MPa.

was only increased by 10 MPa due to the achievement of good bonding after holding for 5 min (c.f. Fig. 2(e) and (f)). Noticeably, diffusion time should be not too long because the microstructure would be coarsened (Fig. 3).



Fig. 14. The evolution of the joint shear strength with solution temperature by holding for 10 min. The standard deviation is within 10 MPa.

4.2. The analysis of post-solution treatment

After holding at 1100 °C for 5 min under a pressure of 10 MPa, the joint shear strength reached a high value of 650 MPa. However, it was lower than the shear strengths of 740 and 920 MPa respectively



Fig. 13. (a, c, f) Microstructures adjacent to the bonding interfaces, (b, d, g) fracture surfaces adjacent to SAE 2205 and (e, h) the distribution of iron and chromium across the bonding interface after post-solution for 10 min at (a, b) 1050 °C, (c–e) 1100 °C, (f, g) 1200 °C and (h) 1250 °C. δ is ferrite and γ is austenite.

obtained from SAE 2205 and SAE 2906 under the same diffusion bonding condition. The observed brittle σ phase (Fig. 3(c) and Fig. 8(d)) was harmful to the joint shear strength [33]. The formation of σ phase was due to the partitioning of Cr from SAE 2906 to SAE 2205 (Fig. 11(a, c, e, g, j)). In order to solve this problem, post-solution treatment was carried out. When the solution temperature was at 1350 °C, increasing the holding time from 0 to 10 min, σ phase was fully dissolved and the partitioning of elements was almost reached a balance (Fig. 11). As a result, the joint shear strength was significantly increased up to 750 MPa. With a further increase in the holding time up to 20 min, the shear strength was only increased by 10 MPa (Fig. 12). Noticeably, these values were higher than the shear strength of the base SAE 2205.

The solution temperature also has a great effect on the joint shear strength. When the solution temperature was 1050 °C, the joint shear strength was only 620 MPa (Fig. 14) due to the precipitation of σ phase (Fig. 13(b)). When the solution temperature was equal to the diffusion bonding temperature of 1100 °C, the σ phase was dissolved (Fig. 13(d)) and the voids were continuously shrunk due to further partitioning and volume diffusion (c.f. Figs. 13(e) and 11(h)). It led to a significantly increased joint shear strength of 808 MPa, which was much higher than the shear strength of the base SAE 2205. Although continuously increasing the solution temperature promoted partitioning and void shrinkage, a decreased fraction of hard austenite resulted in a decreased joint shear strength (Fig. 14). Thus, the post-solution at the same temperature of diffusion bonding was the best choice.

5. Conclusion

The effect of holding time and pressure on the dissimilar diffusion bonding between SAE 2205 and SAE 2906 duplex stainless steels was studied, followed by the investigation on the improvement of the joint shear strength using post-solution treatment, leading to the following conclusions:

(1) Holding at $1100 \,^{\circ}$ C for 5 min under a pressure of 10 MPa followed by post-solution at $1100 \,^{\circ}$ C for 10 min achieved a joint shear strength of 808 MPa. This value was larger than the shear strength of the base SAE 2205 processed by the same diffusion boning condition, indicating the achievement of a sound bonding.

(2) With an increase in the holding time and pressure during diffusion bonding, the joint shear strength was improved through local plastic deformation, void shrinkage by diffusion, and grain boundary migration across the bonding interface. In addition, the critical value of holding time and pressure were determined, before which the joint shear strength was fast increased and after which it was slowly increased.

(3) With an increase in the holding time when post-solution was at 1350 °C, the joint shear strength was first increased rapidly and then was increased sluggishly due to the quick dissolution of brittle σ phase and slow void shrinkage by volume diffusion.

(4) The post-solution at 1100 °C for 10 min achieved the best joint shear strength. When the solution temperature was lower than 1100 °C, the joint shear strength was deteriorated due to the precipitation of σ phase. When the solution temperature was higher than 1100 °C, although the bonding interface became seamless, the joint shear strength was reduced due to a decreased fraction of hard austenite.

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