Diffusion Bonding of Surface Self-nanocrystallized Ti–4Al–2V and 0Cr18Ni9Ti by Means of High Energy Shot Peening

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By means of high energy shot peening (HESP), surface nanostructured layers were synthesized on the ends of Ti–4Al–2V titanium alloy and 0Cr18Ni9Ti austenitic stainless steel bars. Using the peened surfaces for mating surface, constant temperature and pressure diffusion bonding (CTPDB) was applied to prepare joints of Ti–4Al–2V/0Cr18Ni9Ti in the temperature range of 800–900°C for 20 min under a uniaxial load of 8 MPa in vacuum. Tensile strengths of joints were tested, longitudinal section microstructures of joints were analyzed by using scanning electron microscope (SEM), meanwhile, concentrations of Fe, Ti, Cr and Ni in diffusion layers were measured by using energy dispersive spectroscopy (EDS), and the diffusion coefficient and diffusion activation energy of Fe atom in the Ti–4Al–2V side were calculated. Results show that the maximum tensile strength of 327 MPa is obtained when bonding temperature is 850°C, the diffusion activation energy of Fe atom is far less than that in coarse grain sample, and the diffusion coefficient of Fe atom is higher than that in conventional coarse grain sample.

KEY WORDS: high energy shot peening (HESP); constant temperature and pressure diffusion bonding (CTPDB); diffusion coefficient; diffusion activation energy.

1. Introduction

Displaying low density, high specific strength, rich corrosion resistance, the good high temperature strength and low temperature ductility, titanium and titanium alloy were applied widely to multiple fields such as the aerospace, petrochemical industry and nuclear power. But, due to the low Young's modulus and poor high temperature creep,^{1,2)} titanium alloy is commonly jointed with stainless steel by means of diffusion bonding to fabricate workpiece used in hot temperature environment. However, the tremendous physicochemical performance differences between titanium alloy and stainless steel results in formation of brittle intermetallic compounds, in the vicinity of interface, growth of grains, and the increment of joint deformation while bonding time is superlong and bonding temperature is overhigh, which deteriorate badly the performance of joints.

It is well known that nanocrystalline (nc) materials exhibit many novel properties relative to their coarse-grained counterparts, such as high strength and hardness, good superplasticity at low temperature and excellent wear property,³⁾ furthermore, a large number of defective grain boundaries (GBs) in nanostructured materials may act as fast atomic diffusion channels, hence greatly enhancing atomic diffusivities.^{3,4)} Since the Germany scientist H. Gleiter⁴⁾ firstly produced nc sample by means of metal gas condensation and in situ consolidation in the early 1980s, various preparation methods have been developed to synthesize bulk nc materials. K. Lu *et al.*⁵⁾ have developed a

new approach, surface self-nanocrystallization (SSNC), to prepare bulk nc materials. The nc materials from the technology include a large volume fraction of stored energy and non-equilibrium defective resulting in refining of coarse grains such as dislocations, vacancy, subboundary, which may promote chemical reaction of different atoms and enhance atomic diffusion coefficient.^{6,7)} The SSNC has been applied to chemical heat treatment of many metal materials, research results^{8–13)} showed that atomic diffusion coefficient were greatly enhanced in the nanostructured surface layer, so as to can decrease heat treatment temperature and shorten heat treatment time greatly, which will inaugurate new path for low-temperature and high-efficiency chemical heat treatment of metal materials.

However, so far, applying the SSNC to diffusion bonding of metal materials has not been read in all literatures. In present work, the SSNC was applied to diffusion bonding of Ti–4Al–2V titanium alloy and 0Cr18Ni9Ti stainless steel bars, aiming at increasing atomic diffusion coefficient in the course of diffusion bonding, reducing bonding temperature, shorting bonding time, and improving mechanical property of joints. And the strength of joints and Fe atomic diffusion kinetics in diffusion layer were researched.

2. Experimental

2.1. Experimental Materials

Commercial Ti–4Al–2V titanium alloy and 0Cr18Ni9Ti stainless steel bars ($\Phi 12$ mm $\times$ 30 mm in size), with compo-

Table 1. Chemical composition of tested materials (wt%).

Alloy	Fe	Ti	C	Si	Ni	Cr	S	Mn	P	Al	V	N	H	O
0Cr18Ni9Ti	rest	0.29	0.04	0.47	8.37	17.22	0.02	1.21	0.034	—	—	—	—	—
Ti-4Al-2V	0.069	rest	0.01	0.04	—	—	—	—	—	4.5	2.2	0.023	0.003	0.05

Fig. 1. Optical photographs of samples. (a) Ti-4Al-2V titanium alloy; (b) 0Cr18Ni9Ti stainless steel.

sition as shown in **Table 1**, were selected for experiment. To eliminate the effect of rolling deformation and obtain homogeneous coarse equiaxial grains, the Ti-4Al-2V bars were annealed for 3 h at 750°C in vacuum furnace, and the 0Cr18Ni9Ti stainless steel bars were hold 1h for solution treatment at 1100°C in vacuum furnace. After heat treatment, the Ti-4Al-2V microstructure constitutes mainly coarse equiaxial α grains (20–50 μm) and a small amount β phase (as shown in **Fig. 1(a)**), the stainless steel microstructure chiefly consist of austenite with twins and a small volume fraction martensite(as shown in **Fig. 1(b)**).

2.2. Surface Self-nanocrystallization Treatment

Before diffusion bonding, the ends of Ti-4Al-2V titanium alloy and 0Cr18Ni9Ti stainless steel bars were treated to obtain surface nanostructured layer by means of HESP. **Figure 2(a)** shows a schematic illustration of the HESP treatment set-up used in the present work. Steel balls of 1 mm in diameter were thrown with high velocity to the end face of the bar samples under peening pressure of about 0.6 MPa. Each peening of the ball will result in severe plastic deformation on the surface of the treated sample (as shown in **Fig. 2(b)**),⁸⁾ repeated impacting on the sample surface over short period of time (in this work, treatment time is 5 min) leads to a progressive refinement of coarse grains into nanometer regime. Prior to the treatment, the treated ends were polished manually with silicon carbide papers to 1200[#] and then mechanically polished with Al₂O₃ polishing powder to achieve finish surface. It was note that a restrain sleeve part with 2₋₀^{+0.5} mm inner diameter was machined to grip the treatment end of sample and impede metal to flow away rim of the surface (as shown in **Fig. 2(a)**).

Longitudinal-sectional observations of the treated samples were preformed on a TESCAN VEGA II scanning electron microscope (SEM, operating at a voltage of 30 kV). Microstructure features in the surface layer were characterized by using a Philip Tecnai-20 transmission electron microscope (TEM, operating at a voltage of 200 kV). Plane-view and cross-sectional thin foils for TEM observations were prepared by means of cutting, grinding, dimpling and a final ion thinning at low temperatures. The quantitative X-ray diffraction (XRD) analysis of the surface

Fig. 2. (a) The illustration of the HESP treatment set-up and (b) the local plastic deformation in surface layer by the peening of the shot.

layer in the HESP treated samples were carried out on a D/max-2400 X-ray diffractometer (12 kW), with Cu K_{α} radiation (wavelength $\lambda_{\alpha 1}=0.154056$ nm, $\lambda_{\alpha 2}=0.154439$ nm was eliminated by a graphite monochromator using its (0002) reflection), slit parameters of the diffractometer are 1°–1°–0.30 mm. The average grain size and mean microstrain on the top surface were calculated from line broadening of Bragg diffraction peaks.

2.3. Diffusion Bonding Experiment

After the HESP treatment, the roughness of the treated surface was increased in some short. In order to satisfying with interface condition of diffusion bonding, treated surface used for mating surface of diffusion bonding were prepared by conventional metallographic technology, cleaned in acetone and dried in air. Constant temperature and pressure diffusion bonding (CTPDB) of Ti-4Al-2V titanium alloy and 0Cr18Ni9Ti stainless steel bars was carried out in the Gleeble 1500-D vacuum chamber, under 8 MPa uniaxial load for 20 min, the diffusion bonding process drawing is shown in **Fig. 3**, the bonding temperature are 800, 825, 850, 875 and 900°C respectively, both heating rate and cooling velocity are set as 5°C/s. Two joints were prepared at each bonding temperature, one was experimented for tensile-strength, and other one was used for observed microstructure of longitudinal section.

After diffusion bonding, joints were annealed for 1 h at 400°C in vacuum furnace to eliminate bonding stress.

Tensile samples were machined to $\Phi 10\text{ mm} \times 50\text{ mm}$ in size, and experimented on the Instron-1342 style instron. Longitudinal sections were taken from joints and prepared by usual grinding and polishing techniques. The titanium alloy and stainless steel sides were etched with Kroll reagent (6 mL HCl, 2 mL FH and 92 mL H₂O) and aqua regia (60 mL HCl, 20 mL HNO₃ and 20 mL H₂O), respectively. Microstructures in the reaction zone were observed by a TESCAN VEGA II scanning electron microscope (SEM, operating at a voltage of 30 kV), meanwhile, distributions of Fe, Ti, Cr and Ni in diffusion layers on the titanium alloy side were monitored by using a fully quantitative X-ray energy dispersive spectroscope (EDS).

3. Results

3.1. Microstructure of Surface Self-nanocrystallization Treatment

After HESP treatment of samples, the longitudinal-sectional SEM morphology of titanium alloy shows a plastic deformation layer of about $253.52\text{ }\mu\text{m}$ was generated below top surface (as shown in Fig. 4(a)), the plastic deformation was gradually exacerbated with decreasing of depth to the top surface, while the depth was reduced to about $121.81\text{ }\mu\text{m}$, severe plastic deformation was created due to

Fig. 3. Technology curve of constant temperature and pressure diffusion bonding.

greater stress and strain. Cross-sectional TEM morphology about $20\text{ }\mu\text{m}$ depth to the top surface in titanium alloy indicates nanograins with about 50 nm were formed (as shown in Fig. 4(b)), the corresponding selected area electron diffraction (SAED) pattern shows these nanograins have nearly random crystallographic orientation. Within $50\text{ }\mu\text{m}$ below the top surface in the titanium alloy samples, nanograins with below 100 nm were produced after 5 min HESP.

In the 0Cr18Ni9Ti stainless steel samples, the plastic deformation layer thickness was about $615.84\text{ }\mu\text{m}$, and the severe plastic deformation layer was about $184.56\text{ }\mu\text{m}$, as shown in Fig. 4(c). Cross-sectional TEM morphology about $30\text{ }\mu\text{m}$ depth shows that equiaxial nanograins with about 60 nm were formed in the 0Cr18Ni9Ti stainless steel sample, the corresponding SAED pattern reveals nanograins have random crystallographic orientation (as shown in Fig. 4(d)). By TEM observation, nanograins layer about $80\text{ }\mu\text{m}$ thickness was generated in the 0Cr18Ni9Ti after HESP.

On the top surface of treated samples, Full Width at Half Maximum (FWHM) of diffraction peaks were measured by the MDI Jade5.0 (a X-ray analysis software). The average grain size and mean microstrain on the top surface were calculated from broadening of diffraction peaks, by using formula as follows:

$$\frac{FW(S) \cdot \cos(\theta)}{\lambda} = \frac{K}{D} + 4 \cdot \varepsilon \cdot \frac{\sin \theta}{\lambda} \quad \text{.....(1)}$$

Where $FW(S)^D = FWHM^D - FW(I)^D$ is FWHM of peaks after detracting broadening led by the diffractometer itself; K is a constant, $K=0.89$ here; λ is the wave length of the $\text{Cu } K_{\alpha 1}$ irradiation; θ is Bragg diffraction angle; D is mean grain size; ε is mean lattice strain. By performing a least square fit $(FW(S) \cdot \cos(\theta)/\lambda)$ plotted against $(\sin \theta)/\lambda$ for all of the measured peaks of a sample, one is able to determine the mean grain size D and the mean lattice strain ε (as shown in Table 2).

Fig. 4. SEM and TEM of treated samples by means of HESP. (a) Longitudinal section SEM of treated Ti-4Al-2V; (b) TEM of treated Ti-4Al-2V at the $20\text{ }\mu\text{m}$ depth from top surface; (c) longitudinal section SEM of treated 0Cr18Ni9Ti; (d) TEM of treated 0Cr18Ni9Ti at the $30\text{ }\mu\text{m}$ depth from top surface.

Table 2. Average grain size and lattice distortion of treated samples.

Alloy	Average grain size (nm)	Average lattice distortion (%)
Ti–4Al–2V	33.3	0.087
0Cr18Ni9Ti	50.0	0.170

Table 3. Tensile strength and compressibility of bonded joints.

Bonding temperature (T/°C)	Tensile strength (σ_b /MPa)	Compressibility (ϵ /%)
800	Fracture when machining	0.03
825	250	1.13
850	327	4.43
875	275	7.12
900	229	15.5

3.2. Tensile Strength of Joints

Tensile strength of joints at different bonding temperature are summarized in **Table 3**, thereinto, the strength of joint at 800°C has not been obtained due to fracture of joint while machining. From the Table 3, one can know the maximum strength of 327 MPa with 4.43% compressibility was achieved while bonding temperature being 850°C, subsequently, the bond strength decrease with increasing of bonding temperature, at 900°C, the minimum strength of 229 MPa was achieved. In the course of diffusion bonding, compressibility of joints is increasing with rising of bonding temperature.

CTPDB of titanium alloy and stainless steel have been already researched quite a bit, but bond strength of joint has not been enhanced effectively, improving joint strength is still effort of material researcher. So far, in the all literatures about CTPDB of titanium alloy and stainless steel, the maximum tensile strength of 306 MPa, which CP Ti and MDSS (commercial pure Ti and micro-duplex stainless steel) was bonded for 90 min at 850°C and 3 MPa pressure, was obtained by S. Kundu *et al.*,¹⁴⁾ obviously, the bonding time is superlong and efficiency is lower in the process. CTPDB was applied to produce joints of TA1 alloy and 1Cr18Ni9Ti stainless steel by Yingcai Zhang,¹⁵⁾ and the joint strength reached only 31.5–207.5 MPa at 800–1100°C, 50–100 MPa, for 30–180 min, in the bonding process, bond time is longer, upper limit of bond temperature is higher, and the joint strength have not been enhanced effectively. In our prior period researches, CTPDB was applied to produce joints of Ti–4Al–2V alloy and 0Cr18Ni9Ti stainless steel with coarse grains, the maximum strength of 264 MPa was achieved at 850°C, 8 MPa for 30 min.¹⁶⁾ In present work, tight junctions of Ti–4Al–2V and 0Cr18Ni9Ti bars with nanostructured interface were obtained by means of CTPDB, the maximum strength of 327 MPa was achieved within no more than 20 min, which shows surface self-nanocrystallization treatment for bond interface redounds to improvement of the bond strength.

3.3. SEM and EDS of Joints

The SEM image of the transition joints are given in **Fig.**

Fig. 5. SEM and concentration profiles of major elements across interface.

5. On the longitudinal sections, A, B, C, D and E regions were divided in the light of microstructures feature. Region A is titanium alloy matrix consisted of fine grains (less than 10 μm in size), region B is acicular α – β Ti occurring from the decomposition of β -Ti during cooling, region C is β -Ti solid solution, region D is intermetallic compound which presents black groove after etched, and region E is stainless steel base.¹⁷⁾ Longitudinal sections of all joints also are consisted of five regions mentioned above, just that each region magnitude vary with increasing of bonding temperature (as shown in Fig. 8). Corresponding line scanning energy spectrum (in the Fig. 5) shows Fe, Cr and Ni atoms have diffused into titanium alloy with different distances, and Ti atoms diffused into stainless steel also. Diffusion distances of Fe and Ti atoms were measured respectively, and summarized in **Table 4**, they both do not include thickness of intermetallic compound, *i.e.* the thickness of region D. The Table 4 shows the diffusion distances of Fe and Ti atoms increase with increasing of bonding temperature, the minimum diffusion distances of Fe and Ti are 17.8 μm and 3 μm respectively when bonding temperature is 800 °C. Comparing with in coarse grain sample, diffusion distances of Fe and Ti in the sample with nanostructured microstructures are increased aggressively. Literature¹⁸⁾ shows the diffusion distance of Fe atom in coarse grain titanium alloy sample is only about 10 μm when Ti–4Al–2V and 0Cr18Ni9Ti was bonded for 30 min at 875°C and 8 MPa, at

Table 4. The diffusion distances of Fe and Ti atoms at different temperatures.

Bonding temperature (T/°C)	800	825	850	875	900
Fe diffusion distance (d/μm)	17.8	24.4	32.0	32.5	32.4
Ti diffusion distance (d/μm)	3.0	4.0	6.1	9.2	10.2

Table 5. Measured concentration of Fe by EDS in diffusion zone.

Temperature (T/°C)		Distance to interface (x/μm) and concentration of Fe (C/at.%)						
800	x	5.30	10.50	15.80	21.0	26.30	31.6	36.8
	C	7.89	5.74	2.56	1.37	1.13	0.64	0.27
825	x	4.42	8.84	13.26	17.68	22.10	26.52	30.94
	C	10.18	6.77	5.38	4.70	3.14	3.49	1.32
850	x	2.50	7.80	13.10	18.40	23.70	29.00	34.30
	C	14.20	9.30	7.01	4.73	3.94	2.92	1.28
875	x	3.00	8.31	13.42	18.73	24.04	29.35	34.66
	C	14.21	8.51	7.46	5.03	2.94	2.03	1.38
900	x	2.00	9.30	16.60	23.90	31.20	38.50	45.80
	C	14.30	9.03	6.87	4.74	3.67	2.22	1.81

Fig. 6. Profiles of Fe atom concentration *versus* diffusion distance on titanium alloy sides. (a) Profiles of Fe atom concentration *versus* diffusion distance at 850°C; (b) profiles of Fe atom concentration *versus* diffusion distance at 900°C.

the same time, Ti atom diffused hardly into in coarse grain stainless steel base, and only formed intermetallic compound on the bonding interface.

In diffusion layer on the titanium alloy sides, concentrations of Fe atom were measured by EDS and summarized in **Table 5**.

3.4. Diffusion Kinetics of Fe Atom

The diffusion coefficients of Fe atom in titanium alloy sides at different bonding temperatures can be gained according to Fick's law. New phase should not be formed in diffusion layer in the light of the Fick's law. However, multiple intermetallic compounds generated in the interface of joints. But, when the concentration of Fe atom is no more than its solid solubility in Ti–4Al–2V, the Fick's law can be still applied to calculate the diffusion coefficient of Fe atom in Ti–4Al–2V alloy. The maximum solid solubility of Fe atom in β -Ti at 595°C is 15at% according to Fe–Ti binary phase diagram,¹⁹⁾ therefore, while calculating the diffusion coefficient of Fe atom in titanium alloy, its concentration in β -Ti should be no more than 15at%. In addition, in the present work, the diffusion of Fe atom in β -Ti is equal to the diffusion in semi-infinite rod, the initial concentration of Fe atom in stainless steel base was supposed C_0 , therefore,

the concentration of Fe atom on the diffusion interface should be $C_0/2$, and C_0 should be no more than 30at%.

When diffusion coefficient in different concentrations is constant, the Fick's law may be expressed by Eq. (2):

$$\frac{\partial C}{\partial t} = D \frac{\partial^2 C}{\partial x^2} \dots\dots\dots (2)$$

Where D is diffusion coefficient of Fe atom, C is concentration of Fe at x point, and t is diffusion bonding time. According to the Eq. (2), if the diffusion coefficient D maintains a constant at different concentration, the concentration C should be quadratic function of diffusion distance.

According to datum in the Table 5, scatters of Fe atom concentration *versus* distance (x) to diffusion interface in titanium alloy sides were drawn (as shown in **Fig. 6**). Following quadratic function, scatters were simulated into quadratic curves by Matlab software. Results shows the simulated profiles are great agreement with the measured datum, as shown in Fig. 6, which demonstrate the diffusion coefficient of Fe atom in titanium alloy is a constant at specific temperature. Therefore, the Eq. (2) can be applied to deduce diffusion coefficients of Fe atom in titanium alloy.

Aiming to shortening length of the paper, it is well note that just simulated profiles at 850°C and 900°C are listed in

Table 6. Diffusion coefficient and diffusion activation of Fe atom at different temperature.

Bonding temperature (T/°C)	800	825	850	875	900
Diffusion coefficient in nanostructured sample ($D/\text{m}^2\text{s}^{-1}\times 10^{-13}$)	1.059	1.378	1.555	2.143	2.502
Diffusion coefficient in coarse grain sample ^[18] ($D/\text{m}^2\text{s}^{-1}\times 10^{-13}$)	0.540	0.83	1.095	1.428	-----
Diffusion activation energy in nanostructured sample ($Q/\text{kJ}\cdot\text{mol}^{-1}$)			71.61		
Diffusion activation energy in coarse grain sample ^[18] ($\text{kJ}\cdot\text{mol}^{-1}$)			202.75		

the paper (as shown in Fig. 6).

According to the above simulated curves, the Fe atom concentrations range between 13–15at% while x being equation to 0, which indicates the initial concentrations of Fe atom are no more than 30at% and meeting the above supposed condition, therefore, the C_0 may be determined as 30at%.

While D being a constant, the Eq. (2) has following error function (3):

$$C(x,t) = \frac{C_0}{2} \left[1 - \operatorname{erf} \left(\frac{x}{2\sqrt{Dt}} \right) \right] \dots\dots\dots(3)$$

Thereby, the Eq. (2) may be transformed into Eq. (4) as following:

$$\operatorname{erf} \left(\frac{x}{2\sqrt{Dt}} \right) = 1 - \frac{2C}{C_0} \dots\dots\dots(4)$$

Supposing, $Z = x/2\sqrt{Dt}$, Z can be determined according to error function table. Consequently, the diffusion coefficient D can be obtained following Eq. (5):

$$D = \frac{x^2}{4Z^2t} \dots\dots\dots(5)$$

At last, diffusion coefficients of Fe atom in titanium were obtained at different temperatures, as shown in **Table 6**. The results shows the diffusion coefficients of Fe atom in Ti–4Al–2V alloy with nanostructured surface layer are larger than that in coarse grain counterpart, and the diffusion coefficients of Fe are increased with increasing of bonding temperature.

According to Arrhenius diffusion equation,²⁰⁾ a equation can be deduced here as following Eq. (6):

$$D = D_0 \exp \left(-\frac{Q}{RT} \right) \dots\dots\dots(6)$$

Where D_0 is diffusion coefficient constant, Q is mole activation energy, R and T are gas constant and Kelvin temperature respectively. In term of specific material, D_0 and Q are constants which are independent of temperature and can be determined by experiment.

Fig. 7. Plot of $\ln D$ versus $1/T$ in the temperature range of 800–900°C.

The Eq. (6) can be transformed into Eq. (7):

$$\ln D = \ln D_0 - \frac{Q}{RT} \dots\dots\dots(7)$$

Obviously, $\ln D$ becomes linear function of $1/T$. Therefore, D_0 and Q can be determined by simulating profile of $\ln D$ versus $1/T$, and Eq. (6) can be determined also. Thus, the diffusion coefficient of Fe atom in the Ti–4Al–2V can be obtained in different bonding temperature. The straight-line of $\ln D$ versus $1/T$ can be plotted according to diffusion coefficients in the Table 6, as shown in **Fig. 7**, and a simulated Eq. (8) can be obtained:

$$\ln D = -8.622 \times 1000/T - 21.79 \dots\dots\dots(8)$$

According to Eqs. (7) and (8), the D_0 Q and can be determined as following:

$$D_0 = 3.4413 \times 10^{-10} \text{ m}^2 \cdot \text{s}^{-1}$$

$$Q = 71.61 \text{ kJ} \cdot \text{mol}^{-1}$$

Substituting the values of D_0 and Q into Eq. (6), the equation can be transformed into Eq. (9):

$$D = 3.4413 \times 10^{-10} \exp \left(-\frac{71605.73}{RT} \right) \dots\dots\dots(9)$$

4. Discussion

Prior period research of the present work shows, in the course of surface self-nanocrystallization by means of

HESP, coarse grains in Ti-4Al-2V alloy are transformed into low density single-system twin lamella under low stress and strain rate, with increasing of stress and strain rate, the twin density is incremented and multi-system twin is generated. While the strain rate is increased to a certain level, dislocation slip is excited in fine twin lamella and formed high density dislocation walls (DDWs) on the twin boundaries, and then fine twin lamella is divided into low angle subgrains, with further increasing of stress and strain rate, sub-boundaries are rotated due to dislocation slip, at last, large angle equiaxial nanograins with an average size of 35 ± 5 nm are formed on the top surface, and some $50 \mu\text{m}$ thick surface nanostructured layer with nanograins no more than 100 nm are formed.^{21,22)} However, as for 0Cr18Ni9Ti stainless steel, under low stress, dislocations slip along themselves gliding planes and accumulate because of intercrossing each other, with increasing of stress and strain rate, twin deformation are induced on the dislocation accumulating place, and then multi-system twin deformation refined coarse grains within $80 \mu\text{m}$ to top surface into nanograins with below 100 nm, at the same time, induced martensite phase transformation are yielded on the twins intersections.^{23,24)}

Therefore, surface nanostructured layers on the ends of Ti-4Al-2V and 0Cr18Ni9Ti were formed after surface self-nanocrystallization by means of HESP, and a great many of non-equilibrium grain boundaries and subboundaries were introduced into the nanostructured layer. Research¹⁰⁾ shows the ultrafinegrained structure provides about 30 vol% grain boundaries for average grain size of 10 nm, and a few volume percents triple junctions. In present work, at certain diffusion bonding temperature, these numerous non-equilibrium grain boundaries and subboundaries in the nanostructured layer act as fast atomic diffusion channels. Meanwhile, these numerous non-equilibrium phase possess higher Gibbs free energy which may facilitate the chemical reaction than that conventional grain boundaries, which decreased the atomic diffusion activation energy¹⁰⁾ (as shown in Table 6). Therefore, in the course of diffusion bonding of Ti-4Al-2V and 0Cr18Ni9Ti, atomic diffusion in the diffusion layer with nano-microstructures is boosted. Comparing with coarse grain samples, the diffusion coefficient and the diffusion distance of atom in the HESP treated samples are larger, as shown in the mentioned above 3.3 and Table 6.

Figure 8 shows the longitudinal section SEM images of bond joints at different bonding temperature, regions A, B, C, D and E correspond with ones respectively in the Fig. 5. Obviously, the breadth of region B (acicular α - β Ti), C (β -Ti solid solution) and D (intermetallic compound) become increasing with rising of bonding temperature.

CTPDB was applied to bonding of CP-Ti and AISI304 stainless steel at 850°C , 3 MPa and 120 min by M. Ghosh *et al.*, and the maximum bond strength of 222 MPa was obtained, M. Ghosh^{25,26)} held reduction of bond strength resulted chiefly from increasing of brittle intermetallic compound and segregation layer, as well as growth of grains in the diffusion layer. Literature²⁷⁾ insisted bonding property of titanium alloy and stainless steel is chiefly dependence of the thickness of brittle intermetallic compound and combination strength of base materials, early stage of diffusion bonding, the intermetallic compound influences slightly the

Fig. 8. Longitudinal section SEM of joints at different temperature.

bond strength due to no formation of continuing compound layer on the bond interface, $3\text{--}5 \mu\text{m}$ thick continuous brittle intermetallic compound was generated when bonding temperature is exorbitant or bonding time is more long, which decreased heavily the bond strength. In present work, while bonding temperature being less than 850°C , atoms have not diffused completely in diffusion layer after 20 min diffusion bonding, and intermetallic compound is not chief factors of influencing bond strength. However, intermetallic compound layer was increased obviously when the bonding temperature was more than 850°C , which decreased the bond strength primarily after 20 min diffusion bonding (as shown in Table 3), the thickness of intermetallic compound is about $2 \mu\text{m}$ at 850°C , and it is about $3.5 \mu\text{m}$ at 900°C (as shown in Fig. 8), the experiment result is good consistent with research result in the literature.²⁷⁾

Due to fine nanograins and plenty of grain boundaries in the diffusion layer after surface selfnanocrystallization, in the course of diffusion bonding, Fe, Cr, Ni and Ti atoms can diffuse into matrix as soon as possible before nanograins do not grow completely, which result in reducing the thickness of intermetallic compound. Furthermore, after diffusion bonding, grains (about several micron in size, as shown in Fig. 8) closing diffusion interface are homogeneous and far less than convention coarse grain dimension. These factors facilitate to improve the property of joints. However, as for counterpart coarse grain samples, a plenty of atoms can not diffuse quickly into matrix, which results in accumulating a majority of atoms at the interface and increases the thickness of intermetallic compound. Consequently, at same diffusion bonding process, the bond strength of SSNC sample is higher than that of convention sample.

5. Conclusion

(1) On the ends of Ti-4Al-2V titanium alloy and 0Cr18Ni9Ti austenite stainless steel bars, some $50 \mu\text{m}$ and $80 \mu\text{m}$ thickness surface nanostructured layer consisting of nanograins with below 100 nm were obtained respectively by means of HESP.

(2) Using the treated surface for mating surface, CTPDB was applied to prepare joints of Ti-4Al-2V titanium alloy and 0Cr18Ni9Ti austenite stainless steel, and tightening joints were obtained in the temperature rang of $800\text{--}900^\circ\text{C}$ for 20 min under a uniaxial load of 8 MPa in vacuum. When bonding temperature was at 850°C , the maximum tensile strength of 327 MPa was obtained.

(3) In the course of diffusion bonding of treated Ti-4Al-2V alloy and 0Cr18Ni9Ti stainless steel bars, the diffusion activation energy of Fe atom in titanium alloy is far less than that in coarse grains sample, the diffusion coefficient of Fe atom is higher than that in coarse grain sample, the bond strength of joints are increased greatly. Therefore, the surface self-nanocrystallization for mating surface by means of HESP redounds to improving tensile property of joints between Ti-4Al-2V alloy and 0Cr18Ni9Ti stainless steel rods.

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