Interfacial evolution and mechanical behavior of explosively welded titanium/steel joint under subsequent heat treatment process

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Abstract
Explosive welding (EXW) has advantages in joining metallurgically incompatible metals due to its high quality and simple process. Successful titanium/steel clad plates fabricated by explosive welding have been extensively reported in recent years. However, no research has been reported on the fabrication of explosively welded TA2/15CrNi3MoV clad plates and the effect of subsequent heat treatment processes on the clad plates. This study aims to clarify the microstructure evolution and mechanical behavior of explosively welded TA2/15CrNi3MoV clad plates during the subsequent heat treatment process. The explosively welded TA2/15CrNi3MoV clad plates were heat-treated in the temperature range of 700–1000 °C for 30 min. The results elucidated that the heat treatment process accelerated the interdiffusion of constituent elements (e.g., Fe, Ti, and C) at the TA2/15CrNi3MoV interface, which formed FeTi, Fe2Ti, and TiC. For the samples heat-treated at 700 °C, a continuous TiC layer was generated at the TA2/15CrNi3MoV interface, which hampered the formation of FeTi and Fe2Ti. However, for the samples heat-treated at 800 °C, the continuous TiC layer was broken up, which resulted in a significant increase in the Fe-Ti layer to 1.21 μm. This microstructure evolution dramatically dropped the tensile strength to 329.2 MPa. For samples heat-treated in the temperature range of 900–1000 °C, the thickness of the brittle Fe-Ti layer dramatically increased and caused a rapid drop in tensile strength.

Keywords Ti/steel clad plates · Titanium · Mild steel · Explosive welding · Heat treatment

1 Introduction
Clad plates have been increasingly designed due to their potential ability to combine the advantages of base and clad materials [1]. Clad plates such as Al/Mg [2], Ti/Steel [3], and Ti/Al [4] have been successfully fabricated and applied in recent years. Among these clad plates, the pure titanium TA2 and high strength 15CrNi3MoV steel clad plates in this work have broad application prospects in corrosion-resistant and pressure-bearing equipment. The main reason is that TA2 plates have superb corrosion resistance and a high strength/density ratio [5]. However, the high manufacturing cost hinders the widespread application of TA2 [6]. To overcome this problem, 15CrNi3MoV steel with high strength and low processing cost was introduced to fabricate TA2/15CrNi3MoV clad plates to reduce the consumption of TA2. Therefore, TA2/15CrNi3MoV clad plates have a high strength/density ratio, superb corrosion resistance, and low fabrication costs, and they are widely used in pipes, vessels, and marine equipment.

However, due to the metallurgical incompatibility of titanium and steel, brittle intermetallic compounds (IMCs) are easily generated at the titanium/steel bonding interface, which makes it difficult to manufacture titanium/steel clad plates by conventional fusion welding methods [5, 6]. In addition, due to mismatches in their mechanical properties (e.g., heat transfer and thermal expansion), massive residual stresses are produced in the Ti/steel interface region and deteriorate the bonding interface [7]. To avoid residual stresses and brittle IMCs at the Ti/steel interface, rolling bonding [8], diffusion bonding [9], and explosive welding [10, 11] were employed. Among these solid-state metal joining processes, explosive welding is a reliable technology to join titanium and steel due to its simple process and high quality [11].
Explosive welding (EXW) offers an efficient alternative to joining metallurgical incompatible metals [12, 13]. It produces a weld joint via the high-speed collision of the flyer plate and base plate caused by the vast energy of detonation [14, 15], as shown in Fig. 1a. The explosive welding interface plays a crucial role in the mechanical properties of the clad plates. Previous papers [16] revealed that the explosively welded interface underwent fast heating \((10^9 \text{ K/s})\) and cooling \((10^7 \text{ K/s})\), which generated a melted zone at the explosively welded interface; hence, a metallic bond was formed. This high heating and cooling rate during explosive welding reduces the interdiffusion of atoms and minimizes the formation of intermetallic compounds at the explosively welded interface [17]. However, the explosive welding method is not ideal for complex geometric shapes [18, 19]. Consequently, explosive welding is often used in combination with other welding methods, such as tungsten inert gas (TIG), laser beam welding (LBW), and electron beam welding (EBW) [13, 18, 19]. Gao et al. [18] achieved the dissimilar laser welding of Ti alloy to steel by using TA2/Q235 explosive welded clad plates. However, Wang et al. [19] reported that brittle intermetallic components such as FeTi and Fe_2Ti were formed at the TA10/Q345 explosive welded interface during the electron beam welding of titanium alloy and mild steel using the explosive welded TA10/Q345 transition joint.

![Fig. 1 Process diagram of a explosive welding, b heat treatment processing, c microstructural characterization, and d mechanical property tests](image-url)
During these various subsequent manufacturing processes, the Ti/steel explosive welding interface may be exposed to high temperatures. Therefore, it is essential to study the heat treatment processing on Ti/steel explosive-bonded joints to reveal the temperature effects on the bonding interface. Previous literature reported that brittle intermetallic components such as Fe₂Ti and FeTi were generated in the bonding interface due to the interdiffusion of constituent elements [19, 20]. These intermetallic compounds deteriorate the Ti/steel interface strength [21, 22]. Meanwhile, residual stresses are produced at the interface due to the mismatch in their mechanical properties [17]. However, the effect of the heat treatment temperature on explosively welded TA2/15CrNi3MoV clad plates remains unknown.

This work studied the influences of the heat treatment temperature on the tensile strength and microstructure of the explosively welded TA2/15CrNi3MoV interface. The microstructure evolution of the explosively welded TA2/15CrNi3MoV interface during the subsequent heat treatment process was systematically investigated, and the influence of the heat treatment temperature on the tensile strength of the explosively welded TA2/15CrNi3MoV interface was carefully discussed.

2 Materials and methods

In this investigation, Ti/steel explosively welded clad plates were fabricated by commercial purity titanium (TA2) plates and mild steel (15CrNi3MoV) using the explosive welding method. The TA2 plate was chosen as the flyer plate, and the 15CrNi3MoV plate was chosen as the base plate during the explosive welding process, as shown in Fig. 1a. The explosive material in this article was a mixture of ammonium nitrate fuel oil. The chemical compositions of the TA2 plate and 15CrNi3MoV plate are presented in Table 1. The dimensions of the TA2 plate and 15CrNi3MoV plate were 370 mm × 70 mm × 75 mm and 370 mm × 70 mm × 8 mm, respectively. To relieve the internal stress of the explosively welded TA2/15CrNi3MoV clad plates, an annealing treatment was performed at 550 °C for 1.5 h. After annealing treatment, the explosively welded TA2/15CrNi3MoV interfaces were heat-treated in the furnace under an argon atmosphere. The microstructures also show a vortex generated in the wave peak interface, where there are brittle IMCs with electron backscattered diffraction (EBSD) and energy dispersive spectroscopy (EDS), as illustrated in Fig. 1c. A transmission electron microscope (TEM) was used to reveal the microstructure, and the samples for TEM observation were fabricated by focused ion beam (FIB). The width of IMCs was measured at various locations of the wave peak interface by SEM-BSE examination. The IMCs formed at the explosively welded TA2/15CrNi3MoV interface were analyzed by X-ray diffraction (XRD). A series of tensile tests were conducted on AGX-plus at room temperature to evaluate the tensile strength of the explosively welded Ti/steel clad plates. Three samples were prepared from each parameter and subsequently tested at a speed of 1.5 mm/min. All samples were fractured at the TA2/15CrNi3MoV explosive welding interface during tensile tests, as shown in Fig. 1d. Tensile test samples were machined by electric discharge machining (EDM) with dimensions schematized in Fig. 1d. SEM/EDS was conducted to analyze the fracture surfaces of the specimens after the tensile tests.

3 Results and discussion

3.1 Microstructure of the as-annealed TA2/15CrNi3MoV explosive welding interface

Table 1 Chemical compositions of the TA2 plate and 15CrNi3MoV plate (wt%)

| Alloy     | C   | Ti   | Fe  | Si  | Mn  | S   | P   | Cr  | Ni  | Mo  | V  |
|-----------|-----|------|-----|-----|-----|-----|-----|-----|-----|-----|----|
| TA2       | <0.01 Bal | <0.01 | –   | –   | –   | –   | –   | 1.02 | 2.9 | 0.21 | 0.95 |
| 15CrNi3MoV | 0.15 | 0.002 | Bal | 0.26 | 0.43 | 0.002 | 0.005 | 2.9 | 0.21 | 0.95 |

Figure 2 shows the SEM-BSE images of the as-annealed TA2/15CrNi3MoV explosive welding interface. A typical wave structure was clearly observed at the explosively welded TA2/15CrNi3MoV interface, as shown in Fig. 2a. Investigation of the microstructures also shows a vortex structure at the interface, where there are brittle IMCs and microcracks. The generation of IMCs at the explosively welded Ti/steel interface can be related to the intense impact during the explosive welding process. The intense collision induced by explosive welding will have caused rapid and severe plastic deformation in the interface region, which generates heat at the interface [23–25]. A previous study

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[16] reported that the explosively welded interface underwent fast heating and cooling and formed a melted zone at the interface. Chu et al. [23] revealed that the melted zone at an explosively welded Ti/steel interface mainly consisted of FeTi. In addition, during the rapid cooling process, the mismatch in mechanical properties (e.g., heat transfer and thermal expansion) between the base plate and the flyer plate generates microcracks at the interface. Previous studies indicated that the microcracks were restricted within the IMCs and did not expand into the base materials [17]. The elemental distribution maps of the vortex region are shown in Fig. 2c–d. Fe-Ti intermetallic compounds were detected in the vortex. The generation of Fe-Ti IMCs can be related to the heating generation and accumulation caused by the severe plastic deformation during the explosive welding process.

Next, we determined the microstructure distribution between TA2 and 15CrNi3MoV matrix based on the TEM observation. Figure 3a–b shows the fabrication procedure of the TEM sample. The TEM sample was prepared by FIB in the explosively welded TA2/15CrNi3MoV interface.

![Fig. 2 Back-scattered SEM images of the as-annealed TA2/15CrNi3MoV explosive welding interface: a wave morphology of the interface; b a vortex of the wave interface; c distribution of Ti; d distribution of Fe]

![Fig. 3 TEM results of the as-annealed TA2/15CrNi3MoV explosive welding interface: a TEM sample cut from the TA2/15CrNi3MoV explosive welding interface and b prepared by FIB, c HAADF images, d–e HRTEM and FFT images at the corresponding selected area]
Figure 3c shows the HAADF images of the wave peak interface region. At the interface region, there were two reaction layers: 2 and 3. Figure 3d–e shows high-resolution TEM (HRTEM) images of layer 2 and layer 3. Analysis of HRTEM images by fast Fourier transform (FFT) shows that layers 2 and 3 were TiC and FeTi and approximately 0.095 μm and 0.118 μm thick, respectively. The formation of FeTi and TiC was mainly induced by explosive welding and annealing processes, respectively [23, 26], which will be analyzed in detail later in this paper. Thus, for the as-annealed sample, the phases from the 15CrNi3MoV side to the TA2 side were α-Fe, TiC, FeTi, and α-Ti.

Figure 4 shows the EBSD results of the wave peak region of the explosively welded TA2/15CrNi3MoV interface. The image quality (IQ) maps and inverse pole figure (IPF) images demonstrate two distinct grains on the steel side: equiaxed grains and elongated grains. The elongated Fe grains were mainly distributed on the steel side away from the explosive welding interface. In contrast, the equiaxed Fe grains were primarily distributed on the steel side near the interface. The elongated Fe grains were mainly caused by the severe plastic deformation during the explosive welding [27]. Simultaneously, heat was generated and accumulated at the Ti/steel interface due to rapid and severe plastic deformation [16, 23]. This deformation heating caused the recovery and recrystallization at the explosive welding interface; then, the equiaxed Fe grains were generated, as shown in Fig. 4. In addition, the annealing treatment facilitated the recovery and recrystallization. Figure 4c, f shows the kernel average misorientation (KAM) maps of the interface, which reflects the dislocation density/local strains [3]. Many local strains were distributed at the elongated Fe grains, while few local strains were distributed at the equiaxed Fe grains. The elongated grains experienced intense plastic deformation and contained many local strains. However, at the equiaxed Fe grains, the local strains sharply decreased due to the recovery and recrystallization.

3.2 Microstructure evolution of the explosively welded TA2/15CrNi3MoV interface after heat treatment

Figure 5 shows the back-scattered SEM images of the explosively welded TA2/15CrNi3MoV interface after different heat treatment processes. The microstructure of the TA2/15CrNi3MoV interface significantly changed with increasing heating temperature. For the samples heat-treated at 700 °C, a gray layer was formed at the interface. This reaction layer, which can be determined by previous works, was β-Ti [10, 20]. According to previous studies [10, 20], β-Ti stabilizer elements (e.g., Fe, Cr, and Ni) diffused from the steel side to the titanium side during the heat treatment,

Fig. 4 EBSD analysis of the as-annealed Ti/steel interface: a, d IQ maps; b, e IPF maps; c, f KAM maps
which decreased the phase transformation temperature of α-Ti to β-Ti. Hence, the β-Ti layer was produced at the interface after the heat treatment. When the temperature of the heat treatment was increased to 800 °C, a continuous β-Ti layer and Widmanstätten α-β structure were generated at the TA2 side, as shown in Fig. 5c. On the TA2 side away from the explosively welded interface, the amount of β-Ti stabilizer was not sufficient to retain the β-Ti to the natural temperature; hence, the Widmanstätten α-β structure was generated. When the heat treatment temperature continued to
increase, the β-Ti layer at the interface thickened. When the samples were heat-treated at 1000 °C, the SEM-BSE images clearly indicate that the Ti/steel explosive welding interface region consisted of several layers, as shown in Fig. 5f. The generation of these layers will be analyzed in detail later in this paper.

Figure 6 shows the SEM-BSE images of the explosively welded TA2/15CrNi3MoV interfaces at higher magnification. For the samples heat-treated at 700 °C, a continuous black reaction layer was generated at the interface. According to the TEM results (Fig. 3) and previous studies [26, 28], this black layer was TiC. TiC was a diffusion barrier, which hampered the diffusion of Fe atoms from the steel side to the titanium side [28–30]. Hence, the Fe-Ti IMCs (i.e., Fe2Ti and FeTi) and β-Ti layer slowly increased when the heat treatment temperature was lower than 700 °C. When the heat treatment temperature was 800 °C, the TiC layer was broken and existed in a semi-continuous state, which indicates that the Fe and Ti atom diffusion barrier disappeared; then, the Fe-Ti IMCs sharply increased to 1.2 μm, as shown in Fig. 6c. For the samples heat-treated at 1000 °C, Fe-Ti IMCs were generated at the interface, which contained fine blocky TiC (see Fig. 6f). Under this condition, the explosively welded Ti/steel interface region can be divided into several different layers. The chemical compositions of the different layers are presented in Table 2. Layer I mainly contained Fe atoms (95.99 at%), which indicates that this layer was α-Fe. Layer II in Fig. 6 was composed of Fe (65.93 at%), Ti (30.87 at%), and C (3.20 at%), and layer III was composed of Fe (43.44 at%), Ti (51.98 at%), and C (4.58 at%). Both layers II and III contained large amounts of Ti and Fe atoms, which indicates that Fe-Ti IMCs existed on these layers and was the primary phase. Layer IV and layer V mainly contained Ti atoms, while the Fe atoms in layer IV (10.21 at%) were more abundant than those in layer V (4.18 at%). Combined with the SEM-BSE images of the interface, layer IV and layer V were β-Ti and β-Ti + α-Ti, respectively. Hence, the explosively welded TA2/15CrNi3MoV interface from the 15CrNi3MoV side to the TA2 side can be divided into five layers: α-Fe, Fe2Ti, FeTi, β-Ti, and α-Ti + β-Ti, as shown in Table 2.

The element distribution across the TA2/15CrNi3MoV cladding interface was used to predict the phases that form at the interface, as shown in Fig. 7. The line profiles of the Fe, Ti, and C element distribution graph present several platforms at the interface and prove the formation of IMCs. For the as-annealed sample, the line profile of Fe and Ti elements has a significant gradient, which indicates that the diffusion of Fe and Ti elements has a significant gradient, which indicates that the diffusion of Fe and Ti elements was limited. For the samples heat-treated at 700 °C, enrichment was observed at the interface. This result confirms that the TiC layer was generated at the TA2/15CrNi3MoV interface. In contrast, this phenomenon was not observed in the processing temperature range of 800–1000 °C, which may be attributed to the dissolving and breaking up of the TiC layer. After the TiC layer broke up, the diffusion distance of Fe atoms in titanium significantly increased [28–30]. For the samples heat-treated at 1000 °C, the Fe and Ti element distribution presented two platforms, which indicates that Fe-Ti IMCs were generated at the interface.

![Element distributions across 15CrNi3MoV to TA2: a as-annealed; b 700 °C; c 800 °C; d 850 °C; e 900 °C; f 1000 °C](image-url)
The XRD patterns of the explosively welded TA2/15CrNi3MoV interface at different temperatures are shown in Fig. 8. The change in diffraction peak intensity reflects the change in the corresponding phase content [3]. The as-annealed specimens show typical diffraction peaks of α-Fe and α-Ti. After the heat treatment, new IMCs (e.g., FeTi, Fe₂Ti, and TiC) were detected. For the samples heat-treated at 700 °C, the diffraction peak intensities of Fe-Ti IMCs were small. In contrast, for the samples heat-treated at 800 °C and 1000 °C, FeTi and Fe₂Ti were the main IMCs at the explosively welded TA2/15CrNi3MoV interface. In addition, the diffraction peak intensities of FeTi and Fe₂Ti increased with temperature, which indicates that the contents of FeTi and Fe₂Ti increased with temperature.

To determine the microstructure evolution of the explosively welded TA2/15CrNi3MoV interface after the subsequent heat treatment process, the standard free Gibbs energy (ΔGθ) changes with the temperature of the IMCs (i.e., TiC, FeTi, and Fe₂Ti) should be considered [28, 31], as shown in Fig. 9. The TiC formation had the lowest ΔGθ value in the temperature range of 200–1400 °C, which implies that TiC is preferentially formed at the interface compared to FeTi and Fe₂Ti. After the continuous TiC layer was generated, the formation of Fe-Ti IMCs (i.e., FeTi and Fe₂Ti) was hindered because the continuous TiC layer hampered the diffusion of Ti and Fe atoms across the TA2/15CrNi3MoV interface [29, 30]. Hence, the generation of IMCs (i.e., TiC, FeTi, and Fe₂Ti) at the Ti/steel interface can be divided into two stages.

For the as-annealed sample, two reaction layers (i.e., FeTi and TiC) were generated at the TA2/15CrNi3MoV interface (see Fig. 3c). During explosive welding, the intense plastic deformation of the welded materials caused substantial amounts of heat at the TA2/15CrNi3MoV interface, which formed local melting zones at the TA2/15CrNi3MoV interface region, where FeTi was observed [23]. During the annealing process, the C element diffused from the steel side toward the Ti side and finally formed TiC [26]. Hence, the formation of FeTi and TiC at the TA2/15CrNi3MoV interface was mainly induced by explosive welding and stress-relief annealing treatment processes, respectively.

The continuous TiC layer hampered the diffusion of constituent elements (e.g., Ti and Fe) across the bonding interface, which hindered the formation of Fe-Ti IMCs at the TA2/15CrNi3MoV interface [29, 30]. For the TA2/15CrNi3MoV interface heat-treated at 700 °C, continuous TiC hampered the formation of FeTi and Fe₂Ti, as shown in Fig. 6b. For the TA2/15CrNi3MoV interface heat-treated in the temperature range of 800–1000 °C, the continuous TiC layer broke up, so the barrier of the formation of FeTi and Fe₂Ti at the interface disappeared, as shown in Fig. 6c–f. Thus, the FeTi and Fe₂Ti layers significantly thickened with temperature in the temperature range of 800–1000 °C. According to previous studies, the Fe-Ti IMCs...
were very hard and brittle, which deteriorated the combination of the explosively welded TA2/15CrNi3MoV interface. Furthermore, the increase in thickness of the β-Ti layer confirms the accelerated diffusion of Fe atoms after the TiC layer broke up.

3.3 Mechanical properties and fracture analysis

The tensile strengths of the explosively TA2/15CrNi3MoV clad plates at different heat treatment temperatures are presented in Fig. 10. The tensile strength decreased
with increasing temperature. The tensile strength of the as-annealed samples was 385.3 MPa, which was the maximum tensile strength of the explosively welded TA2/15CrNi3MoV interface. For the TA2/15CrNi3MoV interface heat-treated at 700 °C, the tensile strength slowly decreased to 329.2 MPa, which was 14.56% lower than that of the as-annealed samples. However, when the temperature of the heat treatment was 800 °C, the tensile strength sharply decreased to 247.3 MPa, which was 35.82% lower than that of the as-annealed samples. This result can be attributed to the substantial number of IMCs (i.e., TiC, FeTi, and Fe2Ti) at the interface, which deteriorates the combination of the TA2/15CrNi3MoV interface. During the tensile tests, the stress concentration generated at the TA2/15CrNi3MoV interface was due to the deformation mismatch between IMCs and base materials, which eventually caused the explosively welded Ti/steel interface to fracture under a lower load [21]. For samples heat-treated at 850 °C and 900 °C, the tensile strength of the explosively welded TA2/15CrNi3MoV interface sharply decreased to 232.6 MPa and 222.9 MPa, respectively, due to the increase in IMCs. When the temperature of the heat

Fig. 11  a Pulsed TIG welding thermal cycling curves and b mechanical properties of the TA2/15CrNi3MoV clad plates under different conditions [13]

Fig. 12 Fracture surface of TA2/15CrNi3MoV clad plates under different conditions: a as-annealed; b 700 °C; c 800 °C; d 850 °C; e 900 °C; f 1000 °C
treatment was 1000 °C, the thickness of the Fe-Ti IMCs at the bonding interface increased to 4.23 μm, so the tensile strength of the explosively welded TA2/15CrNi3MoV interface sharply decreased to 163.6 MPa.

According to our previous study [13], the tensile strengths of the explosively welded TA2/15CrNi3MoV clad plates at different pulsed TIG welding thermal cycles are shown in Fig. 11. With the increase in peak temperature (Tm) of the pulsed TIG welding thermal cycles, the tensile strength gradually decreased. For the clad plates under Tm = 873 °C, the tensile strength slowly decreased to 346.5 MPa. Under this condition, continuous TiC hampered the diffusion of Ti and Fe atoms across the bonding interface, which could prevent the FeTi and Fe2Ti generation. When Tm = 987 °C, the continuous TiC layer broke up, which resulted in a sharp drop in tensile strength to 279.7 MPa. When Tm = 1081 °C, the TA2/15CrNi3MoV interface was remelted, and a significant quantity of FeTi and Fe2Ti was generated, which deteriorated the combination of the clad plates. As a result, the tensile strength of the clad plates sharply decreased to 120.5 MPa.

Figure 12 shows the fracture surface of the explosively welded TA2/15CrNi3MoV clad plates. All the tensile test samples fractured at the Ti/steel interface. There were bulk fragments on the fracture surface for the as-annealed samples, as shown in Fig. 12a. For specimens heat-treated in the range of 700–1000 °C, the failure mechanism of the explosively welded TA2/15CrNi3MoV interface was brittle fracture. Chemical compositions measured by EDS reveal that the fracture surfaces mainly contained FeTi and Fe2Ti.

4 Conclusions

TA2/15CrNi3MoV explosive welding clad plates were heat-treated in the temperature range 700–1000 °C for 30 min. The microstructure morphology of the interface and tensile strength were investigated, and the conclusions are listed below:

1. The heat treatment process accelerated the interdiffusion of constituent elements, which facilitated the formation of FeTi, Fe2Ti, and TiC. For the Ti/steel interface heat-treated at 700 °C, the continuous TiC layer hampered the formation of Fe-Ti IMCs. For the explosively welded TA2/15CrNi3MoV interface heated in the temperature range of 800–1000 °C, continuous TiC broke up, and the thickness of Fe-Ti IMCs sharply increased to 4.2 μm.

2. The tensile strength of the explosively welded TA2/15CrNi3MoV interface decreased with higher heat treatment temperatures. At 700 °C, the tensile strength slowly decreased to 329.2 MPa. At 800 °C, the tensile strength sharply dropped to 247.3 MPa due to the increase in number of brittle Fe-Ti IMCs at the TA2/15CrNi3MoV interface.

In a real service environment, Ti/steel clad plates may be exposed to high temperatures under subsequent welding processes (e.g., LBW and EBW). During such a process, the microstructure evolution and mechanical behavior of the Ti/steel clad plates are more complex. Future works will focus on the plates’ behaviors in this process based on the basic mechanism in the current research.

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Code availability Not applicable.

Declarations

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Consent for publication Not applicable.

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