Forming mechanism and mechanical property of pulsed laser welded Ti alloy and stainless steel joint using copper as interlayer

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In this work, Cu was used as an interlayer to prevent the formation of these brittle Ti-Fe intermetallics when joining Ti alloy to stainless steel (SS). Microstructures of the joint were analyzed by optical microscopy, scanning electron microscopy, transmission electron microscopy and X-ray diffraction. The tensile strengths of the joint were also measured. The laser was focused on the SS side of the joint, which joined the SS and Cu by fusion welding. At SS-Cu interface, a weld zone was formed due to dilution of the Cu and mixing with the SS. At the Ti alloy-Cu interface, a eutectic reaction was responsible for joining. The presence of unmelted Cu interlayer ensured that crack free welds were obtained and no brittle Ti-Fe intermetallics were observed. Tensile strength of the joint can reach to 320 MPa.

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1. Introduction

Recently, aerospace and nuclear industries have a strong demand of dissimilar Titanium (Ti) alloy to stainless steel (SS) for welded joint [1]. Partial replacement of steel components by Ti alloy will become an important approach to reduce mass of spacecrafts [2]. However, in spite of such vital applications, success in producing reliable and strong Ti-SS joints has been limited, primarily due to the lack of metallurgical compatibility that leads to the formation of brittle intermetallics between these materials [3,4]. Direct heat fusion welding of Ti alloy and stainless steel can result in the formation of a variety of intermetallics such as TiFe, TiFe2, and so on. Hardness of weld metal was in the range of HV 740–1324 [5,6]. These highly brittle intermetallics impair the mechanical properties of the Ti alloy-SS joints [7,8]. Furthermore, Ti alloy and stainless steel have wide differences in their physical properties. In particular is the mismatch of their coefficients of thermal expansion with Ti alloy and stainless steel having different coefficient of thermal expansion, respectively, which results in large residual stress in the joints after welding [9]. Unfortunately, the traditional fusion welding has not yet been technically capable of joining Ti alloy with stainless steel because of a metallurgical incompatibility between them, highly brittle Ti-Fe intermetallics cause spontaneous cracking in conventional fusion-welded joints [10]. So far, less literature about successful direct fusion welding of these two alloys has been reported. The direct joining of this dissimilar couple can be performed only by solid state methods such as diffusion bonding [11], friction [12] and explosive [13] welding. Bulent Kurt et al. [11] have studied the diffusion bonding of Ti alloy to SS with respect to bonding temperature concentrating on the shear
strength. The highest shear strength (187 MPa) was obtained for the bonding temperature of 980 °C. Lee and Jung [12] have attempted friction welding of Ti-SS321 and they were able to achieve a joint strength of 400 MPa by altering the microstructure by varying the friction force. Mousavi and Sartangi [13] have arrived at a suitable parametric window both analytically and experimentally for explosive welding of CP Ti-SS304. They have concluded that at low loads formation of intermetallics could be totally avoided. However, the service conditions may make particular processes unsuitable. Taking high temperature applications for example, brazing cannot be candidate. Furthermore, the required joint geometry can make friction welding and explosive welding difficult to apply.

The general method to deal with this problem is using intermediate layer between Ti alloy and stainless steel to reduce the formation of brittle intermetallics that changes character of the interaction in the melted zone and leads to formation other phases than Ti-Fe-rich intermetallics. The most widely used interlayer are Cu and their alloys since Cu does not produce brittle intermetallics with iron, chromium, nickel and carbon. Despite the formation of numerous intermetallic phases in the Cu-Ti system, such as CuTi2 and Cu2Ti, these Cu-Ti intermetallics were found to be less brittle than the Ti-Fe intermetallics. Therefore, Cu was often used as interlayer material for joining Ti alloys to steels due to the compensation of local phase brittleness by its ductility. I.P. Oliveira et al. highlighting the additive manufacturing technologies based on melting and solidification have considerable similarities with fusion-based welding technologies, either by electric arc or high-power beams [14]. The high power beam methods such as laser and electron beam welding are more suitable for joining dissimilar alloys because they provide very local heat supply, rapid heating/cooling gradients and a perfect precision in the weld realization [15]. Both methods were successfully applied for welding of Ti alloy and steel. The first attempt to perform fusion welding of Ti alloy to steel with Cu interlayer was made by Wang et al. [6]. The tensile strength of the weld was equal to 224 MPa and the brittle fracture within intermetallic layer occurred near the Ti/melted zone interface. Yan Zhang et al. [16] found that the tensile strength of pulsed laser welded dissimilar joint of TC4 Ti alloy to 301L stainless steel using Cu as an interlayer was up to 340 MPa. According to their studies, using Cu as interlayer can suppress an interaction between Ti and Fe, thus amount of brittle Ti-Fe intermetallics was reduced in the weld and greatly reduce the brittleness of the weld. Moreover, investigations of adding other interlayers such as Mg, Ni and Co during fusion welding can also partly prevent the interaction between Ti and Fe. However, as long as the interlayer is completely melted, Ti and Fe elements will mix and react in the fusion pool [17].

I. Tomashchuk found that the shift of heat source to Cu-steel interface allowed to reduce the melting of the Ti alloy and subsequently reduced the amount of Ti-Fe. But total metallurgical isolation of welded materials by Cu interlayer was not achieved [18]. Binggang Zhang et al. [19] obtained a highly qualified electron beam welding joint for Titanium-stainless steel clad plate with Cu interlayer. In order to avoid the fusion of Ti alloy during welding process, the electron beam was acted on the stainless steel plate. Ti alloy and Cu sheet were joined by the contact reaction of Ti and Cu. According to Binggang Zhang’s report, a intermetallic layer was formed by contact reactive of Ti alloy and Cu sheet. Fracture occurred in the intermetallic layer and tensile strength can reach to 300 MPa. In view of the above analysis, pulsed laser welding of Ti alloy and SS using Cu as interlayer was proposed in this paper. The welding process was set up to ensure that the interlayer was partly melted. The melted Cu interlayer and SS formed the welding pool. Meanwhile, a reaction layer was formed at the interface between unmelted Cu interlayer and Ti alloy. In this way, unmelted Cu interlayer was left in the joint after welding with the main objective to avoid mixing of Ti and Fe so that Ti-Fe intermetallics was expected to be eliminated in the joint. However, difficulties exist, such as how to control the melting amount of Cu interlayer and formation of reaction layer by optimizing the process parameters. The amount of melted Cu interlayer and formation of reaction layer in the Ti alloy-SS dissimilar joint can be controlled by properly selecting the processing parameters and the location of laser beam spot. The formation mechanism of the reaction layer between the Ti alloy and Cu was discussed on the basis of Ti-Cu phase diagram and the theory of deep-penetration laser welding.

### Table 1 - Main chemical compositions of 301L stainless steel (at.%).

<table>
<thead>
<tr>
<th>Element</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Cr</th>
<th>Ni</th>
<th>N</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>Value</td>
<td>0.00</td>
<td>0.20</td>
<td>0.045</td>
<td>0.03</td>
<td>16.00-18.00</td>
<td>6.00-8.00</td>
<td>&lt;0.20</td>
<td>Bal</td>
</tr>
</tbody>
</table>

### Table 2 - Main chemical compositions of TC4 Titanium alloy (at.%).

<table>
<thead>
<tr>
<th>Element</th>
<th>Al</th>
<th>V</th>
<th>Fe</th>
<th>C</th>
<th>N</th>
<th>H</th>
<th>O</th>
<th>Ti</th>
</tr>
</thead>
<tbody>
<tr>
<td>Value</td>
<td>6.05</td>
<td>4.02</td>
<td>0.14</td>
<td>0.02</td>
<td>0.02</td>
<td>0.006</td>
<td>0.12</td>
<td>Bal</td>
</tr>
</tbody>
</table>

2. Experimental procedure

#### 2.1. Materials

The materials used are 0.8 mm thick plates of SUS 301L stainless steel and TC4 Ti alloy. The specimens for direct welding experiments were machined into 50 mm × 40 mm × 0.8 mm plates. Their chemical compositions and physical properties are given in Tables 1, 2 and 3. 0.4 mm thick Cu sheet was adopted as interlayer and placed in the contact faces. Before welding, the specimens were mechanically and chemically cleaned. The gap between the edges of the two workpieces is very important to prevent porosity formation. Therefore, before the welding process the edges of the workpieces have been made smooth as much as possible using milling cutter. The workpieces are clamped each other tightly in order to get the minimum gap formation between the edges [22,23].

#### 2.2. Welding method

Nd:YAG pulsed laser with average power of 1.05 kW, spot diameter of 0.1 mm and wavelength of 1.064 μm. A schematic diagram of the welding procedure is shown in Fig. 1a. In order to ensure that Cu interlayer was not completely melted, the laser can be bias to the SS with the distance of 0.2 mm away.
Specimens

from the Cu-SS interface. The welding process parameters were: laser beam current of 75 A, pulse width of 13 ms, defocusing distance of +2 mm, pulse frequency of 4 Hz, welding speed of 150 mm/min.

2.3. Characterization methods

Specimens for microstructural characterization were prepared metallographically and then etched using a reagent containing 2 ml HNO₃ and 6 ml HF. The nature and the localization of intermetallic phases in the welds have been studied by SEM, TEM, EDS, XRD and microhardness measurements. Vickers microhardness tests have been carried out with load time 10 s and the load of 200 g. The tensile strength of the joints was evaluated at room temperature in tensile testing machine (MTS Insight 10 kN) at a cross head speed of 0.2 mm/min. WRN-191 K sheathed thermocouple (measuring range –250° 350 °C) as the temperature sensor.

3. Results and discussion

3.1. Macro-characteristics

Fig. 2 presents the cross section of the joint. The joint can be divided into three parts: the reaction layer formed at the Ti alloy-Cu interface, unmelted Cu and the weld zone formed at the Cu-SS interface. If the laser beam was fixed on the side of the SS plate with some distance from Cu-SS interface, the weld pool with keyhole will be generated inside the SS plate. The laser beam that was directed into the keyhole was reflected and absorbed repeatedly according to the laser penetration welding mechanism, leading to the increment of the heating output and width of the welding pool [24]. Thus, the edge of the Cu as the keyhole boundary melted sharply due to the high absorption of the laser inside the keyhole. It can be speculated that during laser beam oscillation, the laser beam dwells on both sides of the weld zone for a certain amount of time. Although only a few seconds, but not only promotes heat dissipation but also melts Cu interlayer locally. In addition, the width of surplus Cu was 320 μm and almost 80% of initial sheet thickness, which indicated the major melting had occurred on the SS side, while limited melting took place on the Cu side. For melted Cu interlayer, rapid cooling of the melt resulted in formation of the immiscible flows and droplets between liquid Cu and steel that enhanced the barrier function of the Cu interlayer [25]. The unmelted part of Cu interlayer due to its high thermal conductivity and high absorptivity for pulsed laser (wavelength 1.064 μm) of absorbing a significant amount of heat from the welding pool and transferring it to the Ti alloy side [22]. The temperature was high enough to promote atomic interdiffusion. Thus, eutectic reaction occurred and eutectic liquid was produced in Ti alloy-Cu interface. When the laser beam was offset toward SS, the unmelted Cu acted as a barrier to mixing of the two base materials with the liquid-states mixing between Ti alloy and SS was suppressed fully, thus amount of brittle Ti-Fe intermetallics was reduced in the joint and greatly reduce the brittleness of the joint. It should be noted that precise control of the laser spot position is crucial to obtain a sound joint. If the laser spot is far away from the Cu-SS interface, eutectic reaction of Ti and Cu at the Ti alloy-Cu interface cannot take place. If the laser spot is draw near to the Cu-SS interface, the Ti alloy is beginning to melt in the joint, thereby amount of brittle Ti-Fe intermetallics is greatly increased in the joint and cannot be realized the effective combination between dissimilar materials of Ti alloy and SS.

3.2. Microstructure analysis

Optical image of weld zone is shown in Fig. 3a and no defects were observed in it. SEM image of weld zone is shown in Fig. 3b, the cellular dendritic structure was well evident in weld zone. Based on the EDS analysis results in Table 4, Cu and Fe, as

### Table 3 – Physical properties of TC4 Titanium alloy and 301L stainless steel.

<table>
<thead>
<tr>
<th>Material</th>
<th>Melting point/°C</th>
<th>Tensile strength/MPa</th>
<th>Specific heat capacity/ J kg⁻¹ K⁻¹</th>
<th>Thermal conductivity/ W m⁻¹ K⁻¹</th>
<th>Linear expansion coefficient 10⁻⁶ K⁻¹</th>
</tr>
</thead>
<tbody>
<tr>
<td>TC4</td>
<td>1650</td>
<td>895</td>
<td>536</td>
<td>6.4</td>
<td>8.7</td>
</tr>
<tr>
<td>SUS301L</td>
<td>1450</td>
<td>550</td>
<td>500</td>
<td>16.3</td>
<td>16.9</td>
</tr>
</tbody>
</table>

![Fig. 1 – Schematic diagram of the welding process.](image1)

![Fig. 2 – Optical image of the joint obtained by one pass welding.](image2)
Fig. 3 – Microstructures in the weld zone of the joint: (a) optical image of weld zone centerline; (b) SEM image of weld zone centerline; (c) optical image of fusion line near Cu side; (d) SEM image of fusion line near SS side.

### Table 4 – EDS analysis of different zones in the joint.

<table>
<thead>
<tr>
<th>Region</th>
<th>Composition (at.%)</th>
<th>Potential phases</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>V</td>
<td>Al</td>
</tr>
<tr>
<td>A</td>
<td>8.26</td>
<td></td>
</tr>
<tr>
<td>B</td>
<td>5.82</td>
<td></td>
</tr>
<tr>
<td>C</td>
<td>8.02</td>
<td></td>
</tr>
<tr>
<td>D</td>
<td>0.88</td>
<td>1.12</td>
</tr>
<tr>
<td>E</td>
<td>1.00</td>
<td>1.01</td>
</tr>
<tr>
<td>F</td>
<td>0.70</td>
<td>0.51</td>
</tr>
<tr>
<td>G</td>
<td>0.08</td>
<td></td>
</tr>
</tbody>
</table>

Two main elements, were detected in the weld. However, Ti element was not found, which indicated the absence of Ti-Fe intermetallics. As the Cu-Fe system [23] does not contain intermetallic phases, it is easy to create mechanically stable Cu-steel joints. As shown in Table 4, the average compositions of weld zone were approximately 8 at. % Ni, 7 at. % Cu, 68 at. % Fe and 18 at. % Cr, which were close to that of stainless steel. Therefore, the main microstructure of weld zone was γ-Fe phase. Moreover, some micro-cracks were found in the weld zone (Fig. 3b). It happened because of high cooling rates of Cu and high level of the residual stress [26,27]. Fig. 3c shows the microstructures in the weld near the residual Cu interlayer. Obvious fusion line was observed. It can be seen from enlarged photo of Fig. 3c that the columnar grains at the near fusion line grew essentially along the direction perpendicular to the fusion line and proceeded toward the weld centerline. Because the high thermal conductivity of 398 W/(mK) of Cu causes a large temperature gradient at solid-liquid interface of molten pool [28]. It benefits the nucleation of column grains at the interface and the fast growth. The formation of these microstructure characteristics was in accordance with the theory of welding metallurgy [29]. However, coarse grains close to the Ti alloy-Cu interface are integral, indicating that part of Cu does not melt during welding. Because Cu having higher thermal conductivity allows large amount of heat to flow through it, which indicate unmelted Cu interlayer absorbed a significant amount of energy and transferred it to the Ti alloy to create temperature condition for the formation of reaction layer at the Ti alloy-Cu interface. Fig. 3d shows the interfacial morphology between the weld zone and the stainless steel during welding mode. The weld zone-SS interface was smooth and belonged to cellular dendritic morphology.

The optical image of the reaction layer at Ti alloy-Cu interface is shown in Fig. 4a. The reaction layer had average width of approximately 40 μm. In this region, chemical and metallurgical reaction was complex, which can be confirmed by large numbers of reaction products with a variety of morphology. Microstructural examination through SEM studies was carried out in order to determine the structure and distribution of possible phases in the reaction layer. SEM image of the reaction
layer is shown in Fig. 4b. EDS analysis results of the reaction layer is shown in Table 4. It can be seen that the reaction layer consisted of a lamellar structure including two reaction zones marked as zone I and II according to their different morphological characteristics of the included reactive phases, where zone I was close to Ti alloy and zone II was next to the Cu interlayer. The compositions in different positions of zone I and zone II (denoted by letter D-G in Fig. 4c and 4d) were analyzed by SEM-EDS. Based on the Ti-Cu binary phase diagram [30], the results were analyzed and listed in Table 4. According to EDS stoichiometric ratio, region D was defined as Cu$_2$Ti phase, region E was defined as Cu$_2$Ti + Cu$_4$Ti phase, region F was defined as Cu$_4$Ti phase and region G was defined as α-Cu phase. The TEM micrograph of zone I is shown in Fig. 5e, plate-like Cu$_2$Ti phase can be observed. The TEM micrograph of zone II is shown in Fig. 5e, Cu$_4$Ti phase can be observed and Cu$_4$Ti the characteristic flocculent appearance. Thus, the reaction layer of Ti alloy-Cu interface mainly consisted of Cu$_2$Ti, Cu$_2$Ti + Cu$_4$Ti, Cu$_4$Ti and α-Cu solid solution orderly from the Ti alloy to the Cu interlayer. The forming process of reaction layer was concomitant with atomics inter-diffusion between Ti and Cu. In order to confirm the distribution of elements Ti and Cu at the Ti alloy-Cu interface, the SEM–EDS line analysis was carried out, as shown in Fig. 5. The line analysis started from Ti alloy side, passed through the reaction layer and ended in Cu side. As seen, from the Ti alloy to the Cu interlayer, the
content of Cu element increased gradually while the content of Ti decreased gradually as a whole. Since the diffusion ability of Ti into Cu was stronger than that of Cu into Ti [31], larger amount of Ti element diffused into Cu side than that of Cu element diffused into Ti side. To better reflect the formation of the diffusion weld at Cu-Ti alloy interface, the thermal cycling test was performed on the Cu-Ti alloy interface. The structure was provided with opposite grooves on the Ti alloy side faces, and the interior of a sealing groove formed by the groove was provided with a thermocouple. Fig. 6b present the thermal cycle curve obtained from Ti alloy-Cu interface during welding. It is seen that the maximum experimental temperature was 892°C, which exceeded eutectic temperature of Ti and Cu, but was below the melting point of Cu. This meets the temperature requirement for eutectic reaction.

A schematic of formation process of reaction layer was presented in Fig. 7. At higher temperatures, the thermal energy supplied to the diffusing atoms permitted the atoms to overcome the activation energy barrier and more easily move to new lattice sites [32]. The high temperature of Ti alloy-Cu interface provided the activation energy. Thus, Ti and Cu atoms gathered at Ti alloy-Cu interface, as shown in Fig. 7a. The interdiffusion of Ti and Cu elements occurred until eutectic composition was acquired and liquid phase was formed by the eutectic reaction. At this moment, the dissolution of Ti and Cu into the eutectic liquid occurred under the high concentration gradient, which would change the composition of the liquid, as shown in Fig. 7b. Since the liquid existed for only a short time due to the welding process with rapid heating and cooling, Ti and Cu in the liquid did not have enough time to spread evenly leading to concentration gradient in the liquid. In the subsequent cooling process, the liquid phase with different compositions would experience different reaction. Also because of the pulsed laser welding has faster heating and cooling rate, with further decreasing of the temperature, the degree of supercooling increases in the liquid phase. Thus, nucleation started. Typical reactions were explained as point D and E in zone I. According to EDS analysis results, composition in position D of zone I was of 64.25 at.% of Ti and 33.75 at.% of Cu. So it can be concluded that at position D, Cu2Ti phase was primarily precipitated from the liquid during cooling. The Cu2Ti phase with a high melting point of 890°C, nucleates and grows up at the interface between Ti alloy and Cu interlayer, as shown in Fig. 7c. Then the composition of Cu in the residual liquid increased, which resulted in the eutectic reaction (L → Cu2Ti + Cu4Ti) at temperature 875°C. Cu2Ti + Cu4Ti eutectic grew up along the largest temperature gradient direction in the liquid, as shown in Fig. 4c and Fig. 7d. It was needed to notice that Cu2Ti was a kind of metastable phase existing in the temperature range of 890–870°C. But in this paper it was found in the reaction layer at room temperature because it was retained for the rapid cooling rate during laser beam welding. Typical reactions in point F and G of zone II were explained as follows. Firstly, α-Cu appears from initial liquid first (L→α-Cu). Because the highest temperature gradient was produced along the direction perpendicular to the solid/liquid interface [33], α-Cu gains grew rapidly along this direction, which induced formation of the coarse columnar crystal structures, as shown in Fig. 4d and Fig. 7e. Hence, proeutectoid α-Cu was precipitated from liquid phase firstly, which resulted in the peritectic reaction (L+α-Cu→Cu4Ti) occurred and Cu4Ti was formed, as shown in Fig. 7f. After liquid phase solidified completely, the reaction layer was formed. In this case, Ti-Cu intermetallics were formed at Ti alloy side but α-Cu was formed at the Cu side. We can draw the conclusion that the rate of Ti and Cu atoms diffusion was the key factor of controlling forming process of reaction layer.

![Fig. 5](image_url) **Fig. 5** – Above: (reaction layer from Fig. 4) EDS line scans across the Ti alloy-Cu interface.

![Fig. 6](image_url) **Fig. 6** – Temperature measurement test of Ti alloy-Cu interface: (a) thermocouple distribution; (b) thermal cycle curve of test points.
Fig. 7 – Growth process of the reaction layer during welding: (a) dissolution and diffusion of Ti and Cu in the interface; (b) formation of eutectic liquid in the interface; (c) nucleation and growth of Cu₂Ti; (d) nucleation and growth of Cu₂Ti + Cu₄Ti eutectic; (e) nucleation and growth of α-Cu and solidification of Cu side; (d) formation of Cu₄Ti in the liquid/solid interface.

Fig. 8 – Vickers microhardness measurements at semi-height of the joint (zero point situated in the center of the joint).

3.3. Microhardness tests

As shown in Fig. 8, the microhardness distribution in the joint was non-uniform. TC4 Ti alloy has similar hardness to 301L stainless steel and the hardness of Cu interlayer was low because it was simple metals. But overall, the microhardness distribution in the weld zone was relatively uniform and the average value was 260 HV which was lower than that of stain-

less steel. Hardness distribution further reveals that Cu was uniformly dispersed in the weld zone. Therefore, hardness variation also corroborates the microstructural investigation regarding the extent of Cu mixing in the weld zone. Furthermore, there existed a lower hardness of melting region on the Cu side. That might be attributed to its coarse grained structure as revealed in the microstructural analysis. The hardness of reaction layer was higher than those of Ti alloy. This was mainly resulted from Ti-Cu intermetallics.

3.4. Tensile tests and fractography

The maximum tensile strength and elongation of the joint was 320 MPa and 4.5 %, respectively (Fig. 9a). As shown in Fig. 9b, fracture occurred in the reaction layer at Ti alloy-Cu interface. SEM image of the fracture surfaces were presented in Fig. 9c, fracture surfaces were characterized by a number of secondary crack zone. That means the fracture mode of the joint was brittle fracture. As shown in Fig. 9d, XRD analyses of fracture surface detected the existence of Cu₂Ti and Cu₄Ti phases. This confirmed the presence of a large number of Ti-Cu intermetallics at fracture surfaces. It should be noted that there was no Ti-Fe intermetallics in the weld zone. Reaction layer at Ti alloy-Cu interface become the weak zone of the joint, which resulted in the failure in the tensile test.
4. Conclusions

Dissimilar laser welding of Ti alloy to SS was achieved using a Cu interlayer. Unmelted Cu interlayer acted as a barrier to mixing of the two base materials, which prevented the formation of brittle Ti-Fe intermetallics, while ensuring joining at both interfaces. At the Cu-SS interface, a weld zone formed by fusion welding. No intermetallics except for γ-Fe were produced in the weld zone and cellular dendritic structure was observed. At the Ti alloy-Cu interface, a reaction layer formed by eutectic reaction. The reactive layer mainly contained Cu2Ti and Cu4Ti. XRD analysis result confirmed the presence of Cu2Ti and Cu4Ti in the reaction layer. The tensile strength can reach to 320 MPa.

Conflict of interest

We declare that we do not have any commercial or associative interest that represents a conflict of interest in connection with the work submitted

REFERENCES


