

## Review

# Advancements in Interface Layer Design for Ti and Steel Welding: A Review

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## Abstract

The connection between dissimilar materials, Ti and steel, has been a focal point for global scholars. Establishing a high-strength bond between Ti alloy and stainless steel offers the potential to harness their respective advantages and reduce production costs, holding significant applications and far-reaching implications. Currently, non-transition welding methods for Ti/steel, primarily diffusion welding and vacuum brazing, have been pivotal in the early stages of development. Despite their simplicity and convenience, effectively avoiding the formation of brittle Ti–iron compounds in the weld seam, these methods face challenges such as unwelded defects, posing a risk to the reliability of welded structures under service conditions. This limitation restricts their application in products requiring high reliability. The evolving transition welding process, progressing from a single metal interface layer to a multi-metal interface layer, addresses some of the shortcomings of traditional Ti and steel connections, offering promising application prospects. This article delves into the core issue of selecting interface-layer elements and welding methods. Through an analysis of the metallurgical properties of transition metals in conjunction with Ti and steel, the study investigates the impact of single- or bimetallic elements, such as Cu, V, Nb, and Ni, on preparing interface-layer transition metals. A comprehensive review of existing research on Ti and steel welding is presented, with an emphasis on the metallurgical characteristics of their connection. The influence of element selection and welding processes on the metallurgical features and relevant mechanical properties of the weld metal is systematically analyzed and summarized.

**Keywords:** titanium alloy; stainless steel; dissimilar materials; interface layer; weld microstructure; mechanical properties



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## 1. Introduction

With the continuous advancement of scientific knowledge in the 21st century and the increasing globalization of economies, there is a growing demand in the field of engineering for comprehensive mechanical properties of various metal materials and an understanding of their performance in service environments. In some specialized engineering applications, traditional single-metal materials are no longer adequate to meet the performance requirements. To address these demands, researchers have begun combining metals with different properties and characteristics to form metal matrix composite materials, thus giving rise to the emergence of composite materials [1]. These composite materials combine

the chemical and physical properties of different materials in a complementary manner, thereby meeting the performance and economic requirements of various applications. The uniqueness of metal matrix composite materials lies in their ability to inherit the superior mechanical properties of certain materials while also possessing other advantages such as corrosion resistance and cost-effectiveness, making them widely applicable in various sectors of modern manufacturing.

The dissimilar metal joints of steel and Ti offer significant potential in fields like aerospace, shipbuilding, metallurgical engineering, and biomedical applications due to their high performance and economic advantages [2–5]. Ti is known for its high strength-to-weight ratio, excellent corrosion resistance, and exceptional fracture toughness, while steel is valued for its low cost, high toughness, corrosion resistance, and ease of assembly. However, the occurrence of metallurgical reactions between Ti and steel at high temperatures, resulting in the dispersion of brittle intermetallic compounds (IMCs) like FeTi and Fe<sub>2</sub>Ti at the Ti–steel interface, leads to stress concentration at the interface and subsequent cracking. This is a critical factor affecting the fabrication of Ti–steel composite structures [6,7]. The binary phase diagram of iron–Ti reveals that Ti and Fe readily undergo metallurgical reactions to form brittle IMCs, namely FeTi and Fe<sub>2</sub>Ti [8]. Ti exists in a body-centered cubic lattice at high temperatures, while it takes on a hexagonally close-packed hexagonal lattice at lower temperatures. The body-centered cubic lattice of  $\alpha$ -Ti has very limited solubility for dissolved iron atoms, approximately less than 0.047 wt.%. The solubility of iron atoms in the face-centered cubic lattice of  $\beta$ -Ti increases gradually with temperature, reaching its maximum value of 24.7 wt.% at the eutectoid point at 1085 °C. When the content of dissolved iron exceeds the solid solubility limit of iron in Ti at this temperature, reactions between Ti and iron occur, resulting in the formation of brittle IMCs, thereby significantly reducing the mechanical performance of the welded joint. Moreover, Ti also undergoes metallurgical reactions with carbon (C) elements in low-alloy steel, producing brittle TiC compounds. TiC exhibits poor ductility and further worsens the joint's properties. Additionally, the chemical reactivity of Ti and Ti at high temperatures becomes more pronounced. Starting at 200 °C, with the continuous increase in the melting pool temperature during welding, Ti atoms react with hydrogen (H), oxygen (O), and nitrogen (N) atoms, producing various oxide products that substantially degrade the mechanical properties of the weld metal. Therefore, the key to achieving a stable Ti–steel connection lies in controlling the formation of brittle Fe–Ti phases in the weld. This can be achieved by judiciously adding transition layers to suppress the formation of brittle phases. Furthermore, selecting suitable welding techniques and implementing effective gas shielding measures are crucial in preventing the oxidation of Ti at high temperatures.

To address these challenges, various welding methods have been extensively practiced for investigating the connection between Ti and stainless steel. These methods primarily encompass brazing [9–14], laser welding [15–22], diffusion welding [23–28], explosive welding [29–32], and friction stir welding [32–40]. Different welding methods, categorized based on the heat sources and heating methods employed, can be classified into two main groups: solid-state welding and fusion welding. Solid-phase welding techniques, such as friction stir welding, diffusion welding, and explosive welding, operate without fully melting the materials, achieving material connection under solid-state conditions. In contrast, fusion welding methods involve the complete melting of materials, including common processes like arc welding, laser welding, and electron beam welding, where welds are formed in a high-temperature melting state. The selection between these categories depends on factors such as material properties, welding conditions, and the desired welding quality, tailored to specific engineering requirements. However, both Ti and iron have melting points exceeding 1500 °C, with a narrow difference of only 122 °C. Con-

sidering the superheating of metals during rapid melting in welding processes, they will simultaneously undergo melting. The analogy to fusion brazing theory in steel–aluminum welding (where one metal melts while the other does not) is not directly applicable to the connection between Ti and stainless steel.

Existing reviews on Ti–steel welding have laid a foundation for subsequent research. For instance, Yefei et al. [41] summarized solid-state welding techniques for Ti–steel joints over the past decade, focusing on process parameters and microstructure evolution. Mo et al. [27] reviewed diffusion bonding between Ti alloys and stainless steels, emphasizing the influence of bonding temperature and time. However, these reviews lack a systematic analysis of the design principles of interface-layer elements (e.g., why Cu, V, Ni are preferred) and their synergistic effects with welding processes. Additionally, prior works rarely compare the mechanical performance of joints with different single/multi-layer interface layers or propose targeted optimization strategies. This review aims to fill these gaps by: (1) elucidating the metallurgical mechanism of interface-layer elements with Ti/Fe; (2) establishing a correlation between interface layer design, welding process, and joint performance; (3) providing a critical assessment of existing research limitations; and (4) outlining future research directions.

This article will explore the mechanisms involving different elemental components and various welding methods in the interface-layer connection between Ti and steel. Through a detailed analysis of the impacts of various elements under diverse welding conditions, as well as the interface formation and performance under different welding methods, the objective is to provide a comprehensive understanding of the efficient connection between Ti and steel.

## 2. Joints Without Interlayer

Joints without an interface layer are typically better suited for solid-state welding techniques, such as friction stir welding, diffusion welding, and explosive welding. These approaches do not involve melting during welding but rather achieve material connection under solid-state conditions. Friction stir welding employs specially designed non-consumable tools to stir and squeeze between two workpieces, facilitating atomic diffusion and a robust connection. Diffusion welding solidifies material connection through atomic diffusion at elevated temperatures. Explosive welding relies on the high pressure and temperature generated by explosive impact. The characteristics of these solid-state welding methods confer advantages in handling high-melting-point metals, alloys, and heat-sensitive materials, while also mitigating defects like grain growth, porosity, and oxidation often associated with melt welding. In their past review, Yefei et al. [41] provided a comprehensive overview of solid-state welding techniques applied to the joining of Ti with steels over the past decade. The primary solid-state welding methods highlighted in their work include Diffusion Bonding, Explosion Welding, and Friction Stir Welding. These techniques find applications in diverse fields such as transportation, power generation, desalination plants, the chemical and petrochemical industry, among others. Diffusion welding stands out as the most commonly employed method, attributed to its numerous advantages.

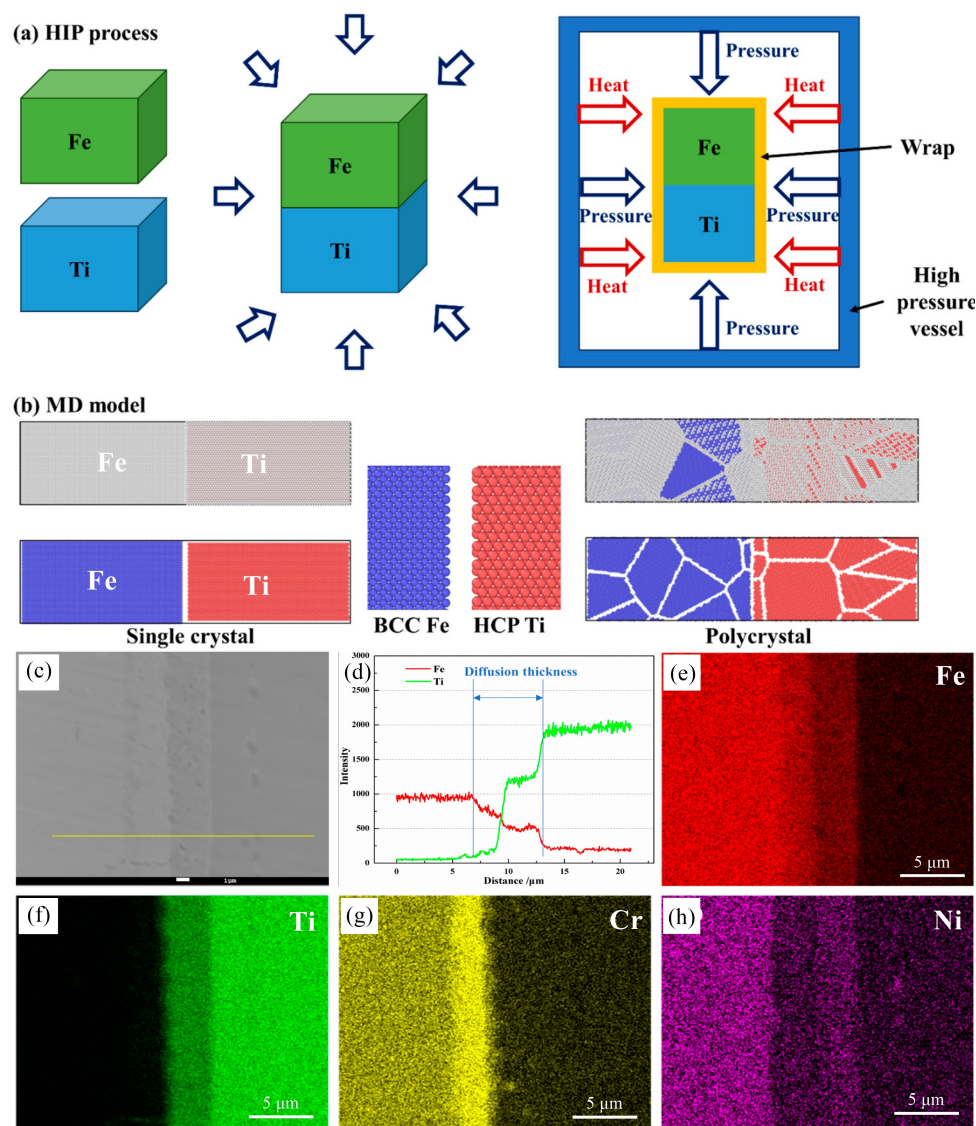
### 2.1. Diffusion Welding

Ghosh and Chatterjee [42] reported on the outcomes of solid-state direct diffusion bonding between commercially pure Ti and austenitic stainless steel. Notably, successful joining of Ti with stainless steels, low alloy steels, and hard-wearing steels was reported through diffusion bonding, with and without the utilization of interlayers such as Ni, Cu, and V.

In order to inhibit the formation of IMCs at the interface, the diffusion bonding between Ti and stainless steel at low temperature was carried out. Velmurugan et al. [43] studied the joint of Ti6Al4V alloy bonded with duplex stainless steel (DSS) by diffusion bonding process in the lower temperature range of 650–800 °C, and applied an axial load of 10 MPa for 30 min in vacuum. At the interface,  $\alpha$ -Fe +  $\lambda$  (solid solution of Fe<sub>2</sub>Ti + Cr<sub>2</sub>Ti),  $\lambda$  + FeTi, and  $\beta$ -Ti were formed; and at a higher bonding temperature, the width of these reaction phases increased. The growth of  $\lambda$  and FeTi phases is attributed to the enhanced atomic diffusion at elevated temperatures, which exacerbates the metallurgical reaction between Ti and Fe. This indicates that low-temperature bonding ( $\leq 700$  °C) is more favorable for reducing brittle IMCs but may compromise joint strength due to incomplete atomic diffusion [43]. Wang et al. [44] conducted thermal diffusion experiments of industrial pure Ti grade 1 steel and 14Cr1MoR steel at different heating temperatures and holding times. The compounds' element diffusion behavior and formation mechanism were observed below or over the  $\beta$  transus temperature. When heated at 850 °C, the reaction diffusion of C and Ti atoms mainly occurs in the contact region and forms some weak binding regions. The transformation of crystal structure leads to the change in diffusion behavior of Fe, C and Ti. The TiC formed at 900 °C is not enough to inhibit the diffusion of Fe to the Ti side. TiFe compounds were formed and dominated at higher temperatures. This suggests that TiC cannot act as an effective diffusion barrier for Fe, which is a key limitation of non-interlayer diffusion bonding. In contrast, the addition of transition elements (e.g., V, Nb) can form stable solid solutions with Ti, thereby blocking Fe diffusion. Gu et al. [45] successfully bonded 316 stainless steel with commercial pure Ti plate by diffusion bonding, as shown in Figure 1. The results showed that the challenge of deviation in diffusion coefficients caused by gaps at the Fe-Ti interface in experiments, this study combines molecular dynamics simulations with diffusion welding experiments, and compares single-crystal with polycrystalline structures as well as analyzes radial distribution functions, to systematically elucidate the effects of temperature, time, and grain boundaries on diffusion behavior at the atomic scale, thereby effectively resolving the key issue of inconsistency between experimental and simulation results [45].

Vigraman et al. [46] utilized EDAX (energy dispersive X-ray analysis) and XRD (X-ray diffraction) to identify compounds within the joints of diffusion-bonded Ti6Al4V and AISI 304L. Besides the presence of Fe<sub>2</sub>Ti, TiNi<sub>2</sub>, Ti<sub>3</sub>Ni<sub>4</sub>, and Fe<sub>2</sub>Ti<sub>4</sub>O, the interface showed the existence of additional phases such as Fe<sub>2</sub>V<sub>3</sub>, Mn<sub>2</sub>Ti, Fe<sub>3</sub>Al<sub>2</sub>Si<sub>4</sub>, Al<sub>6</sub>Ti<sub>19</sub>, Al<sub>4</sub>CrNi<sub>15</sub>, and Ti<sub>2</sub>Si<sub>2</sub>. These phases were confirmed due to the aggregation of V, Mn, Si, and Al. The formation of these phases at the interface led to the depletion of related elements at the grain boundaries, subsequently causing grain growth. The formation of these additional IMCs (e.g., Fe<sub>2</sub>V<sub>3</sub>, Mn<sub>2</sub>Ti) is often overlooked in previous studies, but they can further deteriorate joint performance by increasing brittleness and inducing grain coarsening. This highlights the importance of considering the alloying elements in Ti and steel substrates when designing interface layers. In the study conducted by Kundu [47], it was observed that when bonding was carried out at a temperature of 900 °C for 45 min, joint failure occurred at the  $\lambda$  + FeTi phase, displaying a cleavage pattern indicative of brittleness. However, when bonding was performed at or above 950 °C, joint failure was found to be associated with the  $\sigma$  phase. The fracture morphology of the bonded joints at various temperatures [7]. The successful joining of bimetallic structures is facilitated by the matching surface closure through the diffusion of atoms and reaction products. The properties of these joints are primarily influenced by interfacial porosity. The growth of brittle IMCs governs the strength of the joints. After this, owing to the Kirkendall effect, voids develop in the vicinity of the interface region, leading to a further reduction in mechanical properties. Therefore, the strength of bonded joints is governed by the presence of interfacial porosity and

brittle IMCs. The fracture mode transitions from ductile-dominated to brittle-dominated with increasing temperature, which is consistent with the formation and growth of brittle IMCs and Kirkendall voids. This indicates that the optimal bonding window for non-interlayer diffusion welding is narrow (800–850 °C, 60–90 min) to balance IMC formation and atomic diffusion.



**Figure 1.** (a) Illustration of hot isostatic pressing process; (b) molecular dynamic model of single-crystal and polycrystal Fe-Ti diffusion couple; (c–h) microstructure of Fe-Ti interface after HIP at 950 °C and 30 MPa for 40 min [45].

In summary, the microstructure of diffusion-bonded joints between Ti and stainless steel, without the use of an interlayer, is primarily influenced by three key factors: bonding temperature, bonding duration, and the material composition. Lower bonding temperatures help limit the presence of brittle IMCs (e.g., FeTi, Fe<sub>2</sub>Ti, λ, and σ) at the interface, albeit with limitations on joint strength. For diffusion bonding without an interlayer, the optimal temperature range is typically 800–950 °C over a duration of 60–120 min. Higher bonding temperatures, in accordance with Fick's second law, favor shorter bonding times for achieving higher strength, and vice versa. Furthermore, the specific composition of the Ti and stainless steel significantly impacts the types of IMCs formed at the joint interface. The interaction of elements such as V and Al from the Ti and C, Mn, and Si from the

stainless steel during the diffusion bonding process leads to the creation of IMCs like TiC, Fe<sub>2</sub>V<sub>3</sub>, Mn<sub>2</sub>Ti, Fe<sub>3</sub>Al<sub>2</sub>Si<sub>4</sub>, and Al<sub>6</sub>Ti<sub>19</sub> within the intermetallic layer.

This diffusion bonding process without the interlayer can be summarized as follows: the initial stage involves matching surface closure through the diffusion of atoms at a specific temperature, with the subsequent formation of reaction products that facilitate the bonding of bimetallic structures. The second stage witnesses the development of brittle IMCs, and their width directly influences joint strength. Finally, the Kirkendall effect induces the formation of voids near the interface region in the third stage, further diminishing mechanical properties.

When opting for solid-state welding over fusion welding to join the non-interface layer between Ti and steel, several key considerations come into play. Firstly, solid-state welding, conducted at relatively lower temperatures, mitigates the risk of chemical reactions between Ti and steel, commonly observed at elevated temperatures. This prevention reduces the formation of brittle IMCs, consequently enhancing joint toughness. Secondly, solid-state welding, performed under conditions of incomplete material melting, averts the creation of brittle IMCs in the molten state, further contributing to the weld seam's toughness. Additionally, the lower welding temperatures associated with solid-state welding minimize heat input, reducing the impact on material properties and limiting the formation of heat-affected areas. Lastly, in contrast to fusion welding, which poses risks of oxidation and contamination due to high temperatures and a molten state, solid-state welding avoids such issues, preserving the purity of the weld area.

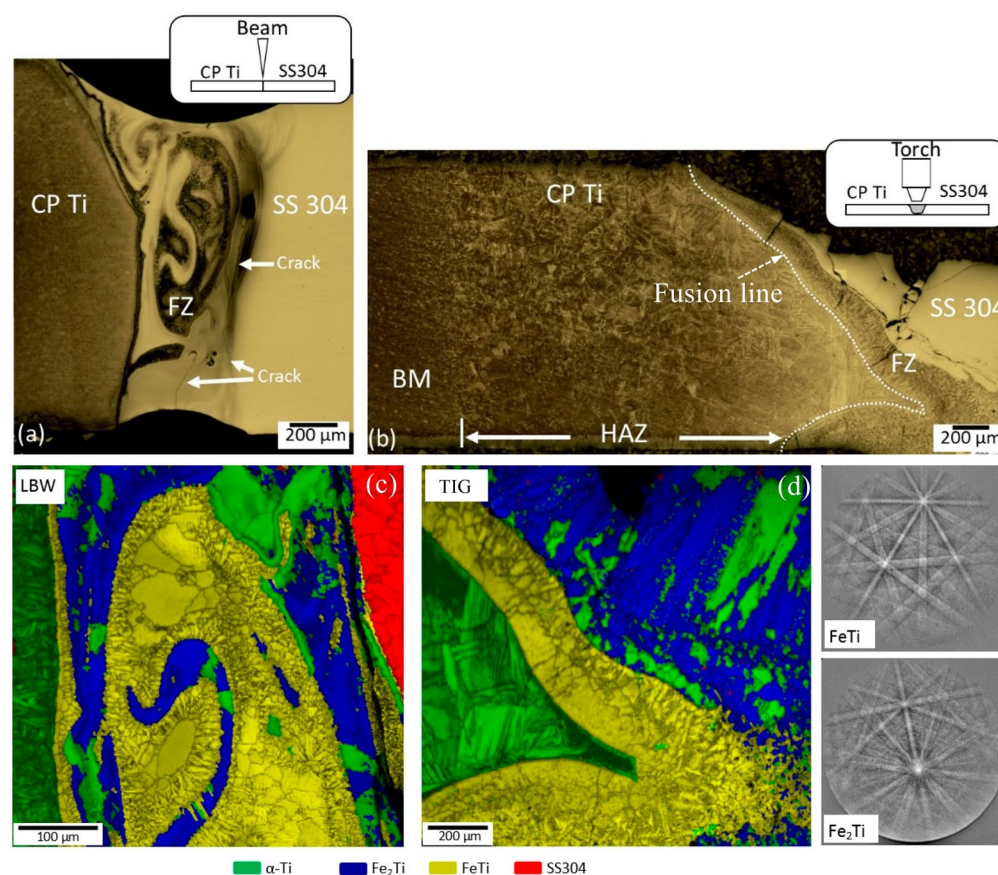
## 2.2. Other Solid-Phase Bonding Methods

Except for diffusion welding, other solid-state bonding methods face limitations and are less frequently reported. For instance, in an experiment [48], it was observed that employing a (Ni)/Cr barrier layer enhanced the shear strength of Ti brazed joints to 304 SS. Notably, Li et al. [49] achieved defect-free butt welds of Ti/steel through friction stir welding, using tungsten rhenium (W-25% Re) cutting tools at speeds of 600 and 950 rpm and a constant traverse speed of 47.5 mm/min. Given the material mixing in friction welding, controlling the formation of brittle IMCs and assessing their impact on the weld seam remains challenging. However, Li et al. [49] did not provide detailed information on the IMC distribution and its influence on joint toughness, which is a critical gap. Future studies should focus on the correlation between tool parameters (e.g., rotation speed, traverse speed) and IMC formation in friction stir-welded Ti–steel joints. There is an urgent need to investigate the friction stir welding performance of Ti/steel to enhance the applicability of Ti/steel components.

Song et al. [50] conducted a comprehensive study on the microstructure of Ti/steel interfaces formed through explosive welding. They highlighted the significant role of brittle IMCs in influencing the overall mechanical properties under shear or bending loading conditions. In a related investigation, Zhou et al. [51] delved into the fracture morphologies and shear properties of TA2/Q235 explosive welding interfaces using in situ SEM analysis. Their findings revealed the presence of microcracks, dimples, and the fragmentation of brittle IMCs on fracture surfaces, indicating a mixed fracture mode characterized by both brittle and ductile components, with the former being predominant. However, even with high cooling rates obtained by this process, it was not possible to make any sound joint. The high cooling rate in explosive welding can suppress the growth of IMCs, but the severe plastic deformation may induce residual stresses, leading to microcracks. This suggests that post-weld heat treatment (e.g., annealing at 500–600 °C) may be necessary to relieve residual stresses and improve joint integrity [51].

### 2.3. Fusion Welding

Moreover, fusion welding is not a viable method for directly bonding Ti and steel due to the fundamental disparities in their melting points and the high energy required for this process. The substantial difference in melting points between Ti and steel makes it challenging to achieve fusion between these materials in a single welding operation. The primary issue with fusing Ti and steel is the formation of zones that contain brittle IMCs such as  $\text{TiFe}_2$  (with a hardness exceeding 1300 HV) and  $\text{TiCr}_2$ . Consequently, this approach is unsuitable for directly joining these materials. It should be noted that “self-welding” (direct fusion welding without filler metal or interlayer) is not a standardized term in the welding community; this paper uses it to specifically refer to the direct fusion welding of Ti and steel without adding any filler metal, interlayer, or barrier layer. In the field of electron beam welding, Ting et al. [52] performed direct electron beam welding on TA15 and 304 stainless steel. This process resulted in the formation of continuously distributed composite phases in the weld seam, predominantly including  $\text{TiFe}$ ,  $\text{Cr}_2\text{Ti}$ , and others. However, numerous cracks appeared within the weld seam, preventing a successful connection from being established. The inability to form an effective joint between CPTi and SS304 using either LBW or GTAW was attributed to the formation of Fe-Ti brittle phases, which led to post-weld cracking. While LBW nearly achieved joint formation, GTAW joints separated immediately after welding, with fracture surfaces dominated by brittle features [53], as shown in Figure 2.



**Figure 2.** (a,b) Optical micrographs of CPTi/SS304 joints by laser welding and TIG welding; (c,d) phase distribution near CPTi/SS304 interface [53].

Similar to electron beam welding, self-welding of Ti and steel did not yield sufficient mechanical strength (approximately 35% of the ultimate tensile strength of Ti). In these welds, brittle failures occurred at the fusion zone (FZ) or at the interface of Ti and steel [54].

High shrinkage and tensile stresses in the steel side led to the formation of dendritic cracks perpendicular to the welding direction, some of which included the  $\lambda$ -Laves phase. The upper welding zone (WZ) primarily consisted of dendritic  $\beta$ -Ti and Ti-Fe IMCs, while the lower part contained single-phase, fracture-resistant  $\beta$ -Ti(Fe). Regardless of the cooling rate employed, self-welding of dissimilar combinations of Ti and steel resulted in cracking and embrittlement, as reported by Shanmugarajan and Padmanabham [55]. The low solid solubility of iron in Ti at room temperature is attributed to its unique lattice structure. This low solubility leads to the formation of Ti/iron IMCs at the overlap joint interface between Ti6Al4V and 42CrMo steel. The thickness of the IMCs increases with higher heat input, which negatively impacts joint strength [56]. Zhang et al. [57] explored the pulse laser welding of Ti and stainless steel without the use of filler metal. When the laser beam was precisely focused on the interface between the Ti and stainless steel, a significant quantity of brittle Ti-Fe IMCs formed within the joint. These Ti-Fe IMCs proved to be highly brittle, ultimately leading to spontaneous joint cracking post-welding. Physical model clearly illustrates the interdiffusion of Ti and Fe atoms at the interface, resulting in the formation of TiFe and TiFe<sub>2</sub> phases. The lack of a diffusion barrier leads to uncontrolled IMC growth, which is the root cause of joint failure. This further confirms that non-interlayer fusion welding is not feasible for Ti–steel joints [57].

The aforementioned experiments collectively affirm that the addition of a non-interface layer is not applicable in the context of fusion welding. This conclusion is drawn from the observed outcomes where the non-interface layer, when introduced, fails to yield desirable results in fusion welding processes. The inherent characteristics of fusion welding, involving high temperatures and the molten state, seem to hinder the effectiveness of the non-interface layer, leading to suboptimal welding performance. As evidenced by the experimental results, fusion welding processes are not conducive to the successful incorporation or utilization of a non-interface layer.

#### 2.4. Conclusions

In summary, joining Ti alloy and steel without an interlayer is feasible only for solid-state welding methods (such as diffusion bonding, friction stir welding, and explosive welding), whereas fusion welding is unsuitable. Solid-state welding, conducted at relatively lower temperatures or without full melting, effectively suppresses the formation of brittle intermetallic compounds (IMCs) at the Ti/steel interface, thereby improving joint toughness and reducing heat-affected zones and oxidation/contamination issues. Among these, diffusion bonding is the most commonly employed method. The microstructure of diffusion-bonded joints without an interlayer is governed primarily by bonding temperature, holding time, and material composition. Lower temperatures ( $\leq 700$  °C) help limit brittle IMCs but may compromise joint strength due to incomplete atomic diffusion. Higher temperatures ( $\geq 900$  °C) promote excessive growth of the IMC layer and formation of Kirkendall voids, leading to brittle fracture. The optimal process window is typically 800–950 °C for 60–120 min. Additionally, alloying elements from Ti and steel can form extra IMCs, further degrading joint performance.

Friction stir welding and explosive welding can also achieve joining without an interlayer, but they face challenges such as uncontrolled IMC formation and residual stress-induced microcracks. In contrast, fusion welding methods (including electron beam welding and laser welding) inevitably generate large amounts of brittle Ti-Fe IMCs under high temperatures, accompanied by severe thermal stresses and cracking, making it impossible to obtain reliable joints. Therefore, the non-interlayer joining approach is only viable for solid-state welding of Ti/steel and is not applicable to fusion welding processes.

### 3. Joints with Single Interlayer

Studies indicate that incorporating suitable intermediary layers can effectively prevent the generation of IMCs in direct contact between Ti and steel. In practical manufacturing, common materials for these intermediary layers include V, Zr, Ta, Nb, and Mo, which readily dissolve with Ti. Additionally, materials such as Cu, Ni, etc., are employed to inhibit the diffusion of C elements. Multi-component composite layers, such as V-Cu Ni, Mo Ta V, and amorphous alloys like Al, Fe, Ag, etc., are also utilized in practical production scenarios. Referencing the binary alloy phase diagram [58], it is observed that Ti tends to form stable yet brittle IMCs with most elements, excluding a select few in the same or adjacent groups on the periodic table, such as V, Nb, Ta, Mo, and W. Similarly, Fe is a reactive element known for its capability to create IMCs with various elements, except for Co, Mn, Cu, and Ag.

In the subsequent sections, our focus will be directed towards distinct chemical elements, meticulously examining their influence on the interface layer across diverse process parameters. Through a comprehensive analysis of various element incorporations, we aim to delve deeply into their ramifications on interface layer formation and functionality. This scrutiny is poised to furnish an insightful comprehension, laying the groundwork for further scholarly exploration.

#### 3.1. Cu Element Composition Interface

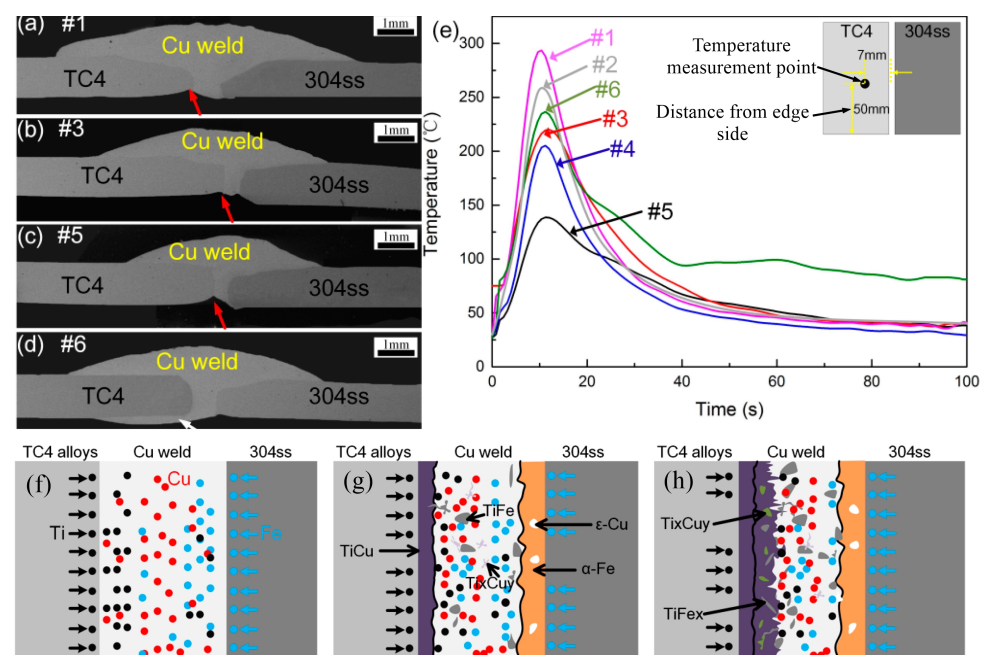
##### 3.1.1. Solid Phase Welding

Elrefaey et al. [59] achieved successful bonding of dissimilar Ti/steel metals using the diffusion bonding process aided by a Cu-based interlayer. While Cu-Ti IMCs are present in the interface structure, their impact on joint strength may be comparatively less than that of Fe-Ti IMCs. Diffusion takes place between the Cu interlayer and the Ti substrate. The diffusion distance from Cu to Ti is extensive, with a plateau observed in a limited section of the elemental dispersion curve, indicating the formation of IMCs in this region. Conversely, a significant portion of the diffusion zone primarily exhibits Ti, with minimal content of other elements. This suggests the formation of a solid solution in this region, typically advantageous for bonding. The Cu concentration decreases gradually from the interlayer to the Ti substrate, with a plateau at ~20–30 at.% Cu, corresponding to the formation of TiCu and Ti<sub>2</sub>Cu phases. These Cu-Ti IMCs have a hardness of ~400–600 HV, which is much lower than that of FeTi (950 HV), thus reducing the brittleness of the joint [59].

In their study of solid diffusion bonding between Ti and 304 stainless steel using Cu interlayers, Kundu et al. [60] noted the presence of ternary Fe-Cu-Ti products at 850 °C and 900 °C but their absence at 950 °C. The effectiveness of the Cu interlayer in preventing element diffusion was found to be insufficient. The highest bonding strength (318 MPa) and ductility (8.5%) were achieved at 900 °C due to improved surface coalescence, while strength decreased at 950 °C due to an increase in brittle IMCs and at 850 °C due to incomplete surface coalescence. The formation of Fe-Cu-Ti ternary phases (e.g.,  $\tau_2$ ,  $\tau_3$ ) at 850–900 °C is a key factor limiting joint strength. These ternary phases are more brittle than binary Cu-Ti IMCs, highlighting the need to control bonding temperature to avoid their formation [60]. Zakipour et al. [61] employed 50, 100, and 150  $\mu\text{m}$  interlayers for dissimilar stainless steel 316/Ti6Al4V diffusion bonding. Ti<sub>2</sub>Cu, TiCu, Cr<sub>2</sub>Ti, Fe<sub>2</sub>Ti, Cu<sub>0.8</sub>Fe<sub>0.2</sub>Ti, Fe<sub>4</sub>Cu<sub>3</sub>, and TiCu<sub>2</sub> were identified at the interface. Increasing the interlayer thickness led to diverse interface microstructures and reduced shear strength, dropping from 284 MPa to 145 MPa. At a bonding temperature of 850 °C for 60 min and 900 °C for 60 min, Cu-Ti and Fe-Cu-Ti ternary products were observed at the diffusion interface. However, with increased bonding temperature and time, the formation of Fe-Cu-Ti and Cu-Ti interphases did not occur at the diffusion interface. This process yielded a notable tensile strength

of 322 MPa and a ductility of 8.5%. The decrease in shear strength with increasing Cu interlayer thickness is attributed to the increased formation of Cu-Ti IMCs and Fe-Cu-Ti ternary phases. A 50  $\mu\text{m}$  Cu interlayer is optimal as it balances diffusion barrier efficiency and IMC formation [61].

Following the Cu-Ti phase diagram, in the temperature range of 890 to 960  $^{\circ}\text{C}$ , there are two eutectic transformations, allowing the introduction of a limited amount of the liquid phase at Cu-Ti joints by selecting an appropriate bonding temperature. Utilizing this principle, the transient liquid-phase (TLP) bonding technique is employed to join Ti and stainless steel. In TLP bonding, a Cu interlayer is positioned between the Ti and stainless steel, and the components are heated to a predetermined temperature, typically close to the eutectic transformation temperature [26,62]. Norouzi et al. [63] examined how varying bonding temperatures (870–960  $^{\circ}\text{C}$ ) affected transient liquid-phase bonding between Ti6Al4V and AISI 304 austenitic stainless steel. At 960  $^{\circ}\text{C}$ , complete elimination of eutectic and intermetallic zones was observed, indicating successful isothermal solidification. Fracture analysis at this temperature revealed a brittle-ductile fracture mode. The successful isothermal solidification at 960  $^{\circ}\text{C}$  is a critical finding, as it eliminates brittle eutectic phases and IMCs, resulting in a joint with high strength and ductility. This demonstrates that TLP bonding with a Cu interlayer is a promising method for Ti–steel joining, provided the bonding temperature and time are optimized [63]. Norouzi et al. [64] also investigated the influence of bonding time on transient liquid-phase bonded joints using Cu as the interlayer, calculating the time needed for full isothermal solidification as per [65]. The main difficulty in joining TC4 to 304 stainless steel via cold metal transfer is the formation of brittle Ti-Cu and Ti-Fe intermetallic phases at the Cu-weld/TC4 interface, especially under high heat input, leading to cracking and poor strength; this can be effectively overcome by using a Cu-based filler wire combined with preheating and a lower heat input, which improves wettability, increases the effective bonding area, suppresses brittle phases and crack extension, thereby raising the tensile fracture load by 47% (to 365 MPa) compared to joints without preheating, as shown in Figure 3 [66].



**Figure 3.** (a–d) Macrophotograph of the cross-section of the Ti/steel joint; (e) thermal cycles under different welding conditions; (f–h) schematic of element diffusion and microstructure evolution [66].

### 3.1.2. Electron Beam Welding

Wang et al. [67] pioneered the fusion welding of Ti to steel using a 1 mm Cu interlayer. Their two-pass electron beam welding resulted in a melted zone with TiFe<sub>2</sub> IMCs within Cu. The weld had a mechanical strength of 224 MPa but exhibited brittle fracture near the Ti/melted zone interface. The presence of TiFe<sub>2</sub> in the melted zone indicates that the Cu interlayer did not fully isolate Ti and Fe, possibly due to insufficient interlayer thickness or improper heat source positioning. Increasing the Cu interlayer thickness to 1.5 mm or shifting the electron beam towards the steel side may improve isolation efficiency [67]. Subsequent research [68] explored welding Ti6Al4V to stainless steel using pulsed Nd:YAG laser and electron beam with a 0.5 mm Cu foil interlayer. Placing the heat source at the Cu–steel interface reduced Ti melting but did not achieve full metallurgical isolation. Rapid cooling led to immiscible flows and strengthened the Cu barrier. Similar results were observed when AISI 304L was welded with Ti-6Al-2Mo-2V-2Zr alloy [69]. Tomashchuk et al. [70] conducted a comprehensive study investigating the impact of electron beam heat source shift and welding speed on the morphology, phase distribution, and mechanical properties of the weld joint between  $\alpha$ - $\beta$  Ti and AISI 316L stainless steel, with the aim of optimizing dissimilar interfaces. Their research unveiled that a slower welding speed led to near-equivalent fusion between the steel and Ti sides and uniform component dispersion in the Cu medium. Additionally, their numerical analysis [71] revealed that a positive beam offset resulted in the steel side of the weld being influenced by a Cu-rich flow. The velocity of this flow increased with welding speed, intensifying the interaction with the steel side. This was supported by the presence of substantial areas of liquid steel separated from the steel/FZ interface. The welding speed and beam offset significantly affect the mixing of Ti, Fe, and Cu. A welding speed of 1 m/min and beam offset of 0.25 mm result in uniform element distribution, minimizing IMC formation. This provides a practical guideline for electron beam welding with Cu interlayers.

The inner interface layers on the Ti side exhibit similar compositions. The formation of these phases can be elucidated by local invariant reactions:

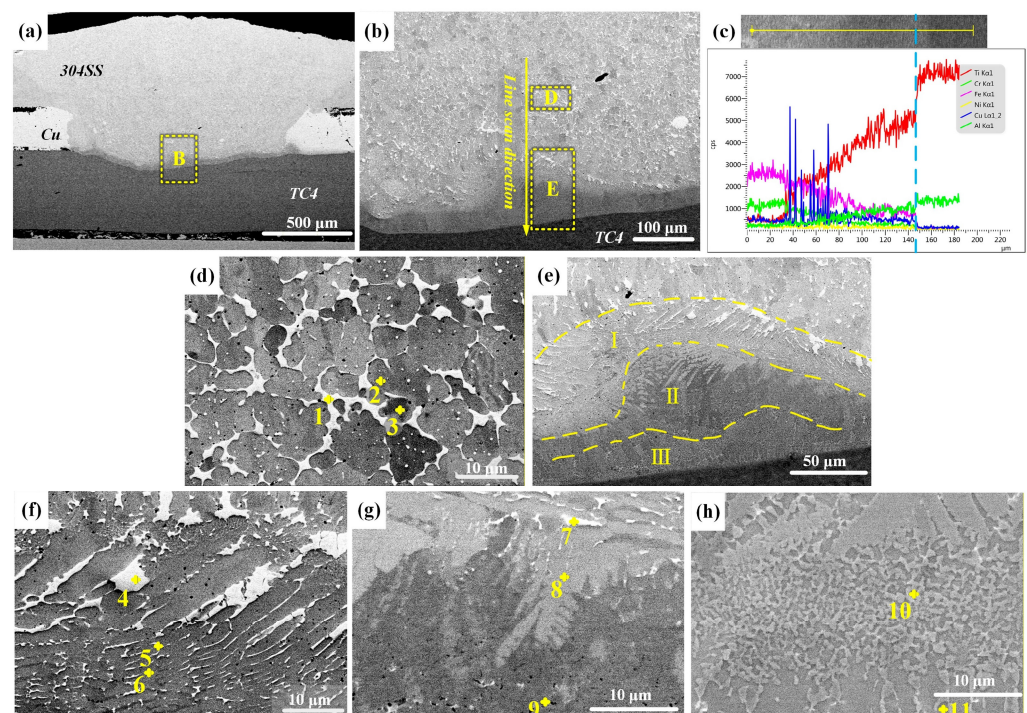


Analysis of the experimental data reveals that the mechanical stability of joints, especially those produced with an electron beam offset towards the steel side, is contingent on the diffusion path length within the Ti-rich interface. The optimal diffusion path length falls within the range of 40–80  $\mu\text{m}$ . This optimal range can be achieved by employing a 570  $\mu\text{m}$  Cu interlayer, utilizing high welding speeds (approximately 2 m/min), and positioning the electron beam precisely at the interface between the Cu interlayer and the steel.

### 3.1.3. Laser Welding

In the field of laser welding, Zhang et al. [21] achieved a remarkable tensile strength of up to 340 MPa in a pulsed laser-welded dissimilar joint, uniting Ti6Al4V Ti with 301 L stainless steel using Cu as an interlayer. Their research underscores that the use of Cu as an interlayer effectively mitigates the interaction between Ti and Fe, consequently reducing the formation of brittle Ti-Fe intermetallics in the weld and significantly enhancing its ductility. However, it is crucial to note that complete melting of the interlayer results in the mixing and reaction of Ti and Fe elements within the fusion pool [17]. The key to achieving high strength with Cu interlayer laser welding is to control the melting degree of the interlayer. Partial melting of Cu can form a liquid barrier, preventing Ti-Fe contact, while complete melting leads to element mixing and IMC formation [21]. While direct

welding of Ti to SS relies on a  $\sim 20\ \mu\text{m}$  thick TiFe intermetallic layer, the introduction of a Cu interlayer enables fusion welding through Ti-Fe-Cu ternary mixing, though it also leads to localized brittleness at the TC4 side due to the formation of  $\text{TiFe}_2$  and  $\text{TiCu}_4$  phases, in contrast to the hybrid bonding mechanisms offered by Mo and Nb interlayers, as shown in Figure 4 [72]. Cherepanov et al. [73] validated that applying a negative defocus of the laser when targeting the Cu insert positioned between Ti and steel resulted in an inhomogeneous molten pool characterized by the presence of porosity. Conversely, when the focal point was directly placed onto the Cu insert, cracks emerged, leading to joint failure. Within the FZ, a supersaturated solid solution of Fe, Cr, Ni, and Ti was observed within the crystal lattice of Cu. Notably, uniformly distributed IMCs were present, including  $\text{Ti}(\text{Fe}, \text{Cr})_2$ , ranging in size from 5 to  $200\ \mu\text{m}$ , and  $\text{TiCu}_3$ , measuring between 1 and  $10\ \mu\text{m}$ . Furthermore, diffusion boundaries with dimensions of 0.1–0.2 mm formed at both the SS and Ti sides. The formation of porosity and cracks is attributed to improper laser focus position. A positive defocus (focus above the Cu insert) may reduce heat input concentration, improving molten pool homogeneity and reducing defects [73]. Zhang et al. [74] employed pulse laser welding to join Ti and stainless steel, incorporating Cu as an interlayer. The inclusion of an unliquified Cu intermediate layer serves to ensure the creation of a weld seam devoid of cracks, with the absence of brittle Ti-Fe IMCs. The resultant joint exhibits a tensile strength of up to 320 MPa. The reaction layer mainly contains  $\text{Cu}_2\text{Ti}$  and  $\text{Cu}_4\text{Ti}$  at the Ti/Cu interface. SEM image reveals a continuous and uniform  $\text{Cu}_2\text{Ti}/\text{Cu}_4\text{Ti}$  reaction layer with a thickness of  $\sim 5\ \mu\text{m}$ . This thin reaction layer provides sufficient bonding strength without inducing brittleness, which is the reason for the high joint strength [74]. Additionally, high-resolution TEM (HRTEM) analysis of this interface confirms the coherent interface between  $\text{Cu}_2\text{Ti}$  and Ti, with a lattice mismatch of  $\sim 3\%$ , further enhancing the bonding stability.



**Figure 4.** SEM images of Cu-interlayer-assisted TC4/304SS laser-welded joints: (a) Cross-section of joint; (b) enlarged view of area B; (c) line scanning analysis; (d) magnified image of area D; (e) magnified image of area E; (f) magnified image of area I; (g) magnified image of area II; (h) magnified image of area III [73].

### 3.1.4. Other Welding Methods

Hao et al. [75] employed a Cu-filled welding wire for the arc welding process to join 304 stainless steel with Ti alloy. They observed the generation of Fe-Ti IMCs with increased heat input. A Ti-Cu IMC layer with a duplex structure formed between the Cu weld and the Ti6Al4V plate. Arc welding has a higher heat input compared to laser/electron beam welding, leading to more severe element diffusion and IMC formation. Using a Cu-filled welding wire with a low Cu content (e.g., 50–60 wt.%) may reduce Ti-Cu IMC formation while maintaining the diffusion barrier effect [75]. Kundu et al. [60] chose Cu as the intermediate layer and employed a diffusion welding process at 900 °C for 1–1.5 h to connect commercial pure Ti and 304 stainless steel, resulting in a tensile strength of 320 MPa. The excellent plasticity of copper led to a maximum elongation of 8.5% in the joint. Brittle Ti-Fe-Cu ternary compounds were identified in both diffusion layers on both sides, denoted as  $\tau_2$ ,  $\tau_3$ , and  $\tau_5$ . Kumar et al. [76] utilized a pre-placed 0.65 mm pure Cu as the intermediate layer metal, achieving a joint strength of 523 MPa. Energy spectrum analysis revealed separate reactions of Ti and Fe atoms with Cu, with Cu acting as an effective barrier. The exceptionally high tensile strength of 523 MPa achieved by Kumar et al. [76] is attributed to the formation of a thin and continuous Cu-Ti reaction layer ( $\sim 3 \mu\text{m}$ ) and the absence of Fe-Ti IMCs. This highlights the importance of interlayer thickness control in diffusion welding [76].

### 3.1.5. Conclusions

The compatibility of the Cu element with Ti and Fe elements provides a promising solution for connecting Ti and Fe. Due to the minimal solid solubility between Cu and Fe, the formation of IMCs is effectively prevented. The resulting connection predominantly comprises Cu-based solid solutions and Fe-based solid solutions. While Cu-Ti IMCs may emerge, they exhibit a “soft phase” compared to the brittleness associated with Ti-Fe compounds [40,74]. Consequently, Cu stands out as the preferred intermediate layer material for various applications, including rolling connections of Ti/steel composite plates and diffusion connections between Ti and steel. Additionally, the interface design with Cu interlayers should consider the synergistic effect of interlayer thickness, welding process, and heat input. For example, thin Cu interlayers (50–100  $\mu\text{m}$ ) are suitable for laser/electron beam welding (low heat input), while thicker interlayers (150–200  $\mu\text{m}$ ) may be required for arc welding (high heat input) to ensure effective isolation of Ti and Fe. Future research should focus on optimizing the Cu interlayer composition (e.g., adding trace V, Nb) to further improve joint strength and corrosion resistance.

## 3.2. V Element Composition Interface

### 3.2.1. Beam Welding

Furthermore, apart from the work of Wang et al. [53] mentioned previously, Nogami et al. [77] achieved a flawless fusion between pure V and 316L stainless steel using electron beam welding. Notably, the weld metal exhibited a hardness below 600 HV, indicating the absence of the  $\sigma$ -phase. In the Fe-V system [78], extensive areas of continuous solid solutions exist at elevated temperatures ( $>1219 \text{ }^\circ\text{C}$ ), along with a metastable  $\sigma$ -phase that occurs within a temperature range of 600–1219 °C, encompassing compositions of 19–78% V by weight. The presence of this brittle  $\sigma$ -phase sometimes limits the direct fusion of Fe and V alloys. However, effective control over these  $\sigma$ -phases in the Fe-V system can be achieved by reducing the V content in the molten steel/V zone to levels below the  $\sigma$ -phase formation range or through rapid quenching of the Fe-V melt to temperatures below 650 °C. This leads to the formation of an Fe-V solid solution, hereafter referred to as (Fe, V)<sub>ss</sub>, instead of the  $\sigma$ -phase. The key design principle of V as an interface layer is its ability to form

continuous solid solutions with both  $\beta$ -(Ti, V)<sub>ss</sub> and (Fe, V)<sub>ss</sub>, thus avoiding brittle IMC formation. However, the formation of the  $\sigma$ -phase in the Fe-V system is a critical challenge that must be addressed.

### 3.2.2. Lasing Welding

Tomashchuk et al. [15] explored the continuous laser bonding of AISI316L stainless steel and a Ti-6Al-4 V alloy using a pure V interlayer, achieving a joint with a strength of 367 MPa and low hardness (<600 HV). The formation of (Fe, V)<sub>ss</sub> in the melted zone between AISI 316L stainless steel and V was attributed to rapid cooling, leading the initial solid solution to segregate into V-rich and V-poor clusters. It was noted that regions with compositions within the  $\sigma$ -phase homogeneity range are prone to cracking, emphasizing the importance of reducing V content in the molten steel/V region below the  $\sigma$ -phase formation threshold. The rapid cooling rate in laser welding (103–104 °C/s) is crucial for suppressing  $\sigma$ -phase formation. This highlights the compatibility of V interlayers with high-energy beam welding processes, which offer precise heat input and rapid cooling [15]. While existing studies have primarily focused on the effect of V composition on post-weld heat treatment hardness [77], investigations specifically addressing the impact of V content on tensile properties in welded joints remain limited. In most successful Ti-Fe-V joint studies, high-density fusion welding processes like laser or electron beam welding are preferred due to their precise, localized heat input and rapid heating/cooling characteristics.

The Melt Zone #1 (MZ1) situated between Ti-6Al-4V and V primarily consisted of a continuous ( $\beta$ Ti, V)<sub>ss</sub> phase with a V composition of up to 26 at.% [15]. The boundary between V and MZ2 exhibited an FeV solid solution, while the interface between 17-4PH and MZ2 included Fe + (Fe, V)<sub>ss</sub> + (CrV) solid solutions. This region displayed a uniform distribution of Fe and V. The composition of the FeV solid solution corresponded to the  $\sigma$ -phase. However, XRD analysis confirmed the presence of (Fe, V)<sub>ss</sub>, aligning with the phases observed via SEM/EDS examination. This affirms that the rapid cooling post-welding inhibited the formation of the (Fe, V)  $\sigma$ -phase, instead yielding an FeV solid solution, in agreement with Ustinovshikov's proposal [79]. HRTEM analysis reveals that the  $\beta$ -(Ti, V) ss phase has a BCC structure with a lattice parameter of ~0.33 nm, which is consistent with the  $\beta$ -Ti lattice, indicating a coherent interface between Ti and V. This coherence enhances the bonding strength and ductility of the joint.

Adomako et al. [80] used vanadium intermediate films for laser welding of Ti 6Al 4V and 17-4PH stainless steel, emphasizing the need for caution when adjusting the laser beam to avoid very low V composition, as this may lead to incomplete fusion or unwelded areas. Tomashchuk [15] investigated the feasibility of continuous Yb: YAG laser joining of Ti6Al4V with a ~316 mm pure V interlayer to AISI 316L stainless steel. The interface between AISI 316L and MZ1 consists of a light layer of  $\gamma$ -Fe + (Cr, V) solid solutions, followed by a dark layer with a composition corresponding to a mixture of (Fe, V)' $\sigma$  + (Cr, V), and zones of (Fe, V)  $\sigma$  and (Fe, V)'  $\sigma$  solid solutions. In the melted zone adjacent to the MZ1/V interface, numerous cracks are present. These cracks can be attributed to a combination of residual stresses near the V interface and an increase in the ductile-brittle transition temperature in this region due to the high levels of Cr and Fe in the V matrix [15,81]. The MZ2 formed between Ti6Al4V and V consists of continuous (Ti, V) solid solutions and does not exhibit any defects. Local heterogeneities, determined by insufficient convection for MZ homogenization, can be observed. The cracks highlight the limitation of V interlayers: high Cr/Fe content in the V matrix can induce  $\sigma$ -phase formation and residual stresses. To address this, a multi-layer interlayer (e.g., V/Cu) may be used to reduce Cr/Fe diffusion into V.

### 3.2.3. Conclusions

V was initially deemed a favorable transition metal due to its remarkable ability to form a stable solid solution with Ti, as evidenced by the binary alloy phase diagram for the Ti-V system. The beta-phase Ti exhibits a comprehensive range of solid solutions with V, while the alpha-phase Ti displays more limited behavior in this regard. The advantageous properties of V as a transition metal are further accentuated by thermal expansion coefficients, resulting in a Ti:V ratio of 8.5:8.3 [82]. Some researchers have employed V as an intermediate layer in the argon arc welding of pure iron and pure Ti. However, the extensive heat input and arc coverage in arc welding lead to the formation of numerous IMCs, adversely impacting joint performance. Consequently, the strength achieved with an intermediate layer is limited to less than 300 MPa. When utilizing V as the intermediate layer, joint performance is contingent upon the interface characteristics between stainless steel and V. The brittleness of the  $\sigma$  (Fe, V) phase, resulting from the interaction between Fe and V, contributes to a decline in joint performance. It can be deduced that the application of melt pool metallurgy holds the potential to optimize weld microstructure and enhance joint performance. To summarize, V interlayers are most suitable for high-energy beam welding (laser/electron beam) due to their ability to form solid solutions with Ti/Fe and the rapid cooling that suppresses  $\sigma$ -phase formation. The key design principles for V interlayers include: (1) controlling V content in the Fe-V melted zone below 19 at.% to avoid  $\sigma$ -phase; (2) using rapid cooling processes to inhibit phase transformation; (3) optimizing heat source positioning to ensure uniform mixing. Future research should explore alloying V with trace elements (e.g., Ti, Nb) to further stabilize the (Fe, V)<sub>ss</sub> phase and reduce residual stresses.

### 3.3. Ni Element Composition Interface

#### 3.3.1. Diffusion Welding

Muralimohan et al. [83] investigated the friction welding of 304L stainless steel with commercially pure Ti, employing an electroplated Ni interlayer. The findings indicated that, under optimal parameters, atomic diffusion was effective, and the formation of Fe-Ti IMCs was absent. With increasing friction time, the thermal degradation zone expanded, leading to a reduction in the thickness of the interlayer material. Microhardness testing at the connection interface revealed the impact of solution hardening in the weld zone. Furthermore, under a constant friction time, the tensile strength exhibited an increase with the rise in forging pressure. The absence of Fe-Ti IMCs is attributed to the formation of a Ni-Ti reaction layer ( $\text{Ni}_3\text{Ti}$ , NiTi) that acts as a diffusion barrier. Increasing forging pressure enhances interfacial contact, promoting atomic diffusion and improving joint strength [83].

Kundu and Chatterjee [84] used Ni as the interlayer and found that 300  $\mu\text{m}$  thick Ni intermediate film can completely prevent the diffusion of Ti towards the stainless steel side, with a maximum connection temperature of 900 °C. The maximum tensile strength of the transition joint processed at 302 °C is 219 MPa, and the shear strength is 900 MPa. Sam et al. [84] employed a 150  $\mu\text{m}$  Ni alloy (comprising Ni 78.76 wt.%, Fe 15.6 wt.%, Mo 4.9 wt.%, and Al 0.74 wt.%) as an interlayer for the diffusion bonding of Ti and microduplex stainless steel (MDSS). The microstructure of joints created at 900 °C for 45 min is depicted. At the MDSS-NiA interface, no intermetallic phases are observed [84]. Upon closer examination, layers A to C reveal the presence of  $\text{Ni}_3\text{Ti}$ , NiTi, and  $\text{NiTi}_2$  phases, all possessing a relatively lower hardness of 662 HV compared to FeTi phases (950 HV). These joints exhibit a maximum tensile strength of 560 MPa. The high tensile strength of 560 MPa achieved with the Ni alloy interlayer is due to the formation of ductile Ni-Ti IMCs ( $\text{Ni}_3\text{Ti}$ , NiTi) and the absence of Fe-Ti phases. The addition of Mo and Al to the Ni interlayer enhances the stability of the Ni-Ti reaction layer and improves corrosion resistance.

Thirunavukarasu et al. [85] explored the impact of bonding time on the interfacial reaction and mechanical properties of Ti6Al4V diffusion-bonded joints utilizing Ni as an intermediate material and 304 stainless steel. Their findings revealed that, under a 4 MPa compression load at 1023 K (750 °C) for 60 min, the diffusion-bonded joint achieved a peak tensile strength of approximately 382 MPa and exhibited substantial ductility of around 3.7% [85]. The tensile strength increases with bonding time up to 60 min, then decreases. This is because prolonged bonding time promotes the growth of Ni-Ti IMCs, leading to brittleness. The optimal bonding time of 60 min balances IMC formation and atomic diffusion.

### 3.3.2. Beam Welding

Wang et al. [86] conducted electron beam welding experiments to join Ti and stainless steel, employing various filler metals such as Ni, V, and Cu. They utilized optical microscopy, scanning electron microscopy, and X-ray diffraction to meticulously investigate the joint's microstructure. Moreover, the mechanical properties of the joint were assessed through tensile strength and microhardness measurements. This research delved into the influence of filler metals on the microstructural features and mechanical attributes of electron beam-welded joints between Ti and stainless steel. The findings revealed that all filler metals played a pivotal role in curbing the formation of Ti-Fe IMCs. Irrespective of the filler metal used, the joints consistently exhibited solid solution phases and interfacial IMCs. However, the precise composition of these solid solutions and interfacial IMCs hinged on the metallurgical reactions between the filler metals and the base materials. Notably, the interfacial IMCs observed in joints with Ni, V, and Cu filler metals included  $\text{Fe}_2\text{Ti} + \text{Ni}_3\text{Ti} + \text{NiTi}_2$ ,  $\text{TiFe}$ , and  $\text{Cu}_2\text{Ti} + \text{CuTi} + \text{CuTi}_2$ , respectively. The formation of  $\text{Ni}_3\text{Ti}$  and  $\text{NiTi}_2$  phases in Ni-filled joints contributes to moderate hardness and high strength, while Cu-filled joints have lower hardness due to ductile Cu-Ti IMCs. V-filled joints have high hardness due to (Fe, V)ss formation but lower ductility [86].

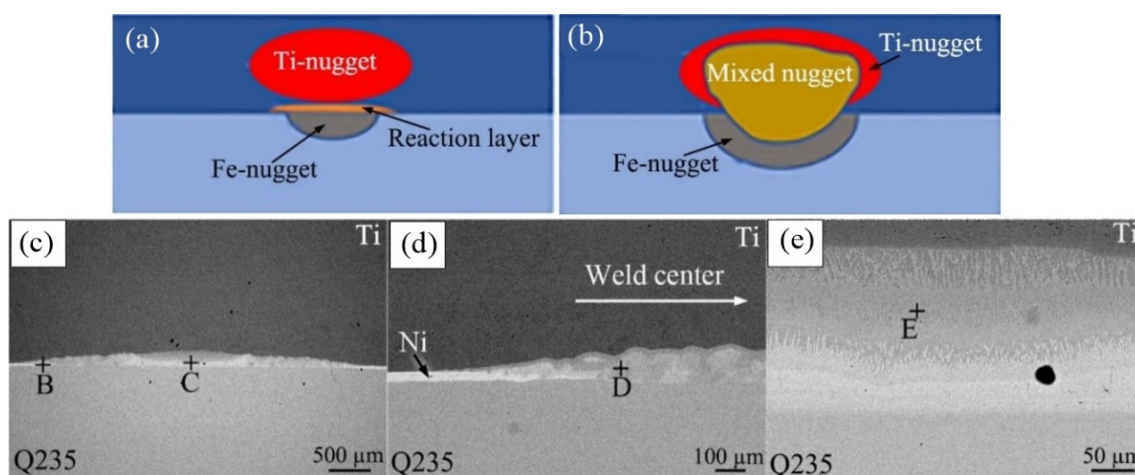
Filler metal choice significantly affects joint performance, notably influencing microhardness and tensile strength [86]. V-filled joints display elevated microhardness, indicating substantial hardening effects, while Cu-filled joints exhibit comparatively lower microhardness, suggesting milder strengthening effects. In contrast, Ni-filled joints strike a balance, showcasing moderate enhancements in both hardness and strength. This versatility makes Ni an eclectic option for engineering applications requiring a combination of hardness and tensile strength. However, the tensile strength of Ni-filled joints (124 MPa) is lower than that of Cu-filled joints (234 MPa), which may be due to the formation of more brittle Ni-Ti-Fe ternary phases. Optimizing the Ni filler metal composition (e.g., adding Cu) may improve ductility.

### 3.3.3. Other Welding Methods

Wang et al. [87] employed pure Ni as the transition metal in the laser cladding method to create a transition layer for composite plate transition welding. Initially, laser welding facilitated self-fusion welding of the Ti-clad plate, followed by the formation of a pure Ni transition layer using the laser cladding method. Subsequently, GMAW was employed to weld the steel substrate. The laser cladding method for transition layer preparation effectively reduced the dissolution of Ti and steel layers during welding, resulting in the generation of an approximately 200  $\mu\text{m}$  thick TiNi compound layer on the adjacent Ti side. The TiNi compound exhibited the presence of  $\text{Fe}_2\text{Ti}$  and  $\text{Ti}_6\text{O}$  phases near the steel side, contributing to an increased joint hardness, reaching a maximum of 720 HV. The average tensile strength of the joint was measured at 396 MPa, with an impact absorption energy

ranging between 74 and 84 J. The fracture surfaces of the Ti layer and transition layer displayed brittle characteristics.

Zhang et al. [88] utilized Ni<sub>30</sub>CrMo alloy as a filler material in composite plate transition welding through the TIG method. The microstructure of the joint revealed that the primary component in the Ti weld seam structure was  $\beta$ -Ti, with the presence of NiTi<sub>2</sub>, Ni<sub>3</sub>Ti, TiFe, and TiFe<sub>2</sub> phases, indicating incomplete blocking of the Ti and Fe combination. In the nickel transition layer,  $\gamma$ -Ni was the predominant component, and there was an approximately 50  $\mu$ m width between the Ti layer and the nickel layer. The transition layer contained a significant amount of needle-like NiTi<sub>2</sub>, NiTi, and CrNi<sub>2</sub> compounds, resulting in increased weld hardness but lower joint strength and toughness compared to the base material. The key challenge in Ti/steel welding is the formation of brittle Ti-Fe intermetallic compounds (IMCs) when the Ti and steel nuggets mix at high currents (>8 kA); this is effectively addressed by using a Ni interlayer and keeping the nuggets unmixed at lower currents (e.g., 8 kA), which promotes favorable Ti-Ni and Ti<sub>2</sub>Ni IMCs, thereby achieving a maximum tensile shear load of approximately 8.02 kN, as shown in Figure 5 [89].



**Figure 5.** (a,b) Schematic diagram of the Ti/steel resistance spot welding joint with DS-nugget and C-nugget; (c–e) SEM images of the Ti/steel joint with DS-nugget [89].

### 3.3.4. Conclusions

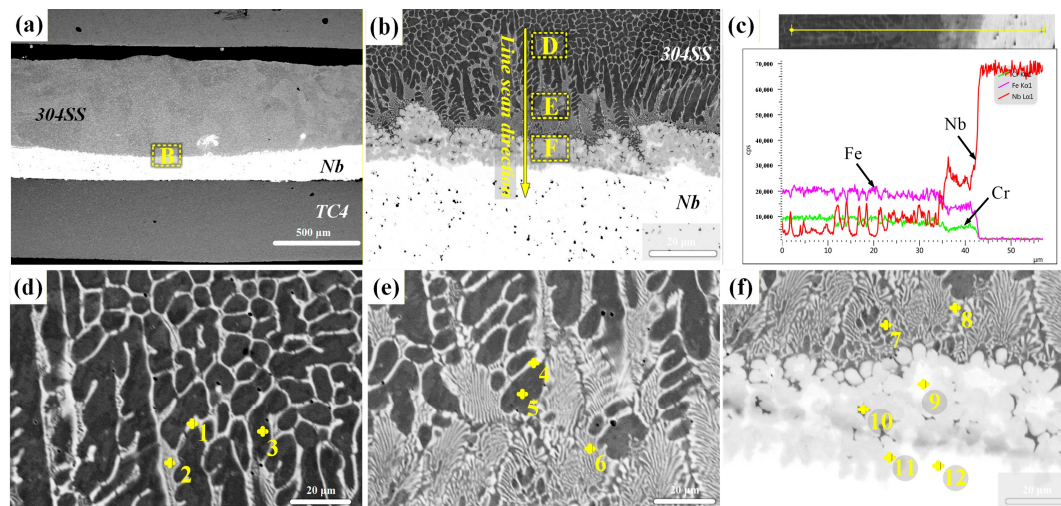
Ni exhibits excellent metallurgical compatibility with Fe in steel, allowing for infinite dissolution and rendering it an optimal choice for steel connections. In contrast to the Ti-Fe compounds formed by Ti, the IMCs resulting from the reaction between Ni and Ti elements demonstrate lower brittleness. Consequently, pure Ni is extensively utilized as an intermediate transition layer metal to facilitate the connection between Ti and steel. In comparison to joints filled with Cu and V, Ni-type filled joints strike a favorable balance between hardness and tensile strength. These Ni-filled joints not only provide ample strength but also maintain a requisite level of toughness, thereby enhancing the overall performance of the connected components. This equilibrium makes Ni-type filled joints a prudent and efficacious choice for linking Ti and steel.

## 3.4. Other Element Composition Interface

### 3.4.1. Nb Element Composition Interface

To prevent the undesirable mixing of Ti and Fe, Zhang et al. [21] introduced a laser offsetting technique utilizing a Nb interlayer with a higher melting point than the parent materials. This approach resulted in a hybrid joint characterized by fusion welding at the Ti-Nb interface and inter-atomic diffusion bonding, accompanied by eutectic reactions at the Nb-SS interface, yielding a robust joint with an impressive ultimate tensile strength

(UTS) of 370 MPa. The Nb on the Ti side underwent mixing without the formation of any IMCs, while the SS side remained unmelted, with the presence of a reaction layer. This reaction layer consisted of Nb/Fe<sub>7</sub>Nb<sub>6</sub>/Fe<sub>2</sub>Nb +  $\alpha$ -Fe/ $\alpha$ -Fe/SS phases, with Fe exhibiting a faster diffusion rate compared to Nb. The main challenge in direct Ti/steel welding is the formation of brittle Fe-Ti intermetallic compounds (IMCs); this can be addressed by introducing a Nb interlayer, which instead forms favorable Fe-Nb IMCs (Fe<sub>2</sub>Nb and Fe<sub>7</sub>Nb<sub>6</sub>) with the steel side, as evidenced by the phase distributions in areas D-F where honeycomb  $\alpha$ -Fe, network Fe<sub>2</sub>Nb+ $\alpha$ -Fe, and layered Nb-rich phases ensure a sound bonding between 304SS and the interlayer, as shown in Figure 6 [72].



**Figure 6.** SEM images of TC4/304SS laser joints with Nb interlayer [72] (a) cross-section of joint; (b) enlarged view of area B in region (a) and schematic of the line scan path; (c) line scan analysis data corresponding to (b); (d) enlarged image of area D in (b); (e) enlarged image of area E in (b); (f) enlarged image of area F in (b).

### 3.4.2. Ag Element Composition Interface

Deng and Sheng [90] utilized a silver interlayer for the solid-state diffusion bonding of commercial pure Ti with 304 stainless steel. Microstructural analysis unveiled that the resulting joints consisted of the remaining Ag interlayer, TiAg intermetallic phases, and a Ti-Ag solid solution. This approach yielded an optimized bond strength of 414 MPa.

Balasubramanian [91] employed a Box–Behnken design to optimize process parameters for silver foil as an interlayer. Microstructural examination revealed that with increasing temperature, grain length on the stainless steel side notably exceeded that on the Ti side. The diffusion bonding region displayed uniform, equiaxed grains. Grain growth can be attributed to recrystallization and small grains being encapsulated by larger ones. The presence of TiAg is linked to increased bonding strength and achieves a ductility of 18%.

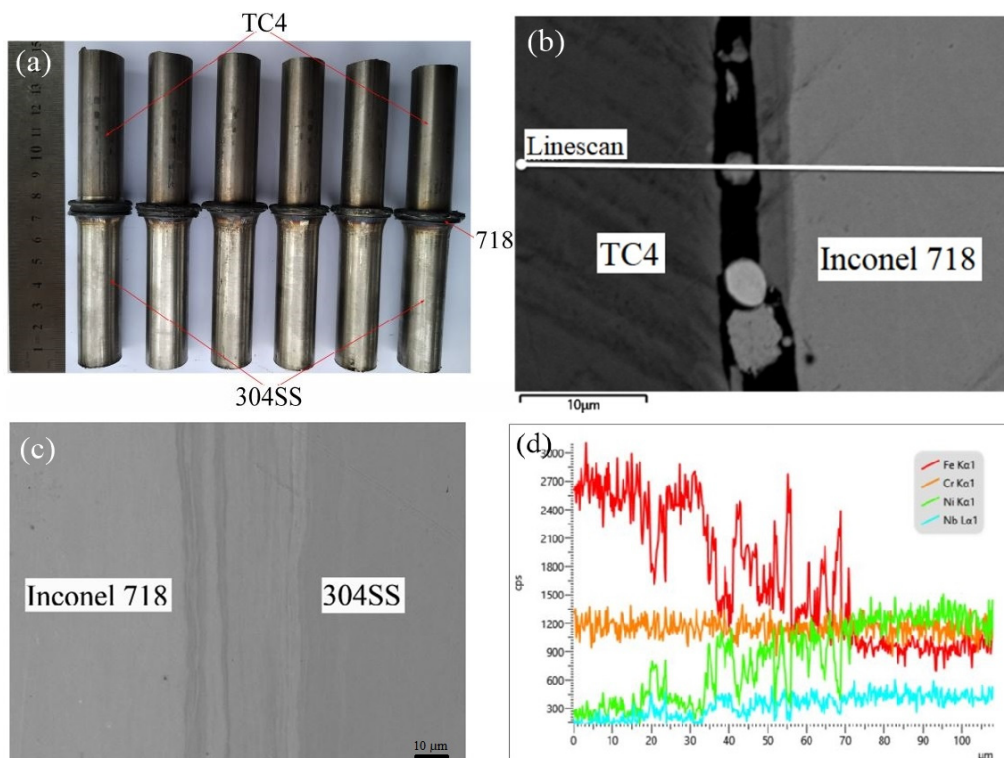
Compared to Cu, Ni, or Ni alloy, the Ag interlayer uniquely eliminates brittle phases at the interface. The formation of the TiAg phase between Ag and Ti facilitates their bonding. Optimal joint strength, up to 400 MPa, is achieved by bonding at 825–875 °C for 20 min with a 50  $\mu$ m thickness, resulting in a ductile fracture nature. A Cu interlayer significantly reduces FeTi and Fe-Cr-Ti phases, lowering joint hardness and brittleness. In line with the CuTi phase diagram, bonding between 890 and 960 °C introduces a small liquid phase in the CuTi joints, reducing diffusion bonding time and enhancing efficiency. Using a Ni alloy interlayer achieves the highest tensile strength of 560 MPa, with Ni<sub>3</sub>Ti, NiTi, and NiTi<sub>2</sub> phases forming sequentially at the Ni alloy and Ti interface after a 45 min hold at 900 °C.

### 3.4.3. Multi-Element Component Interface

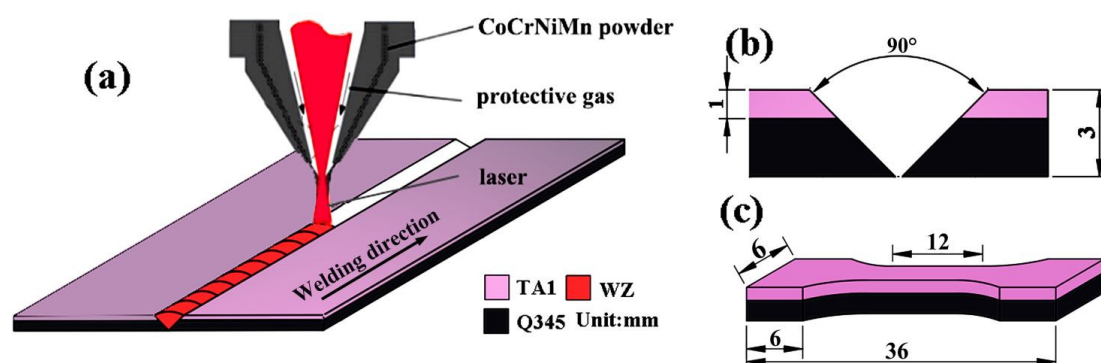
Vacuum brazing offers several advantages, including low heating temperatures, minimal residual stress in joints, and a diverse range of available solders. This method relies on the capillary action of the molten solder to connect and wet the base metal, making it particularly suitable for joining dissimilar high-melting-point metals that are challenging to weld. Given the limited mutual solubility of Ag and Ti, which precludes the formation of IMCs with Fe, Ag-Cu-based filler materials are preferred for brazing due to their favorable wettability to Ti and stainless steel at 800 °C. Yue et al. [9] utilized AgCuTi solder in a high-temperature short-term process (920 °C × 5 min) to achieve a shear strength of 188 MPa in the joint. The microstructure of the joint sequence was as follows: 1Cr18Ni9Ti/CuTi/Ag-rich zone + Cu<sub>4</sub>Ti/Cu<sub>4</sub>Ti/β(Ti)/Ti<sub>6</sub>Al<sub>4</sub>V. Upon increasing the temperature to 980 °C, the Ag-rich zone diminished, the Ti-Cu IMCs increased, and the strength decreased to 123 MPa. This highlights that appropriately conducted high-temperature short-term brazing can effectively suppress the formation of Ti-Fe IMCs. To reduce the occurrence of Ti-Cu IMCs and inhibit the diffusion of Ti atoms in the weld, some employed AgAl-based low-melting-point brazing materials, namely Ag-15Al-Mn-Si, Ag-5Al-Mn-Si, and Ag-8Al-Mn-Si, at 950 °C × 10 min, effectively impeding the diffusion of Ti atoms. When using Ag-8Al-6Sn-1Ni, Ni formed a solid solution with Fe, resulting in the presence of only an IMC reaction layer on the Ti side. However, specific joint strength data were not provided. It is worth noting that brazing, relying on capillary action to wet the base material, yields relatively low joint strength. Additionally, Ag-based brazing materials, though effective for certain specific joints, can be expensive and are less suitable for large-sized and structurally complex workpieces. Moreover, the requirement for overall heating of the workpiece during brazing can adversely affect the mechanical properties of the base material. The main difficulty in directly welding TC4 to 304SS is the formation of brittle Ti-Fe and Cr-Fe intermetallic compounds (e.g., Fe<sub>0.2</sub>Ti<sub>0.8</sub>, Cr<sub>0.7</sub>Fe<sub>0.3</sub>), which severely degrade mechanical performance; this issue can be resolved by introducing an Inconel 718 interlayer, which alters the reaction products to a more complex but less detrimental phase composition (including Fe<sub>2</sub>Ti, Ni<sub>3</sub>Ti, FeNi<sub>3</sub>, etc.), thereby suppressing the most harmful brittle phases and enhancing joint properties, as shown in Figure 7 [92].

The substantial difference in melting points between Cu and Ti poses challenges in forming weld seams. To address this issue, high melting point metal V can be introduced to Cu, and Cu along with V can serve as transition metals for composite plate transition layer welding. Scholars have explored the utilization of Cu-V flux-cored welding wires to create transition layers through the TIG method for composite plate transition layer welding. This approach effectively resolves issues related to the high welding temperature and penetration cracks associated with welding high-melting-point metals [93]. In the weld seam where Ti/Cu-V transition layer/Q345 steel coexist, scholars observed TiFe IMCs and cracks. The introduction of 25% V content led to the disappearance of cracks, and a V-based solid solution emerged in the transition layer. Using Cu30V flux-cored welding wire for transition layer preparation revealed the presence of FeTi at the Ti/Cu-V/Fe interface in the weld seam, with the main presence of β-Ti phase. Nanosized particles were distributed in Ti α-Ti, and FeTi phases, while Fe<sub>2</sub>Ti IMCs were mainly distributed in Cu-based solid solutions. However, the bending test indicated poor toughness, with an average bending angle of 5° [94]. Chu et al. [95] addressed the challenges of poor fluidity in the molten pool and splashing phenomenon by incorporating Ag into Cu-V flux-cored welding wire. Water cooling was employed to reduce the high-temperature residence time of the weld and welding heat-affected zone during the welding process. The temperature field of composite plate welding and the cross-sectional morphology of the transition layer prepared by filling Cu VAg flux-cored welding wire. The addition of a water-cooled

device effectively mitigated the appearance of TiFe compounds, enhancing joint strength to 511 MPa. However, the joint exhibited mainly brittle fracture characteristics. Simulation results demonstrated that after welding the Cu-V-based transition layer, the maximum transverse residual stress of the composite plate was 15.2 MPa, and the longitudinal residual stress was 15.3 MPa. Welding the Ti welding wire on the cover surface significantly increased stress, with the maximum transverse residual stress reaching 691 MPa and the longitudinal residual stress reaching 367 MPa. Stress concentration was observed at the Ti weld seam. The main difficulty in welding TA1 titanium to Q345 steel is the formation of a transition zone with defects such as incomplete fusion and cracks; this can be mitigated by using a CoCrNiMn multi-principal filler to create a high-entropy alloy structure in the weld metal, achieving a tensile strength of 293 MPa (60% of the base metal strength), as shown in Figure 8 [96].



**Figure 7.** (a) TC4/Inconel 718 alloy/304SS welded joints; (b) SEM image near TC4/Inconel 718 alloy interface; (c,d) SEM image near Inconel 718 alloy/304SS interface [92].



**Figure 8.** Experimental material diagram of TA1/Q345 bimetallic sheet welded joint with CoCrNiMn powder (a) TA1/Q345 bimetallic sheet welded joint; (b) the parameter of groove; (c) tensile test specimen [96].

#### 3.4.4. Conclusions

When an interlayer is introduced, element composition plays a critical role. A Nb interlayer (higher melting point than base metals) produces a hybrid joint with fusion welding at the Ti/Nb interface and diffusion bonding at the Nb/steel interface, achieving an ultimate tensile strength of 370 MPa without IMCs on the Ti side, though a reaction layer forms on the steel side. Vacuum hot rolling with Nb improves plasticity, but cracking occurs at 1050 °C due to Fe-Nb IMCs. A pure Ag interlayer in solid-state diffusion bonding gives a maximum strength of 414 MPa, with the joint consisting of residual Ag, TiAg, and Ti-Ag solid solution. Optimized parameters (825–875 °C, 20 min, 50 µm) yield ~400 MPa and ductile fracture. Compared to Cu or Ni, Ag uniquely eliminates brittle Fe-Ti phases. For multi-element systems, vacuum brazing with Ag-Cu-based fillers reaches 240 MPa at 860 °C/15 min, while AgCuTi solder gives 188 MPa. Higher Cu content increases Ti-Cu IMCs and reduces strength. Ag-Al-based solders suppress Ti diffusion, but brazing generally provides lower strength, is expensive, and requires overall heating that may degrade base metal properties. Cu-V-based flux-cored wires with 25% V eliminate cracks and form a V-based solid solution, yet the joint remains brittle. Adding Ag to Cu-V wire together with water cooling reduces TiFe compounds and raises strength to 511 MPa, though fracture is still predominantly brittle, and high residual stresses (up to 691 MPa) appear in the Ti cover layer. In summary, Nb and Ag interlayers offer moderate to good strength, while multi-element systems further increase strength (~511 MPa) but still face challenges in achieving fully ductile joints, controlling residual stress, and managing cost.

#### 4. Joints with Multi-Layers

It is important to highlight that a single intermediate layer can still lead to the formation of IMCs with one of the parent metals, posing challenges for achieving high-quality joints. In contrast, multilayer interlayers offer distinct advantages in the context of diffusion welding for Ti and steel. They facilitate superior interface control, resulting in stronger, more stable, and adaptable welded joints. This approach also mitigates stress concentration, enhances impact resistance, and improves overall quality control.

Kundu et al. [96] utilized composite Ni and Cu interlayers to bond duplex stainless steel with Ti6Al4V. Intermetallic phases including Cu<sub>4</sub>Ti, Cu<sub>2</sub>Ti, Cu<sub>4</sub>Ti<sub>3</sub>, CuTi, and CuTi<sub>2</sub> formed at the Cu and Ti interface. No intermetallic phases were detected at the stainless steel side and the Ni-Cu interface. Fracture occurred at Cu<sub>4</sub>Ti intermetallic, Ni<sub>3</sub>Ti, and  $\sigma$  phases with increasing bonding temperature, yielding a maximum shear strength of 377.4 MPa. Lee et al. [82] obtained a remarkably high strength diffusion-bonded joint of commercially pure Ti and super stainless steel without any detrimental phases using a V-Cr-Ni interlayer. The interface structure was Ti (base)/Ti solid solution/V/Cr/Ni/STS (base).

Li et al. [97] explored the impact of bonding time in the context of a multi-layer Nb/Cu/Ni interlayer at the optimized bonding temperature of 850 °C. They noted that initially, voids were present at the Cu-Nb to SS-Ni interface due to insufficient bonding time. With the progression of bonding time, strength gradually increased, reaching its maximum at 45 min, after which it sharply declined. This phenomenon is attributed to the improved mutual contact at the interface, the disappearance of voids, and increased Ni atom aggregation and diffusion, all contributing to enhanced bonding strength. However, with a bonding time of 60 min, the aggregation of Ni atoms diminishes, disrupting the reinforcing effect of the Cu-Nb solution and resulting in decreased bonding strength.

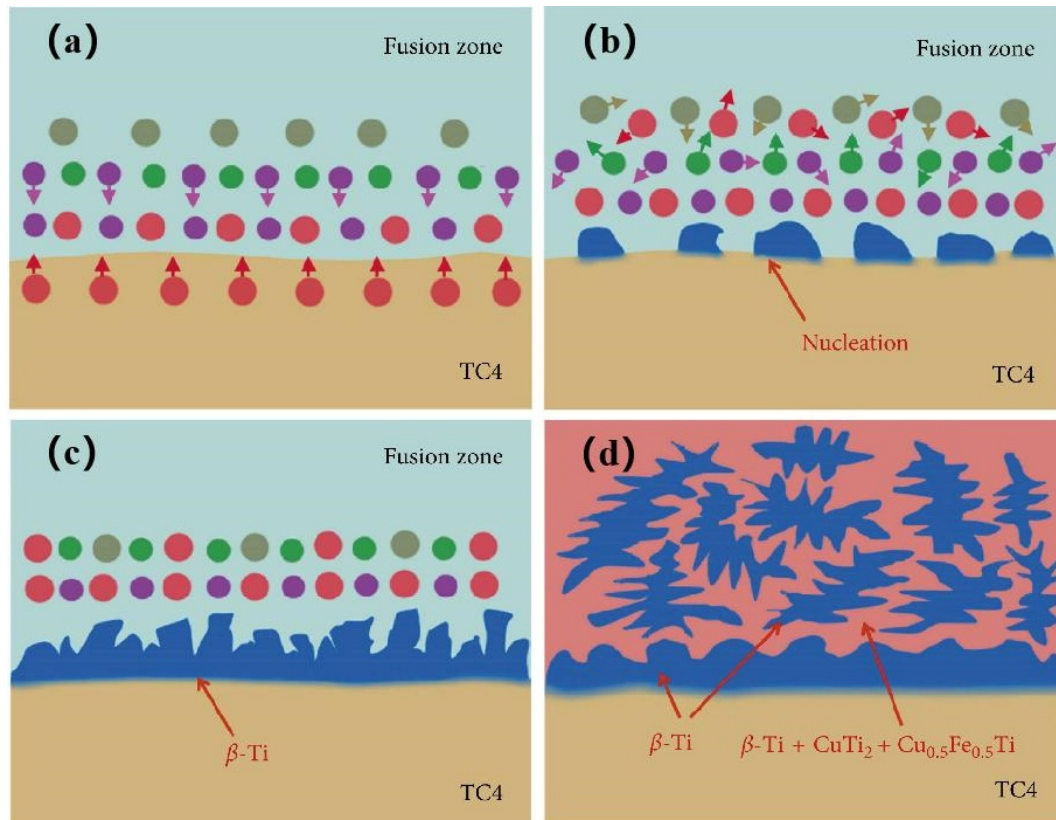
Song et al. [98] vacuum diffusion bonded Ti6Al4V and AISI316L using Cu/Nb interlayers. Optical images show joints at 850 °C, 900 °C, and 950 °C for 90 min. The  $\alpha$ - $\beta$  Ti layer thickened as temperature increased from 850 °C to 900 °C. The joints at 900 °C for 30, 60, and 120 min displayed a widening Ti layer due to increased Nb diffusion into Ti. Nb

acted as a  $\beta$ -stabilizing element in the Ti [97]. Slow cooling precipitated  $\alpha$ -Ti within  $\beta$ -Ti, allowing  $\alpha$ -Ti at room temperature. The joint bonded at 950 °C. The choice of a multi-layer interlayer is guided by two key principles. Firstly, it should prevent the formation of brittle intermetallic phases at the interface between the multi-layer and the parent metal. Secondly, the physical properties of the multi-layer interlayer should facilitate the transition between Ti and stainless steel, exhibiting qualities like good plasticity to alleviate thermal stress and a suitable thermal expansion coefficient matching the parent metal. In contrast to Cu/Ni and Cu/Nb multi-layer interlayers, the adoption of Nb/Cu/Ni and AlCoCrCuNi<sub>2</sub> multi-layer interlayers has demonstrated effectiveness in averting intermetallic phase formation and enhancing joint strength. Nevertheless, it is essential to note that more layers in the interlayer do not necessarily translate to better results. The decision to use a single-layer or multi-layer interlayer should be based on specific requirements and application scenarios, considering that multi-layer setups involve more intricate processes and greater technical demands.

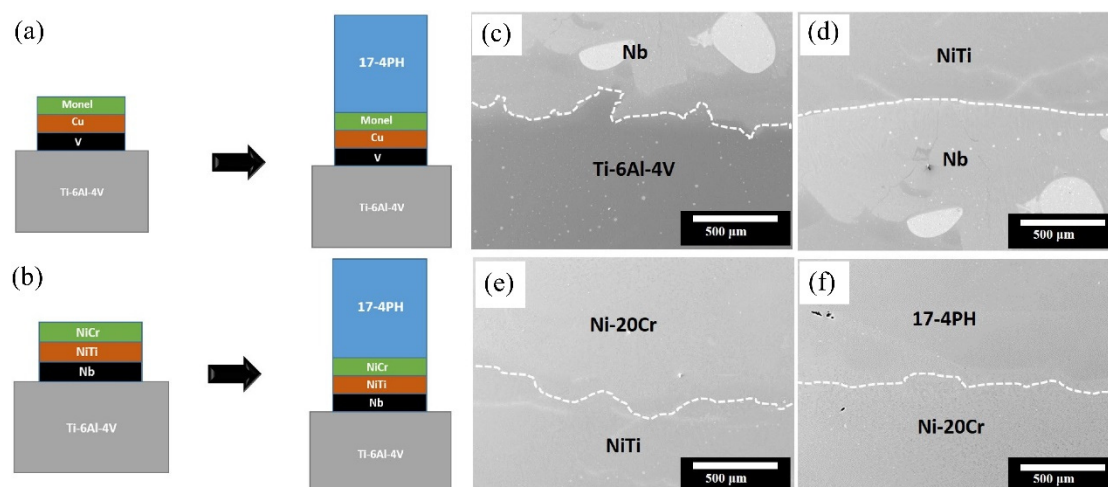
Niu et al. [99] studied the effect of electron beam welding on the microstructure and mechanical properties of NiTi/stainless steel joints, and found that the joint with FeNi interlayer added had excellent mechanical properties, with a fracture strength of up to 343 MPa. Wang et al. [40] conducted high-intensity electron beam welding using a V/Cu-based composite filler metal, joining  $\alpha$ -type Ti and austenitic 304 stainless steel. Their findings indicated that the feasibility of Cu/V filler metal was suboptimal, leading to non-fusion defects within the joint due to the emergence of brittle compounds like TiCu, TiFe, and  $\tau_2$ , particularly under low heat input conditions. However, the utilization of powder metallurgy V/Cu-V filler metal eliminated these defects, yielding a joint with an impressive tensile strength of 395 MPa, and remarkably low detection of IMCs in the Ti/V/Cu-V/Fe joint. Fang et al. [100] explored the application of Cu/Nb composite sandwich electron beam welding in joining Ti6Al4V and Kovar (Fe-29Ni-17Co) Alloys. They delved into the bonding mechanism, as illustrated in Figure 9a–d.

The utilization of multiple interlayers with strong compatibility with Ti-SS combinations serves as a promising strategy to prevent IMC formation and enhance joint strength to match the ultimate tensile strength (UTS) of these interlayers. A multi-pass narrow-gap welding approach, involving multiple interlayers of Ti (TA1), Cu (HS201), and Fe (ER50-6) fillers, was investigated for its impact on the transition zone in butt-welded CP-Ti/Q235B bimetallic sheets commonly used in pressure-bearing welded structures [20]. In comparison to gas tungsten arc welding (TIG) [101], this process significantly reduced the size of the transition zone and resulted in a 1.5–2 times smaller FZ. This reduction in FZ size led to decreased filler volume requirements and a consequent reduction in residual stresses. Li et al. [102] employed a magnesium–neodymium–zirconium (Mg-Gd-Y-Zr) alloy filler to perform coaxial dual-beam laser welding of Ti6Al4V and QP1 high-strength steel. It was observed that at the Mg/steel interface, the lattices of two compounds were linked through the sharing of Gd and Y atoms. In this region, the body-centered cubic (BCC) structured Mg<sub>24</sub>(Gd, Y)<sub>5</sub> compound acted as the heterogeneous nucleation core for  $\alpha$ -Mg. At the Mg/Ti interface, Zr, Y, and Gd atoms diffused into the Ti and formed a diffusion layer with a thickness of 35 nm. Zhang et al. [103] achieved a substantial improvement in the ultimate tensile strength of the Ti6Al4V/SS301L joint, reaching 627 MPa. This was accomplished through the strategic use of multiple interlayers composed of Ta/V/Fe materials, effectively preventing undesirable Ti-Fe intermixing. By focusing the two-pass laser beam on the Ta and Fe layers, they ensured that the V interlayer did not completely melt, thus contributing to the strength enhancement. The research indicates that Ti and Ta can form a body-centered cubic (BCC) solid solution, and similarly, Ti and V are also fully miscible. Within the FZ at the Fe-V interface, a uniform  $\gamma$ -Fe + (Fe, V) solid solution was observed,

which remained free from cracks and effectively prevented the formation of the brittle  $\sigma$ -Fe phase. A major difficulty in TIG welding Ti-6Al-4V to 17-4PH steel is the severe interlayer penetration of Fe and Ni into Cu and Monel alloys, along with nonuniform Ti diffusion along Nb and Ni-Ti layers, which can be detrimental to lamination; however, using carefully designed interlayer sequences (V/Cu/Monel400 or Nb/Ni-Ti/Ni-Cr) enables effective internal bonding without cracks, achieving tensile strengths up to 273 MPa, as shown in Figure 10 [104].



**Figure 9.** Formation process of the fusion zone of the Ti side (a) surface contact; (b) element diffusion and nucleation; (c) growth of grains; (d) formation of the solid solution and IMCs [100].



**Figure 10.** (a,b) Schematics of the fabrication strategy; (c–f) SEM microstructures of the morphology and cross-sectional image of Figure 10b [104].

Steel–Ti welds are plagued by the presence of brittle Fe–Ti IMC joints, leading to immediate joint failure. Achieving adequate joint strength has proven challenging when using single interlayers of Nb, Cu, Mg, and V elements, as each of these interlayers tends to promote the formation of brittle IMCs in either the Ti or Fe part of the joint. More promising outcomes have been observed when employing multiple interlayers of Ti: V: Fe, either through double-pass welding or via the hybrid explosion welding method, effectively preventing IMC formation [105].

## 5. Conclusions

The fundamental challenge in joining titanium and steel lies in their metallurgical incompatibility at elevated temperatures. As shown in the Fe–Ti binary phase diagram, Ti and Fe readily react to form brittle IMCs such as FeTi and Fe<sub>2</sub>Ti, which cause severe stress concentration, cracking, and premature joint failure. Additionally, Ti reacts with carbon in steel to form TiC and with atmospheric gases at high temperatures, further degrading mechanical properties. Direct fusion welding inevitably leads to uncontrolled mixing of molten Ti and Fe, producing large amounts of brittle IMCs and thermal stress-induced cracks, making it completely unsuitable for producing reliable joints. Even solid-state welding without an interlayer, while more promising than fusion welding, still suffers from IMC formation and Kirkendall voids, with a narrow optimal processing window and limited joint strength.

The most effective strategy to suppress brittle Fe–Ti IMCs is to introduce an intermediate layer that physically and chemically separates Ti and Fe. The choice of interlayer element is critical and must be guided by binary phase diagrams: suitable elements (V, Nb, Cu, Ni, Ag) are those that form solid solutions or less brittle IMCs with Ti and Fe. Cu interlayers are widely used because Cu–Ti IMCs (e.g., Ti<sub>2</sub>Cu, TiCu) are much softer than Fe–Ti IMCs, and Cu–Fe systems have negligible mutual solubility. Optimized Cu interlayers with diffusion bonding or laser welding achieve strengths of 320–523 MPa. V interlayers are attractive due to complete solid solubility with Ti ( $\beta$ -Ti, V) and with Fe (Fe, V) under rapid cooling, but the brittle  $\sigma$ -phase (Fe, V) can form if cooling is slow or V content is unfavorable; thus, high-energy beam welding with rapid cooling is essential. Ni interlayers form ductile Ni–Ti IMCs and offer excellent compatibility with steel, achieving the highest reported single-interlayer strength of ~560 MPa in diffusion bonding. Other elements like Nb and Ag also show promise (370 MPa and 414 MPa, respectively), but Ag is expensive, and brazing-based methods yield lower strength. Multi-element interlayers (e.g., Cu–V, Cu–V–Ag) can further increase strength (up to 511 MPa) but often suffer from residual stress and brittle fracture. The key conclusion is that no single element is perfect—each requires a specific welding process and tight parameter control to balance IMC formation, atomic diffusion, and residual stress.

For demanding applications requiring both high strength and ductility, single interlayers often fall short because they can still form brittle phases with one of the base metals. Multi-layer interlayers offer superior interface control by providing a graded transition in thermal expansion, chemical affinity, and mechanical properties. For instance, a Ti/V/Fe multi-layer achieved an impressive 627 MPa ultimate tensile strength by preventing direct Ti–Fe contact and forming only solid solutions  $\beta$ -(Ti, V) and (Fe, V) ss. The design principles for multi-layer interlayers are (1) each adjacent layer must be mutually soluble or form only ductile IMCs; (2) the overall stack should have gradually changing coefficients of thermal expansion to minimize residual stress; (3) the welding process must precisely control heat input to prevent complete melting of all layers. Key remaining challenges include eliminating residual stress, reducing cost, and achieving fully ductile fracture instead of brittle or mixed-mode failure. Future research should focus on computational design of interlayer

sequences, in situ process monitoring for quality control, and post-weld heat treatment strategies to relieve residual stress and further improve joint reliability.

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