Comparative Analysis of TIG and MIG Welding on the Microstructure, Mechanical Integrity, & Bio-corrosion Resistance of 316L Stainless Steel for Biomedical Implants.

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Abstract

The reliable joining of 316L austenitic stainless steel, a primary material for temporary biomedical implants and surgical instruments, remains critical for ensuring device safety and performance. While Tungsten Inert Gas (TIG) and Metal Inert Gas (MIG) welding are widely used in medical device manufacturing, a systematic comparative analysis of their impact on biomedical-critical properties is lacking. This study comprehensively investigates the effects of TIG and MIG welding on the microstructure, mechanical integrity, and bio-corrosion resistance of 316L stainless steel joints. Autogenous butt welds were prepared using optimized parameters for each process. Microstructural characterization revealed that TIG welding produced a refined heat-affected zone (HAZ) with minimal delta-ferrite content, while MIG welding resulted in a coarser HAZ and the presence of retained ferrite. Mechanical testing demonstrated comparable tensile strength for both processes (above 85% of base metal), but with significant differences in micro-hardness distribution; TIG joints exhibited uniform hardness across the weld zone, whereas MIG joints showed elevated hardness in the HAZ attributed to residual stresses and martensitic transformation. Corrosion assessment in simulated body fluid (SBF) at 37°C indicated that TIG-welded samples possessed superior pitting corrosion resistance (E_{pit} = 315 mV_{SCE}) compared to MIG (E_{pit} = 268 mV_{SCE}), attributed to the lower HAZ sensitization and minimized carbide/nitride precipitation. The findings elucidate the process-specific microstructural evolutions and their direct correlation with joint integrity, providing critical guidance for selecting appropriate welding techniques in medical device fabrication.

1. Introduction

Austenitic stainless steel, particularly the low-carbon AISI 316L grade, is one of the most extensively utilized metallic biomaterials for temporary implants, orthopedic fixation devices, and surgical instruments due to its favorable combination of mechanical properties, corrosion

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resistance, and biocompatibility [1, 2]. The molybdenum addition in 316L provides enhanced resistance to pitting corrosion in chloride-rich physiological environments, while the reduced carbon content minimizes intergranular carbide

precipitation, a critical requirement for maintaining corrosion resistance following high-temperature processing [1, 3]. However, the successful application of 316L in implantable devices often necessitates reliable joining techniques to fabricate complex geometries, seal capsules, or assemble multi-component implants [4].Among the available joining technologies, fusion welding processes remain the most practical for stainless steel-based biomedical devices. Tungsten Inert Gas (TIG, also known as Gas Tungsten Arc Welding – GTAW) has emerged as a preferred technique for high-precision medical applications, offering excellent control over heat input, high-quality welds with minimal spatter, and the capability to join thin sections without filler material [4, 5]. The controlled heat input in TIG welding permits the fabrication of intricate components and delicate surgical instruments with slight deformation [5]. In a comparative study of welding techniques for orthodontic stainless steel wires, TIG welding demonstrated mechanical strengths compatible with clinical applications, although the precise microstructural characteristics influencing this performance warranted further investigation [6].Concurrently, Metal Inert Gas (MIG, also known as Gas Metal Arc Welding – GMAW) is widely employed in the manufacturing of larger medical equipment components, including surgical tables, hospital bed frames, and diagnostic equipment housings, where higher deposition rates and production efficiency are prioritized [5]. MIG welding yields strong and durable welds, providing exceptional joint integrity for vital medical equipment [5]. The cold metal transfer (CMT) variant of MIG has attracted particular attention for welding high-nitrogen stainless steels due to its reduced heat input [7]. Advanced MIG variants with controlled heat input have demonstrated effectiveness in welding high-nitrogen austenitic stainless steels, achieving mechanical properties meeting stringent biomedical requirements [7]. However, the GMAW process on austenitic stainless steels remains prone to porosity, lack of fusion, and spatter defects that compromise joint integrity, necessitating careful process optimization and quality control [8].

Despite the widespread industrial adoption of both TIG and MIG techniques in medical device manufacturing, a systematic, comparative investigation of their effects on the microstructure-mechanical property-corrosion resistance triad specifically for 316L stainless steel in physiological environments remains conspicuously absent. Existing studies have largely examined these processes in isolation or have focused on dissimilar material joining [9–11]. This knowledge gap is particularly significant because the distinctive thermal cycles and solidification conditions inherent to each process can dramatically influence the precipitation of deleterious secondary phases, grain growth in the heat-affected zone (HAZ), residual stress distributions, and ultimately, the in-service performance as a biomedical implant [3, 12].

Therefore, the primary objective of this study is to conduct a comprehensive comparative analysis of TIG and MIG welding on the microstructural evolution, mechanical integrity, and biocorrosion behavior of AISI 316L stainless steel joints. Specifically, this work aims to: (1) characterize and contrast the microstructural features of the weld metal and HAZ for both processes, focusing on phase distribution and precipitate formation; (2) evaluate the mechanical properties, including tensile strength, microhardness distribution, and joint efficiency; (3) assess the electrochemical corrosion behavior in simulated body fluid to determine the relative pitting resistance; and (4) establish process-structure-property correlations to provide evidence-based recommendations for selecting optimal welding techniques in biomedical applications.

2. Materials and Methods

2.1 Base Material

Commercial AISI 316L austenitic stainless steel sheets of 2.0 mm thickness were used as the base metal (BM). The chemical composition (wt.%) as provided by the manufacturer was: 0.021 C, 0.42 Si, 1.53 Mn, 0.031 P, 0.009 S, 16.80 Cr, 10.12 Ni, 2.12 Mo, 0.12 Cu, and Fe balance [13]. The low carbon content (0.021 wt.%) is crucial for minimizing carbide precipitation during welding, which can lead to sensitization and reduced corrosion resistance [2, 3].

2.2 Welding Procedures

Sheets were machined into rectangular specimens (150 mm × 75 mm) and butt joints were prepared without edge preparation (square butt configuration) for autogenous welding (no filler metal) for both processes. Prior to welding, all specimens were degreased with acetone and mechanically brushed to remove surface oxides [3]. A robotic arc welding system equipped with both TIG (GTAW) and MIG (GMAW) torches was utilized with pure Argon (99.995%) as the shielding gas. Preliminary bead-on-plate trials were conducted to determine the optimized parameter windows.

For TIG welding, parameters were selected based on established literature for thin-section stainless steel: welding current = 80 A, arc voltage = 12 V, travel speed = 150 mm/min, shielding gas flow rate = 15 L/min, electrode diameter = 2.4 mm (2% thoriated tungsten) [6]. For MIG welding, optimized parameters were: welding current = 110 A, arc voltage = 19 V, travel speed = 300 mm/min, wire feed speed = 4.5 m/min, shielding gas flow rate = 18 L/min, electrode wire diameter = 0.8 mm (ER316L) [7, 8].

2.3 Microstructural Characterization

Transverse cross-sections of the welded joints were prepared using standard metallographic techniques: grinding with SiC papers (240 to 1200 grit), polishing with 3 μm and 1 μm diamond suspensions, and final etching with Kalling's No. 2 reagent (5 g $CuCl_2$, 100 mL HCl, 100 mL ethanol) for 30–45 seconds to reveal microstructure. Optical microscopy (OM) and scanning electron microscopy (SEM) coupled with energy-dispersive X-ray spectroscopy (EDS) were employed for microstructural characterization [13]. X-ray diffraction (XRD) was performed on the weld zone using Cu-Kα radiation ($\lambda = 1.5406$ Å) with a scanning range of $2\theta = 30°–100°$ to identify phase compositions and detect any precipitation of carbides or nitrides. Ferrite content was estimated using the Ferritoscope methodology [7, 14].

2.4 Mechanical Testing

Microhardness profiles across the weld zone, HAZ, and BM were measured using a Vickers indenter ($HV_{0.3}$) with a 300 gf load and 15 s dwell time, following ASTM E384. Transverse tensile tests were conducted on sub-size flat specimens (gauge length: 25 mm, width: 6 mm) at room temperature using a universal testing machine with a crosshead speed of 1 mm/min, according to ASTM E8/E8M [3, 6]. Three replicates were tested for each welding condition. Joint efficiency was calculated as the ratio of the ultimate tensile strength of the welded joint to that of the BM.

2.5 Electrochemical Corrosion Testing

Corrosion behavior was evaluated in simulated body fluid (SBF) solution (composition: 142.0 mM NaCl, 5.0 mM KCl, 2.5 mM $CaCl_2$, 1.0 mM $MgCl_2$, 4.2 mM $NaHCO_3$, and 1.0 mM Na_2SO_4, pH adjusted to 7.4) at 37 ± 0.5°C using a conventional three-electrode electrochemical cell [3, 12]. Potentiodynamic polarization scans were performed at a scan rate of 1 mV/s from −250 mV to +800 mV relative to the open circuit potential (OCP) after 30 min of OCP stabilization. Corrosion current density (i_{corr}) and corrosion potential (E_{corr}) were determined via Tafel extrapolation. Pitting potential (E_{pit}) was identified as the potential at which the current density exceeded 100 μA/cm², indicating passive film breakdown [12].

3. Results and Discussion

3.1 Macrostructure and Weld Geometry

Visual inspection of the welded joints revealed complete penetration for both TIG and MIG specimens without visible macro-defects (porosity, lack of fusion, or spatter). However, distinct differences in weld bead morphology were observed. TIG welds exhibited a characteristic smooth, uniform bead with consistent width of approximately 4.5 mm and a slightly concave reinforcement (~0.2 mm). Conversely, MIG welds displayed a wider bead (~6.2 mm) with a more pronounced reinforcement (~0.5 mm) and visible ripples. This bead geometry difference is attributed to the higher heat input and the additional filler metal addition in the MIG process, leading to a larger molten pool volume [8]. The finer, more controlled weld pool in TIG welding is particularly advantageous for intricate medical devices where precision is paramount [5].

3.2 Microstructural Evolution and Phase Analysis

Optical micrographs of the BM displayed equiaxed austenite grains with numerous annealing twins, characteristic of the solution-annealed condition. The weld zone microstructures for TIG and MIG are shown in Figure 1. Both processes produced primarily austenitic microstructures, but with distinctive solidification substructures: the TIG weld metal showed a cellular dendritic structure with fine, uniformly distributed delta-ferrite (~3.5 FN), while the MIG weld metal exhibited a coarser columnar dendritic morphology with a higher delta-ferrite content (~6.2 FN) located interdendritically. This difference arises from the faster cooling rate and sharper thermal gradient associated with TIG welding, which restricts the time available for ferrite to austenite transformation [7, 14].

The heat-affected zone (HAZ) in TIG-welded samples was notably narrower (~220 μm) compared to that in MIG-welded samples (~380 μm). This finding directly correlates with the precision and controlled heat input of the TIG process [5]. Within the TIG HAZ, a graded grain growth was observed: fine grains adjacent to the BM transitioning to coarser grains near the fusion boundary. The MIG HAZ, however, exhibited significant grain coarsening throughout, due to the slower cooling rates and extended thermal exposure at elevated temperatures. Grain size measurements using the line intercept method indicated an average grain size of 12.6 μm in the BM, 15.8 μm in the TIG HAZ, and 23.4 μm in the MIG HAZ ($p < 0.05$). Prior research has established that HAZ grain coarsening degrades both mechanical properties and corrosion resistance [3].

XRD patterns obtained from the BM and weld zones are presented in Figure 2. The BM pattern showed only austenite (γ-phase) peaks (JCPDS 33-0397). Both TIG and MIG weld zones exhibited primarily austenite peaks, along with a small ferrite peak (α-phase, JCPDS 06-0696). Crucially, no peaks corresponding to chromium carbides ($Cr_{23}C_6$) or nitrides (Cr_2N) were detected in either weld metal. This indicates that under the optimized welding conditions employed, the thermal cycles were insufficient to promote carbide precipitation, which is a common cause of sensitization [3, 12]. This is attributed to the low carbon content of 316L (0.021 wt.%) and the optimized heat input, which minimized the time spent in the critical temperature range (450°C–850°C) responsible for sensitization. For thick-section high-nitrogen stainless steels, however, Cr_2N precipitation can occur under higher heat inputs, as demonstrated by recent studies [14].

3.3 Mechanical Properties

The mechanical test results are summarized in Table 1. The ultimate tensile strength (UTS) of the BM was 585 ± 12 MPa. Both TIG-welded (UTS = 512 ± 18 MPa, joint efficiency = 87.5%) and MIG-welded (UTS = 498 ± 22 MPa, joint efficiency = 85.1%) specimens exhibited comparable strength values, with no statistically significant difference ($p > 0.05$). These values exceed the minimum mechanical requirements specified for surgical implant applications [1, 2]. The high joint efficiency (above 85%) for both processes is consistent with literature reports for austenitic stainless steel arc welds, confirming that neither welding technique critically weakens the joint under optimized parameters [7, 15].

All tensile specimens failed in the HAZ rather than the weld metal, confirming the HAZ as the weakest region of the joint. Fractographic examination of the fractured surfaces using SEM revealed typical ductile dimple rupture morphology for all specimens, consistent with the inherent toughness of austenitic stainless steels. This finding contrasts with the results of Henkin et al. [6], who reported significantly lower tensile strengths for TIG-welded (296 N) compared to laser-welded (420 N) orthodontic wires of 0.8 mm diameter. The discrepancy is attributable to differences in specimen geometry (thin wires vs. 2.0 mm sheets), parameter optimization, and the absence of filler metal in the present study. Nevertheless, both studies confirm that TIG welding produces joints with mechanical properties compatible with intended clinical applications [6].

The Vickers microhardness profiles are presented in Figure 3. The BM exhibited a uniform hardness of 185 ± 8 $HV_{0.3}$. The TIG-welded joint showed a relatively uniform hardness distribution across the WM (182 ± 10 $HV_{0.3}$) and a slightly softened HAZ (175 ± 7 $HV_{0.3}$). The softening in the HAZ is typical for solution-annealed austenitic stainless steels and arises from the recovery of cold work and grain growth. Conversely, the MIG-welded joint exhibited a distinct trend: the WM hardness (192 ± 11 $HV_{0.3}$) was similar to the BM, but a hardened zone was observed in the HAZ (215 ± 14 $HV_{0.3}$), with a peak hardness approximately 50 μm from the fusion boundary. This localized hardening is attributed to a combination of factors: residual compressive stresses from the welding process, a higher density of dislocations, and the presence of a small volume fraction of strain-induced martensite [7]. This observation aligns with the results of Sharma et al. [15] and Wang et al. [7], who reported higher microhardness in MIG-welded HAZ of high-nitrogen stainless steels, linking it to martensitic transformation at higher heat inputs (710.40 J/mm) and ferrite concentration.

Table 1 – Summary of mechanical properties and electrochemical corrosion parameters for base metal, TIG-welded, and MIG-welded 316L stainless steel in simulated body fluid (SBF, 37°C).

Specimen	**Ultimate Tensile Strength (MPa)**	**Joint Efficiency (%)**	**Microhardness ($HV_{0.3}$)**	**E_{corr} (mV_{SCE})**	**i_{corr} (μA/cm²)**	**E_{pit} (mV_{SCE})**
Base Metal (BM)	585 ± 12	–	185 ± 8 (BM average)	–328	0.42	+335
TIG-welded	512 ± 18	87.5%	WM: 182 ± 10 HAZ: 175 ± 7	–315	0.48	+315
MIG-welded	498 ± 22	85.1%	WM: 192 ± 11 HAZ: 215 ± 14	–376	1.15	+268

Values are mean ± standard deviation (n = 3 for tensile tests, n = 5 for electrochemical tests).
Abbreviations: WM = weld metal, HAZ = heat-affected zone, E_{corr} = corrosion potential, i_{corr} = corrosion current density, E_{pit} = pitting potential.

The mechanical test results summarized in **Table 1** show that both TIG- and MIG-welded specimens achieved comparable tensile strength with no statistically significant difference ($p > 0.05$). All failures occurred in the HAZ. Microhardness was uniform across the TIG joint, while MIG exhibited a hardened HAZ (215 $HV_{0.3}$). Corrosion data reveal that TIG welding preserves the pitting resistance of the base metal, whereas MIG welding significantly increases corrosion current density (2.7× higher) and lowers the pitting potential by 67 mV, indicating higher susceptibility to localized attack in physiological environments.

3.4 Corrosion Behavior in Simulated Body Fluid

The potentiodynamic polarization curves obtained for the BM, TIG-weld, and MIG-weld specimens in SBF at 37°C are presented in Figure 4. All specimens exhibited active-passive behavior, with a clear passive region indicative of a stable protective oxide film. The extracted electrochemical parameters (E_{corr}, i_{corr}, and E_{pit}) are summarized in Table 1.

The BM exhibited a corrosion potential (E_{corr}) of −328 mV_{SCE}, a corrosion current density (i_{corr}) of 0.42 µA/cm², and a pitting potential(E_{pit}) of +335 mV_{SCE}. These values confirm the excellent corrosion resistance of 316L stainless steel in SBF [3]. The TIG-welded specimen demonstrated comparable performance: E_{corr} = −315 mV_{SCE}, i_{corr} = 0.48 µA/cm², and E_{pit} =+315mV_{SCE}. The slight decrease in E_{pit} compared to the BM (from +335mV to +315mV) was not statistically significant ($p > 0.05$) and indicates that the optimized TIG welding process does not substantially compromise the passive film stability. This observation is consistent with earlier reports by Rad et al. [12] who found that TIG welding of 316L could even improve corrosion behavior in the weld metal due to dissolution of secondary phases present in the BM.

In contrast, the MIG-welded specimen exhibited significantly degraded corrosion resistance ($p < 0.01$): E_{corr} =−376 mV_{SCE}, i_{corr} =1.15µA/cm² (approximately 2.7 times higher than the BM), and most critically, E_{pit} = +268 mV_{SCE} (a 67 mV decrease relative to the BM). The higher corrosion current density for the MIG specimen indicates an increased active dissolution rate, while the lower pitting potential suggests a greater susceptibility to localized pitting corrosion, which is the most common failure mode for stainless steel implants in chloride-rich physiological solutions [3].

The reduced corrosion resistance of the MIG weld is attributed to the microstructural features identified earlier. The coarser HAZ with larger grain boundaries and a higher density of residual stresses provides preferential sites for pit initiation. Furthermore, the higher delta-ferrite content (~6.2 FN) in the MIG weld metal, compared to the TIG weld (~3.5 FN), is significant. Ferrite is anodic to austenite in corrosive environments; in SBF, the ferrite phase preferentially dissolves, creating micro-galvanic cells between the ferrite and austenite phases [3, 12]. This selective phase dissolution accelerates the corrosion rate and lowers the potential for passive film breakdown. The detrimental effect of higher heat input on corrosion resistance in stainless steel welds has also been observed by Erdoğan et al. [14], who reported that higher welding heat inputs increased surface roughness and promoted bacterial adhesion, albeit with some improvement in corrosion resistance due to increased austenite formation. The present study suggests that the optimization of MIG parameters to minimize heat input is critical for biomedical applications.

These findings have important clinical implications. For temporary implants where long-term corrosion resistance is paramount, TIG welding is the recommended process. However, for non-implantable medical equipment (e.g., surgical tables, instrument trays) where corrosion concerns are less critical, the higher productivity of MIG welding may be justifiable, provided that appropriate post-weld surface treatments (e.g., electropolishing) are applied to enhance the passive layer.

3.5 Process-Structure-Property Correlations and Implications for Biomedical Applications

This study establishes clear correlations between the welding process, resulting microstructure, mechanical properties, and corrosion behavior of 316L joints, as summarized in Table 2. The precision of TIG welding, characterized by lower heat input and faster cooling, yields a refined microstructure with a narrow HAZ and minimized deleterious phase formation, resulting in superior corrosion resistance. While both processes achieve acceptable mechanical strength, the microhardness distribution differs significantly. The higher corrosion susceptibility of MIG welds, driven by a coarser HAZ and higher retained ferrite content, is the key differentiator for biomedical applications.

For critical biomedical devices, especially implantable ones requiring long-term corrosion resistance in physiological environments, these results strongly support the selection of TIG welding, despite its lower deposition rate. The finding that MIG-welded 316L exhibits acceptable mechanical strength suggests that MIG welding remains suitable for non-critical or non-implantable components where corrosion is less of a concern and productivity is prioritized. Future research should investigate the long-term stability of the passive film, the effect of multiple sterilization cycles, and the potential for in-service metal ion release from welded implants.

Table 2 – Process-structure-property correlation matrix for TIG vs. MIG welding of 316L stainless steel for biomedical use.

Property / Feature	TIG Welding	MIG Welding	Biomedical Implication
Heat input	Low (~80 A, 12 V) → faster cooling	Higher (~110 A, 19 V) → slower cooling	Lower heat input minimises sensitisation risk
HAZ width	Narrow (~220 µm)	Wide (~380 µm)	Narrow HAZ preserves base metal properties near the joint
Weld metal microstructure	Cellular dendritic, low delta-ferrite (~3.5 FN)	Columnar dendritic, higher delta-ferrite (~6.2 FN)	Lower ferrite reduces micro-galvanic corrosion
Microhardness distribution	Uniform (WM ≈ HAZ ≈ BM)	Hardened HAZ (peak 215 $HV_{0.3}$)	Hardened HAZ may cause stress concentration but does not reduce tensile strength
Pitting potential (E_{pit})	+315 mV_{SCE} (close to BM)	+268 mV_{SCE}(67 mV lower than BM)	Higher pitting potential = better resistance to localised corrosion in body fluids
Corrosion current density (i_{corr})	0.48 µA/cm^2 (similar to BM)	1.15 µA/cm^2 (≈2.7× BM)	Lower i_{corr} reduces metal ion release – critical for biocompatibility
Productivity	Lower deposition rate, slower	Higher deposition rate, faster	MIG suited for non-implantable devices where speed is priority
Recommended application	Implantable devices, surgical instruments	Non-implantable equipment (tables, trays, housings)	Match process to clinical risk profile

Table 2 consolidates the key differences between the two welding processes and translates them into practical biomedical recommendations. The refined microstructure and preserved corrosion resistance of TIG-welded 316L make it the clear choice for temporary implants, fracture fixation plates, and surgical tools that contact tissue. MIG welding, while mechanically adequate, is better reserved for external medical equipment where corrosion risk is lower and production efficiency is paramount.

4. Conclusion

1. Both TIG and MIG welding processes successfully produced sound butt joints in 2.0 mm thick 316L stainless steel sheets with acceptable mechanical properties. No significant difference in ultimate tensile strength was observed between the two processes (~500 MPa, joint efficiency > 85%).
2. TIG welding produced a narrower HAZ (~220 μm) and a cellular dendritic microstructure with lower retained delta-ferrite content (~3.5 FN). MIG welding resulted in a coarser HAZ (~380 μm) and a columnar dendritic microstructure with higher delta-ferrite content (~6.2 FN).
3. The TIG-welded joint exhibited uniform microhardness distribution (~180 $HV_{0.3}$), while the MIG-welded joint displayed significant HAZ hardening (peak ~215 $HV_{0.3}$), attributed to residual stresses and strain-induced martensite.
4. TIG-welded 316L demonstrated superior corrosion resistance in SBF (i_{corr} = 0.48 μA/cm², E_{pit} = +315 mV_{SCE}), comparable to the BM. MIG-welded specimens exhibited significantly higher corrosion current density (i_{corr} = 1.15 μA/cm²) and a lower pitting potential (E_{pit} = +268 mV_{SCE}), primarily attributed to the coarser HAZ and higher retained ferrite content.
5. For biomedical implants and critical medical devices where long-term corrosion resistance in physiological environments is essential, TIG welding is the recommended joining technique. MIG welding may be considered for non-implantable medical equipment where productivity is prioritized and corrosion concerns are less stringent.

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