

Review

Dissimilar Welding and Joining of Cemented Carbides

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Abstract: Cemented carbides have been widely used in aerospace, biomedical/wearable sensor, automobile, microelectronic, and other manufacturing industries owing to their superior physical and chemical properties at elevated temperatures. These superior properties, however, make it difficult to process these materials using conventional manufacturing methods. In this article, an overview of the welding and joining processes of cemented carbide and steel is given, followed by a few examples of welding processes. Cemented carbides can be successfully joined by sinter-bonding, brazing and soldering, laser beam welding, tungsten inert gas (TIG) welding, diffusion welding, friction welding, electron-beam welding, and chemical vapor deposition. An overview of the benefits and drawbacks of brazing and soldering of cemented carbide and steel is presented, including reports on joint design, processes, and selection of brazing filler metals. The laser welding of cemented carbide and steel is addressed and reviewed, including reports on gap bridging ability, the inclusion/absence of filler metals, interlayers, and laser/TIG hybrid welding. Finally, a section is devoted to explaining the main issues remaining in the welding and joining of cemented carbide, corresponding solutions, and future work required.

Keywords: cemented carbides; dissimilar welding; sinter-bonding; brazing; laser welding; metal-inert gas welding; diffusion bonding; tungsten carbides

1. Introduction

Cemented carbide was patented by Karl Schröter in 1923 [1] as a composite material of a “soft” binder metal, usually cobalt (Co), nickel (Ni), iron (Fe), or a mixture thereof, and “hard” carbides such as tungsten carbide (WC), molybdenum carbide (Mo₂C), tantalum carbide (TaC), chromium carbide (Cr₃C₂), vanadium carbide (VC), niobium carbide (NbC), titanium carbide (TiC), hafnium carbide (HfC), or their mixtures [2–5].

The metallographic microstructure of cemented carbides includes tungsten carbide (α -phase), a binder phase (for example, based on Co, Ni, Fe) (β -phase), and a carbide with a cubic lattice (e.g., TiC, TaC), which may contain other carbides (e.g., WC) in solid solution (γ -phase) [6–8]. Murakami’s reagent can be used as an etchant for the identification of α -phases through attack of the WC phases. The etchant can reveal the microstructures and is a freshly prepared solution of equal quantities of

10–20% (mass fraction) aqueous solutions of potassium hexacyanoferrate (III) [potassium ferricyanide] and potassium or sodium hydroxide (10–20 g in 100 mL of water) [9]. The etching time is dependent on the WC grain size. In general, the etching time is advised to be in the range of 5–6 min and increases with the increasing grain size. To improve contrast, acid-based etchants, such as Nital (nitric acid and ethanol), dilute hydrofluoric acid (HF), ferric chloride (FeCl_3) solutions, or acidic mixtures such as aqua regia (mixtures of nitric and hydrochloric acids) are used as additional etchants to attack the binder phase.

Cemented carbides are widely used for high-speed cutting, printed circuit board drilling, rolling, and mining (die, rings, rolls, blades, slitters, rotors, stators) as hard metal or hard metal–steel composites. Moreover, the micro drills for printed circuit boards (PCBs), oil gas nozzle fuel pumps, and hydraulic components in jet engines, as well as automotive components (fuel pumps, fuel injectors, compressors, and valve trains), are fabricated from cemented carbides.

2. Weld Processes

2.1. Sinterbonding

Sinterbonding is an important method used to join cemented carbides to other materials. Sintering is the central process of powder metallurgy. Powders of most metals will sinter when heated to approximately three-quarters of their absolute melting points while protected from oxidation or other gaseous attack [10]. The metal flows viscously under the effect of surface tension and gas pressure [11]. Kuczynski et al. [12] suggests that the possible densification mechanism is (1) grain-boundary diffusion or (2) volume diffusion with grain-boundary sinks. Generally, extensive grain-boundary sliding occurs in most metals during deformation. In normal polycrystalline metals, such movements are very limited because of blocking of neighboring grains. In a compact of compressed metal powder, Jones [13] reported that neighboring grains were welded to each other at relatively few points and that, in the early stages of sintering, considerable rearrangement of the relative disposition of the powder particles under surface tension forces is possible by this mechanism [14]. Michalski and Rosiński [15] produced sintered composites of diamond particles in a cemented carbide matrix prepared by pulse plasma sintering (PPS) at high temperatures (Table 1). The transition layer consisted of a solid solution of carbon and tungsten in cobalt. The diamond particles were observed to be strongly bonded with the cemented carbide. Rodelas et al. [16] observed that sinterbonding WC–Co powder into a fully dense nickel-iron tungsten heavy alloy (WHA) by hot pressing yielded a consolidated interface devoid of porosity and voids. However, pressureless sintering is unsuitable because of debonding and void formation resulting from differential sintering shrinkage. Co-rich η -carbides form preferentially in regions of low carbon activity at the interface, which has been confirmed by a thermodynamic evaluation of η -carbides as a function of carbon activity. The proposed WC–Co/WHA bonds can be used to produce Friction Stir Welding (FSW) tools. Kitiwan et al. [17] fabricated WC and silicon carbide-coated (SiC-coated) diamond composites by spark plasma sintering (SPS) at 1199–1600 °C for 300 s under 130 MPa under vacuum. The WC–diamond (SiC) composite exhibited high hardness and fracture toughness. Michalski et al. [18] produced a WC–6Co/cBN composite at 1150 °C under a pressure of 100 MPa for 5 min using pulse plasma sintering (PPS). The melting cobalt film spread over the surfaces of the WC and Cubic Boron Nitride (cBN) grains, and strong bonds formed between them. The rapid heating rate was observed to lead to the generation of transient thermal compressive stresses in the WC grains and subsequent grain refinement. The suitability of tungsten/steel joints prepared by PPS has been demonstrated, and PPS is a promising method for fabricating components of divertors [19,20]. The microwave sintering [21] time is too short to prevent grain coarsening and is a useful technique to fabricate nano-sized cemented carbides. Partial transient liquid-phase bonding (PTLP) [22] was also used to fabricate WC–Co/40Cr steel joints. Maizza et al. [23] reported a new solid-state capacitor discharge sinter-welding (CDSW) process to obtain a WC–12Co/AISI M2 steel joint. The advantages are that short processing times help to prevent the coarsening of steel or WC

grains and WC decomposition and minimize microstructural defects. These findings lay a foundation for interface formation and surface science.

Table 1. Dissimilar cemented carbide bonds prepared by sinter-bonding methods. PPS, pulse plasma sintering; SPS, spark plasma sintering; PTLP, partial transient liquid-phase bonding; CDSW, capacitor discharge sinter-welding.

Hard Metals	Counterpart	Load (MPa)	Sintering Method	Temperature (°C)	Mechanical Properties (GPa)	Reference
WC–Co	Diamond	-	PPS	1000	23	[15]
WC–6Co	W–3.5Ni–1.5Fe	30	Uniaxial Hot-Pressing	1325	-	[16]
WC	Diamond (SiC)	130	SPS	1190–1600	30.5	[17]
WC–6Co	cBN	100	PPS	1150	-	[18]
W	Eurofer97 steel	-	PPS	1000	-	[19]
W–La ₂ O ₃	P91	-	PPS	800	-	[20]
WC	Co	-	Microwave Sintering	-	-	[21]
WC–10Co	40Cr	0.3	PTLP	950–1100	-	[22]
WC–12Co	AISI M2	-	CDSW	-	-	[23]
WC–20Co	Invar	-	Liquid-phase sintering	1350	-	[24]

As shown in Figure 1a, the sinter-bonding of Fe–36 wt.% Ni powder to WC–Co was performed at 1300 °C for 2, 8, and 16 h under vacuum [24]. The results indicate that the compacted Fe–36 wt.% Ni/WC–Co sinter-bonded powders yielded a consolidated interface comprised primarily of hexagonal α -WC, cubic Fe_{0.64}Ni_{0.36}, and the presence of the Co₃W₃ complex (Figure 1b), whose content increased as a function of holding time. Prolonged holding times promote bonding of the WC–Co/Fe–Ni component. However, excessive holding times result in significant contractions and failure by cracking. Normal grain growth (NGG) (Figure 1c) and abnormal grain growth (AGG) (Figure 1d) occurred through the WC/Co/WC interfaces, not only in the liquid phase, but also in the solid state, at lower temperatures. According to the results shown in Figure 1e,f, the abnormal grains at the interface exhibited much larger grain sizes than the normal grains. AGG observed at the WC–Co/Fe–Ni interface was characterized by a grain size of 6–9 μ m. The maximal abnormal grains were 6.5–8.5 times larger than the average grain size in the WC–Co base material in the vicinity of the interface. The concentration gradient, together with a high stress imbalance at the WC–Co/Fe–Ni interface, resulted in rotation, repacking, and rearrangement of the particles, which led to abnormal grain formation (Figure 1g).

The actual sintering temperatures are significantly lower than the melting points of cobalt or tungsten carbide because of the formation of intergranular films (complexion) at the WC/Co interface (Figure 1h). The nanosized powders are suggested to be the principal contributors to the initial rapid grain growth because of the inherent large surface areas and short diffusion distances.

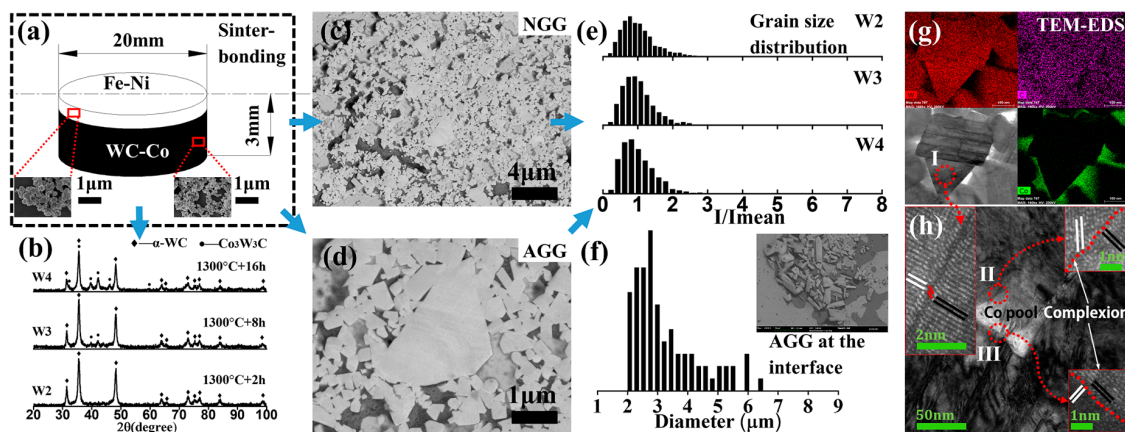


Figure 1. Sinter-bonding of WC–Co cemented carbide and Invar alloys. (a) Schematic illustration of sinter-bonding process. (b) X-Rays Diffraction (XRD) patterns of sinter-bonded joints. (c) Normal grain growth. (d) Abnormal grain growth. (e) Grain size distribution for specimens far from interface and (f) at the WC–Co/Fe–Ni interface. (g,h) Bright-field Transmission Electron Microscope (TEM) photographs using FEI Titan Themis 200 TEM microscope (FEI, Hillsboro, OR, USA) and high-resolution TEM photographs showing the WC crystal lattices and the formation of intergranular films (complexion) at the WC/Co interface. NGG, normal grain growth; AGG, abnormal grain growth. EDS, energy-dispersive X-ray spectroscopy. Adapted from [24], with permission from Elsevier, 2019.

2.2. Brazing

2.2.1. Background

Brazing is one of the cheapest, most reliable, and efficient ways of making dissimilar joints, which is one of the most important methods to join cemented carbide and steel. After fabricating a cemented carbide, Schröter [25] patented the brazing method of producing a cemented carbide/steel joint. The evolution of the cemented carbide joining method had the milestone significance in the history of cemented carbides' development. Table 2 summarizes the brazing processes and characteristic features of the various types of brazing filler metals.

Table 2. Chemical composition of filler metals and their brazing conditions and temperatures. OFHC, oxygen-free high conductivity; GTA, gas tungsten-arc.

Composition	Brazing Conditions	Thickness (mm)	Body Material (1)	Body Material (2)	Temperature (°C)	Reference
Cu-Borax/Mo/Cu-Borax/AuNi, Silver, AgCu, Copper	-	-	Hard metal	Steel or iron	1100	[25]
WRe	Vacuum	0.127	WC-6Co	4340	810–1100	[26]
CuAg, OFHC Copper	GTA braze	0.25	GE-15	-	-	[27]
CuMnCo	Sinter, hydrogen	-	WC-TiC/TaC-Co	-	1400	[28]
AgCuP	Induction, Argon	0.2/0.3	B30	40HM	1040–1120	[29]
AgCuZnCd	Flame, flux	-	WC	steel	470–725	[30]
AgCuZnCd	Flame, flux	-	W-Co-Ti	AISI 4145	710–980	[31]
Zinc, AlSi-alloy	Ultrasound, flux	0.2	WC-15Co	Be-Cu	640–750	[32]
AgZnCuNiMn	Ultrasound, fluxless	-	K10	T11302	-	[33]
AgCu	GTA braze	-	WC-10Co	AISI 1020	1300	[34]
AgZnCuNi	Vacuum	0.8/2	WC-8Co	SAE1045	1100/850	[35]
CuZn(Ni)	Vacuum	0.3	WC-10Co	90MnCrV8	850	[36]
CuZn	-	-	WC-Co/Ni	410	-	[37]
CuZnNi	Vacuum	0.2	WC-8Co	3Cr13	1060–1100	[38]
CuNi	Vacuum	-	WC-TiC-Co	-	940–960	[39]
CuNi	Vacuum	-	WC-Co	Carbon steel	-	[40]
CuMnZn	Vacuum	0.1/0.04	WC-8Co	S45C	1050	[41]
AgNi/CuZn/AgNi	Vacuum	0.2	WC-20Co	16Mn	940–980	[42]
	Induction, flux	0.12	WC-15Co	35CrMo	710–770	[43]

2.2.2. Joint Design

Joint design strongly affects the cooling rates of the weld metal and the heat-affected cemented carbides. Joint design includes the section thickness, arrangement of the seams, geometry of the parts, welded joint geometry, and restraint of the welded joint. These variables determine the weldability and the ease of fabrication of as-welded cemented carbides.

Figure 2a,b present images of a digging machine and a drag pick. The cross-sectional view along section I–I at the top-right corner shows the brazed seam and joint geometry. The microstructures of the cemented carbide base in region “C” in Figure 2b and the grain size distribution are presented in Figure 2c,d, respectively. Compared with as-welded cemented carbide with a high cobalt content, the magnified microstructures in Figure 2e of region “E” (Figure 2b) reveal a much narrower intermediate layer between the cemented carbide and fusion zone. Figure 2f shows the heterogeneous microstructures in the fusion zone, where tungsten carbide exhibits a type of delicate mutual effect with multi-alloys. The further magnified microstructures have cellular structures (Figure 2g). The electron back-scattered diffraction (EBSD) (Hitachi S3400 N SEM with an HKL-EBSD, Tokyo, Japan) patterns in Figure 2h–j show the microstructures, orientation maps, and Smith factor. The proposed joint design (circular sandwich structure: WC–Co/Cu–Ag/9SiCr) was concluded to be helpful for minimizing residual stresses.

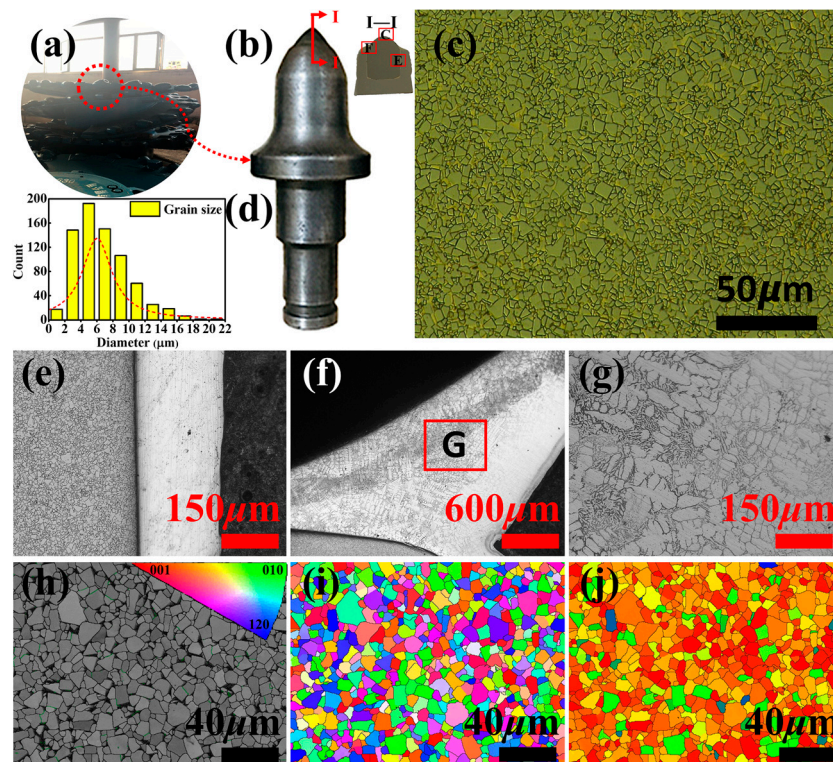


Figure 2. Joint design and example of brazing of WC–Co cemented carbide and steel. (a) Digging machine. (b) Drag pick. The joint geometry is shown in the top-right corner. (c) Microstructures of WC–Co cemented carbide (region “C” in Figure 2b). (d) Coincidence site lattice (CSL) grain boundaries. WC–Co/Cu–Ag/9SiCr joint (e) in region “E”, (f) in region “F” in Figure 2b, and (g) in the fusion zone in region “G” in Figure 2f. (h) Electron back-scattered diffraction (EBSD) pattern; the top-right corner shows the crystal preferred orientation. (i) EBSD orientation maps of heat-affected WC–Co. Step size: 150 nm. The colored areas are binders and tungsten carbides; different colors indicate different orientations (Euler angles) and grains. (j) Smith factor.

2.2.3. Processes

As shown in Table 2, tungsten and its alloys can be successfully brazed by vacuum brazing [26], gas tungsten-arc (GTA) braze welding [27], sintered brazing in hydrogen [28], induction brazing with

protective argon [29], flame brazing with flux [30,31], and ultrasound-assisted induction brazing with flux [32] and without flux [33].

Thorsen et al. [28] discussed treatments to improve wettability while brazing cemented carbides sintered in hydrogen furnaces with CuAg and oxygen-free high conductivity (OFHC) copper as filler metals. The results revealed the wettability mechanism during brazing of cemented carbide. Pieczara et al. [29] presented the brazing technology of B30 grade cemented carbide and 40HM steel with filler metal Cu87Mn10Co3. The experimental joints were made with sandwich structures using distance rods, steel mesh, and nickel mesh. The research provides guidance for the repeatability of the geometry of the brazing process. Amelzadeh et al. [31] investigated the effect of the brazing filler metals on WC–Co/steel joints. Additive nickel with a Na2B4O7 flux was observed to improve the strength of the joints. Although the use of flux enables successful joining, it also generates voids within the joint, which reduces the strength of the connection. Tillmann et al. [33] instead explored the feasibility of brazing of cemented carbides to steel without a flux. The selected filler metals, Ag449, pure Zn, and an AlSi-alloy, successfully wetted both materials and led to a dense connection.

2.2.4. Selection of Brazing Filler Metals

Filler metals used for the brazing of cemented carbide included Zn-based, Ag-based [34–36], and Cu-based [38–42] alloys, as well as “sandwich structure” alloys [43]. The main drawback of using filler metals containing Silver (Ag) and Zinc (Zn) is the high potential of evaporation during soldering. For this reason, in the most recently used filler materials, Zn is replaced by Stannum (Sn), Copper (Cu), Nickel (Ni), Phosphorus (P), or Manganese (Mn), sometimes accompanied by small amounts of Indium (In) or Sn, to decrease the melting temperature [30]. Cadmium (Cd) has been used to reduce the brazing temperature and ensure a strong joint. Because of the restriction of using Cd, higher temperatures are needed to produce sufficient joints with other filler metals; however, the process became more complex by handling the residual stresses [36]. Cu reduces the stress levels because of its higher ductility. Ni is an element that enters into complete solid solution with cobalt (Co) in tungsten carbide. Chen et al. [37] fabricated a new functionally graded WC–Co/Ni component (FGWC) with an Ni layer on the joining surface, which was designed to improve the wettability of solders on the cemented carbide and relax the residual stresses [38]. This research lay the foundation for the joining of cemented carbide and steels by introducing a functionally graded nickel wetted WC–Co/Ni component. Lee et al. [40] patented a novel insert metal consisting of stacked (or double layered) Cu alloy and amorphous Ni (MBF80) alloy. Increased amounts of Cr₃C₂ were observed to be effective in preventing the formation of the η phase (Co₃W₃C) and the coarsening of WC grains. Under the same brazing conditions, the maximum shear strength was higher than that of a similar joint without chromium carbide [41].

Ag-based braze alloys result in excellent wettability of cemented carbide and steels, while the Co is dissolved in the Ag-based filler metals [44]. Triantafyllou et al. [45] reported on a series of wetting experiments on systems of pure Ag as well as Ag–Cu–Ti and Ag–Cu–Ni pseudo-alloys in contact with widely used steels and ceramics to determine the interfacial properties of the above systems. Pure Ag in air exhibited poor wettability with all the substrates with contact angles greater than 90°. The Ag–Cu–Ti pseudo-alloy in vacuum exhibited improved wettability. The Ag–Cu–Ni pseudo-alloy in air exhibited excellent wetting properties with contact angles of less than 10°. Ag-based braze alloys are also used in dissimilar cermet brazing. Feng et al. [46] reported on a TiC cermet/Ag–31Cu–23Zn/steel joint formed by vacuum brazing. Increasing the brazing temperature or time was observed to decrease the amount of Ag-based solid solution in the middle of the braze alloy.

Laansoo et al. [47] investigated the weldability of cermet (WC-15Co, TiC60/FeNi, TiC50/NiMo, CrNi10, CrNi30) and steel (S45C, X10CrNi18-8) with 50–150 μ m thick filler foils between the cermet–steel parts prepared by induction brazing heated for 1 min under a pressure of 2–3 MPa. Northrop [48] reviewed several braze alloys available for the brazing of cemented carbides to steel. The brazing metals ranged from silver solders of low melting point to pure copper, with brazing temperatures

from approximately 600 °C to over 1050 °C. The strength of the joints strongly depended on the braze parameters [49].

2.2.5. Advantages and Application

The main brazing techniques employed are torch brazing, induction brazing, and furnace brazing [48]. To ensure slow and even heating and cooling, large structures should be furnace-brazed; however, no technique can entirely eliminate the thermal stresses because of the differential contraction that occurs as the assembly cools from the brazing temperature to ambient conditions. Smaller components can be torch-brazed; however, most components are mass produced by induction brazing, which is a clean, simple, and rapid method.

Tungsten-based alloy and its welding lay a foundation for the application of dissimilar joints of cemented carbide. Farrell et al. [50] and Lessman et al. [51] investigated the welding issues and weldability of tungsten-based alloys. They showed that cracking is caused by the growth and coalescence of small pores on the grain boundaries.

2.3. Laser Welding

2.3.1. Background

Laser is an acronym for light amplification by stimulated emission of radiation (LASER). Albert Einstein established the theoretical foundations for the laser by optimizing Planck's law of radiation and the probability coefficients for the absorption, spontaneous emission, and stimulated emission of electromagnetic radiation in 1917 [52]. A laser is made up of a laser medium (the material in which the laser is generated), a pump source (provides the necessary activation energy to the laser medium), and a laser resonator.

The application of a laser on welding can decrease the distortion with lower heat input and high efficiency and automation [53,54]. Laser welding is a competitive, high-efficiency process with high user satisfaction and has been applied in a wide range of fields for steel, aluminum, titanium [55], and magnesium alloys [56]. Laser welding is also a potential method to fabricate dissimilar WC–Co/steel joints. The properties of laser-welded hard metals were considered to be sufficient for the fabrication of valves [57] or drag picks to be used in production in the rock mining industry or for oil and gas applications.

2.3.2. Without Filler Metals

Sandig et al. [58] reported on as-welded hard metals joints prepared by a laser and opened the doors to further development of the use of lasers on hard metal. Tian et al. [59] fabricated a dissimilar joint by wetting hard metal with molten steels and revealed the mechanism of dip soldering. The plasma on the specimen surface increases the metallurgical bonding. Miranda et al. [60] reported the application of the CO₂ laser, neodymium/yttrium aluminum garnet (Nd/YAG), and fiber laser and optimized the welding parameters. Nd/YAG (neodymium/yttrium aluminum garnet) was observed to produce better as-welded joints. The research indicated that cobalt volatilization and overheating of the cemented carbide were prevented by focusing the laser spot at the steel side.

Figure 3 presents microstructures in a fusion boundary region in the joint prepared by laser welding without filler metals. The research also indicated that laser is a potential technology for joining cemented carbide and steel. The parameters, including heat input (laser power, scan speed), laser spot position, defocusing amount, and local preheating by a “dry-run” of the laser scan, were observed to affect the bead formation of a 2 mm thick plate. Within the range of parameter variations [61], for a constant heat input, it appears that increasing the laser power has a greater effect on enhancing the penetration and the bend strength than decreasing the laser scan speed.

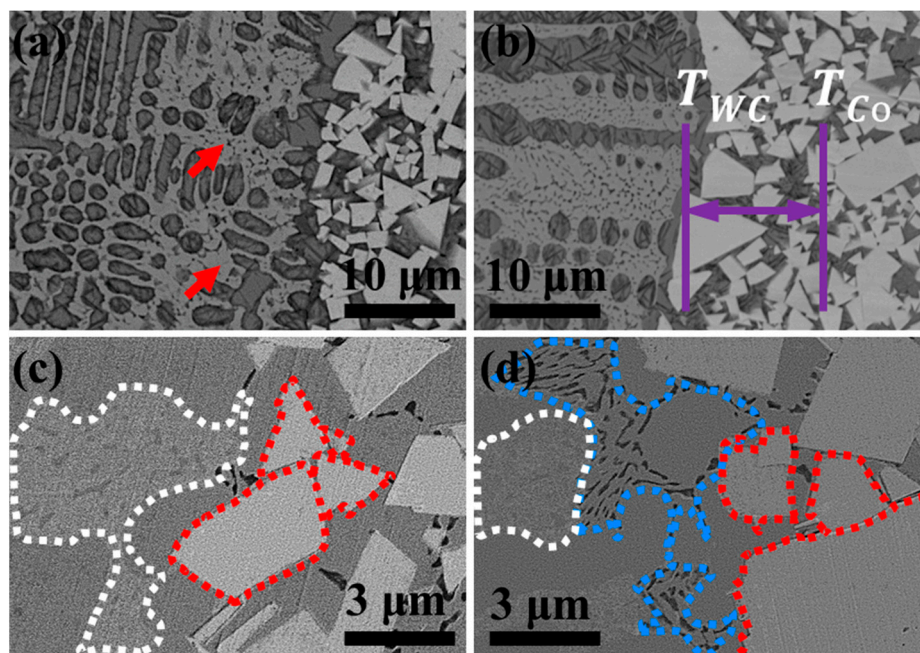


Figure 3. Laser welding of WC–Co cemented carbide and S45C carbon steel without filler metals. (a) Microstructures in a fusion boundary region. The red arrows show the mixed carbides. (b) The higher temperature border (T_{WC}) and lower temperature border (T_{Co}) of the fusion boundary region. (c), (d) Microstructures and grain growth in a fusion boundary region.

At a higher magnification in the literature, the microstructure (Figure 3a) consists of a fusion boundary region that is several WC grains wide ($\approx 30 \mu\text{m}$) and contained more mixed carbides. The higher and lower temperature border of the fusion boundary region are defined by the melting point of tungsten carbide and the cobalt (Figure 3b). The microstructure in the Figure 3c,d consists of half-dissolved WC (enclosed by red dotted lines), iron-rich face-centered cubic (FCC) (enclosed by white dotted lines), face-centered cubic (FCC) cobalt-rich solid solution, mixed carbide, and eutectic microstructure (enclosed by blue dotted lines). Brittle fracture during bending occurred along the fusion boundary and heat-affected zone (HAZ) on the cemented carbide side, where dissolution of WC and penetration of Fe from the fusion zone are believed to have caused embrittlement at the WC–matrix interfaces. Martensite was formed in the fusion zones and exhibited different morphologies in the center of the fusion zone, near the cemented carbide, and on the steel side; the morphologies were affected by the carbon [61,62].

Another study on the laser welding of cemented carbide without filler metals by Costa et al. [63] indicated that the weldability increased with increasing cobalt content, because of its effect in improving the ductility and weldability. Pre-heating, post-heat treatment, and nitridation were observed to be helpful to ensure joint integrity, increase the strength, and minimize the residual stresses [64]. Tungsten was observed to diffuse to the bead, inducing multiple W carbide formation in the martensite structure and contributing to higher bead hardness.

Figure 4a–d present a second electron (SE) micrograph and back-scattered electron (BSE) micrographs of laser-welded super-fine cemented carbide (WC–TiC–TaC–Co) and stainless steel without filler metals. Although black titanium carbide grains were observed at the interface, normal grain growth and abnormal grain growth were observed (Figure 4e–h).

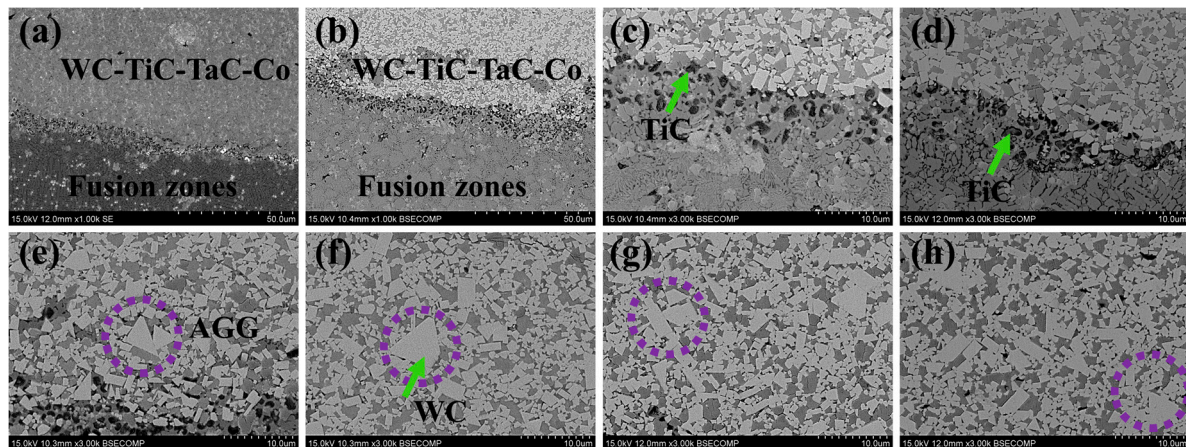


Figure 4. Laser welding of super-fine cemented carbide (WC–TiC–TaC–Co) and stainless steel without filler metals. (a) Second electron (SE) micrograph and back-scattered electron (BSE) micrographs (b,c) at low magnification and (d) high magnification showing the WC–TiC–TaC–Co/316L stainless steel interface. (e–h) Normal grain growth and abnormal grain growth (AGG) of WC grains near the interface.

2.3.3. Buffered with Interlayers

Guillaume first discovered the Invar effect in 1897 [65]. The Invar alloy has a low thermal expansion coefficient over a wide temperature range [66,67]. Yu et al. [68] and Yao et al. [69] utilized an Invar insert as a filler metal to join WC–Co cemented carbide and carbon steel prepared by fiber laser welding. Figure 5 shows the laser welding of cemented carbide and steel with Invar as the interlayer. Figure 5a shows the robotic welding systems used in the present search. The architecture of the preparation of the WC–20Co/Invar/AISI 1045 welded joint consisted of the following: the dimensions of the base metal were $\phi 58 \times 3$ mm, and the Invar interlayer was 0.5 mm thick. After welding, the as-welded WC–Co/Invar/AISI 1045 specimens were evaluated using X-ray nondestructive testing (Figure 5c). X-ray film maps (Figure 5d) were analyzed using image processing methods (Figure 5e).

The as-welded microstructures of the Invar fusion zones contained FCC primary dendrites γ (nickel dissolved in γ -Fe), eutectic colonies, and mixed carbides. Figure 5f,g show the cellular grains (white arrows) and grain boundaries (black arrows) between them, respectively. The Invar dendrites may have further transformed from the bcc crystal structure to bcc $\rightarrow$ fcc constituents upon cooling [18]. Iron atoms play a key role in the formation of dendrites and eutectics. Driven by the carbon depletion mechanism (Figure 5h), with decreasing welding speed and increasing heat input, it is evident that the mixed carbides exhibited coarser eutectics and a higher value closer to the fusion boundary on the WC–Co side.

Figure 5i shows the interfacial microstructures near the WC–Co/Invar interface. The areas surrounded by red, blue, and white dotted lines indicate dissolving WC grains, the eutectic microstructure, and the iron-rich solid solution, respectively. Figure 5j shows the effect on the bend strength for different levels of welding speed. The fracture propagation paths for the face bend-tested specimens are shown in Figure 5k. The 3 mm thick as-welded WC–Co/Invar/AISI 1045 specimens with lower heat input exhibited higher bend strength. Some specimens exhibited transgranular cracking at the WC facet (Figure 5l,m). Cracks were observed to initiate from the HAZ, propagate to the WC–20Co base metal, and finally fail in the HAZ near the cemented carbide. The fracture of the other specimens was initiated from the fusion zone. In addition, the fractures occurred in the fusion zone and propagated to the HAZ near the WC–Co cemented carbide. Dissolved and half-dissolved WC was observed at the fracture surface (Figure 5n), with the red arrows indicating the dissolved WC grains; cracks were also observed along the grain boundaries (Figure 5o). The proportional limit [70] was also concerned.

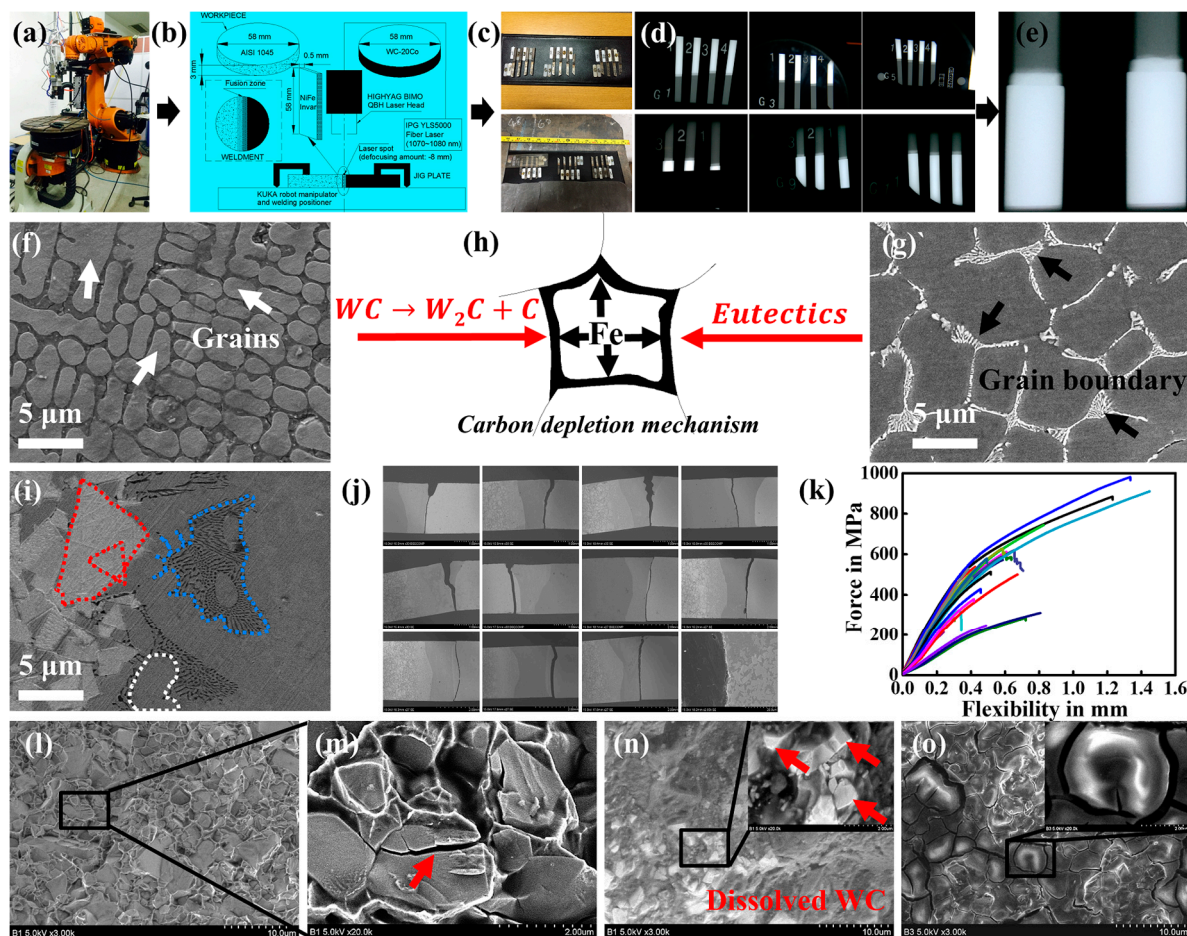


Figure 5. Laser welding of WC–Co cemented carbide and AISI 1045 steels. (a) Robotic welding systems for welding of cemented carbide and carbon steel. (b) Architecture of the preparation of WC–20Co/Invar/AISI 1045 welded joint: dimensions of $\phi 58 \times 3$ mm base metal, interlayer: 0.5 mm thick Invar. (c) X-ray nondestructive testing (NDT) of as-welded specimens. (d) X-ray detection film maps. (e) Image processing showing the weld defect. (f,g) Grains and grain boundaries in the fusion zone. (h) Carbon depletion mechanism during welding. (i) Microstructures showing the grain growth and mixed carbide near the WC–Co/Invar interface. (j) Bend strength. (k) Face bend-tested specimens showing the fracture propagation path. (l–o) Bend fracture surface showing the interdentritic fracture in the fusion zone and transgranular fracture in the heat-affected zone (HAZ). Adapted from [68], with permission from Elsevier, 2016.

As shown in Figure 6a, the as-welded fusion zone exhibited larger plastic deformation than the Invar base, which led to an increase in local strain fields and contributed to the dislocations (the white rectangle showing the deformation). Slip deformation indicated by the white line in Figure 6b–d was considered to be a contributing deformation mechanism for the as-welded Invar Fe–Ni alloys. Bright-field TEM image and high-resolution images show the dislocations and crystal lattice. Crystal plane (111) is the priority.

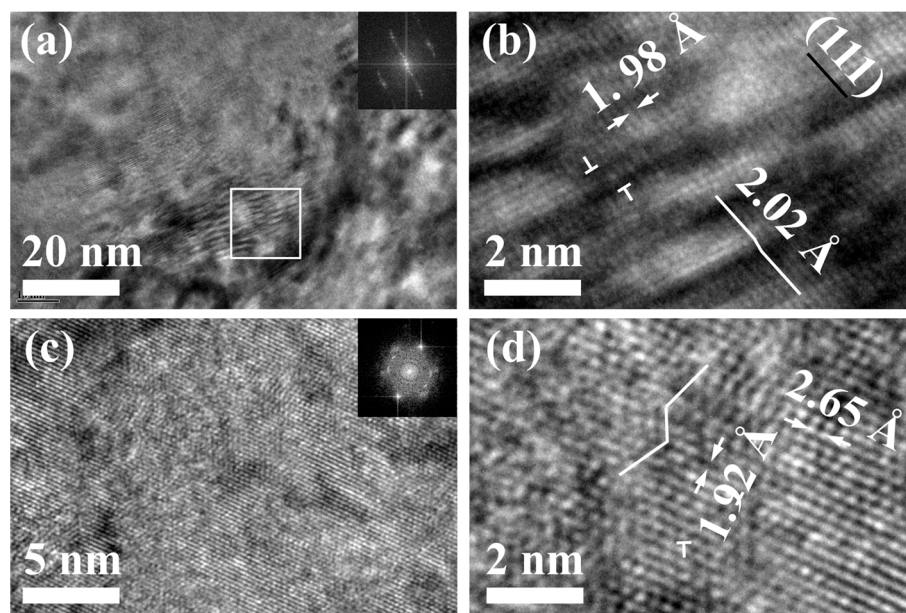


Figure 6. A typical (a) TEM image and (b) high-resolution TEM image of the as-welded Fe–Ni samples with a Fourier transformation of the white rectangle at the top-right corner shows the slip deformation, and the inverse Fourier-filtered image from inside the white rectangle (c) in (a), and (d) in (b) shows slip deformation, Stacking Faults (SFs) (white lines), and dislocations (white T). Adapted from [69], with permission from Elsevier, 2017.

Mirski et al. [71] investigated and reviewed the application of laser welding on hard metals and provided a basis for subsequent research. The research showed the possibilities of laser joints between hard metals and carbon steel with Cu–Ag–Ni interlayers prepared using a Triumphf disc laser. The laser beam focusing on the steel surface was observed to be helpful for avoiding direct interaction between the laser beam and hard metal. Yin et al. [72] reported on the effect of the interlayer thickness on the metallurgical bonding, microstructure, and bend strength. Figure 7a–d present cross-sectional views of the microstructure of the as-welded specimens, base metal, joints, and HAZ. Figure 7e shows a typical as-annealed specimen, and Figures 7f–h and 7i–k show the microstructures of the base metal and HAZ annealed at 1250 °C for 2 h, 8 h, and 16 h. The aggregates and grain growth were observed on the cemented carbide side (Figure 7l–n). Increasing the thickness of the Invar interlayer from 1 to 1.5 mm resulted in continuous metallurgical bonding, free of voids and oxide films. When the thickness was increased to 2 mm, the porosity in the fusion zone increased and cracks developed near the fusion boundary on the WC–Co side, reducing the strength. As shown in Figure 7o, the maximum bend strength was achieved for 4 mm thick specimens with 1.5 mm thick Invar as the interlayer. The fusion boundary, HAZ, and the fusion zone on the WC–Co side were susceptible to cracking.

The Focused Ion Beam (FIB) technique was used to prepare TEM samples (Figure 7p), and the chemical compositions were measured using High angle Annular Dark Field (HAADF) energy dispersive X-ray spectroscopy (EDS) (Figure 7q) [73]. The microstructures at the WC/M interface after heat treatment were characterized using high-resolution TEM, revealing high densities of dislocations (Figure 7r–w). Energy-dispersive X-ray spectroscopy (EDS) analysis suggested that long-range solute diffusion occurred inside the WC grains and at the WC/Fe (Ni)/WC interfaces. These findings are helpful for theoretical research on dissimilar cemented carbides and steels.

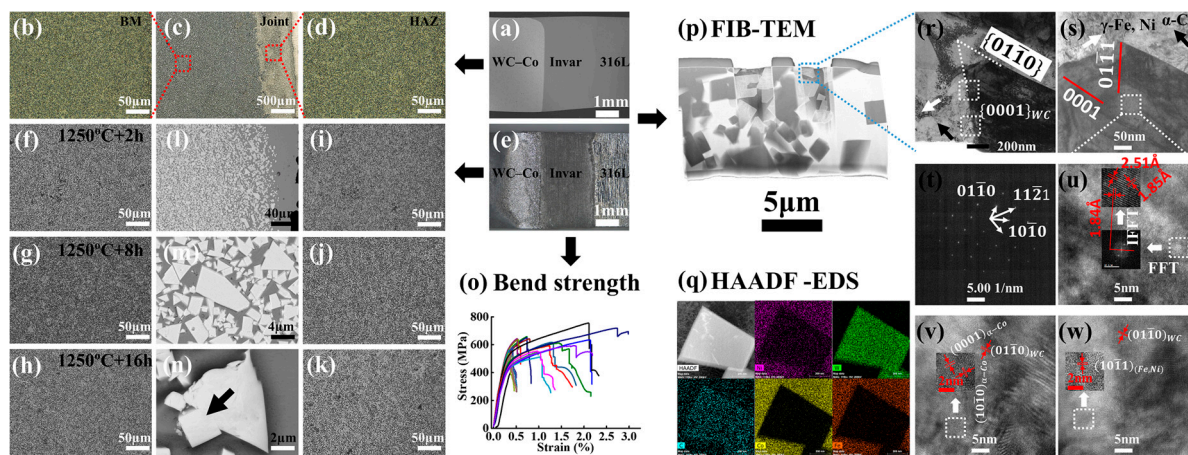


Figure 7. Fiber laser welding of cemented carbide and steel using Invar filler metals. (a) As-welded joint (cross-sectional view) composed of WC–Co (left side), 316L stainless steel (right side), and an Invar interlayer (middle). Higher-magnification images of (b) base material, (c) joint, and (d) HAZ; (e) as-annealed joints of WC–Co (left side) and 316 L stainless steel (right side) with an Invar interlayer (middle) held at 1250 °C; base materials held at 1250 °C for (f) 2 h, (g) 8 h, and (h) 16 h; and HAZ held at 1250 °C for (i) 2 h, (j) 8 h, and (k) 16 h; and (l–n) carbide grain growth after annealing. (o) Bend strength. (p) FIB–TEM image, (q) HAADF–EDS images, and (r–w) bright–field TEM and high–resolution TEM images showing structure and crystal lattice. Adapted from [72,73], with permission from Elsevier, 2018.

2.3.4. Laser–TIG

In laser–TIG (tungsten–inert gas) hybrid welding, a laser beam interacts in the same molten pool as the molten pool created by a secondary heat source. The first attempts to join a laser and conventional welding torch into one welding process were made in the late 1970s at Imperial College, London by a group led by William Steen. The advantages of laser/TIG hybrid welding are as follows:

1. High-efficiency process ($\geq 80\%$);
2. Ability to bridge relatively large gaps (≥ 0.5 mm);
3. Slow cooling rates because of lower welding speed and higher heat input;
4. Highly reflective materials are generally easy to weld [74]. Laser/TIG hybrid welding of cemented carbide to steels followed the wetting mechanism.

Xu [75] reported on the dissimilar welding between cemented carbide and Invar alloy using a CO₂ laser beam and argon arc as heat sources. Taking 4 mm thick and 6 mm thick plates as examples, Figure 8 shows the weld formation and cross-sectional views of WC–Co/steel joints prepared by laser/TIG hybrid welding. Figure 8a–c show the microstructures on the top side, in the middle, and at the bottom of the joint after parameter optimization. WC dissolving to the fusion zone is observed in Figure 8d. The results in Figure 8e–h indicate that the main welding defects were porosities and welding cracks. Cold cracks formed near WC–Co interface, especially in the HAZ on the WC–Co side. However, hot cracks were always observed on the steel side. The main probable reasons for the defects included insufficient heat generation at the beginning stage and a non-uniform laser beam or tungsten arc. The microstructure in the fusion zone exhibited obvious crystal structures (Figure 8i,j). The microstructures labeled I, II, and III in Figure 8k at the WC–Co/Invar interface are shown at a higher magnification in Figure 8l–n and consisted of columnar crystals, cellular crystals, and a eutectic structure with fir-tree crystals and dendritic crystals, respectively. The columnar crystals were surrounded by many fir-tree crystals [57,76].

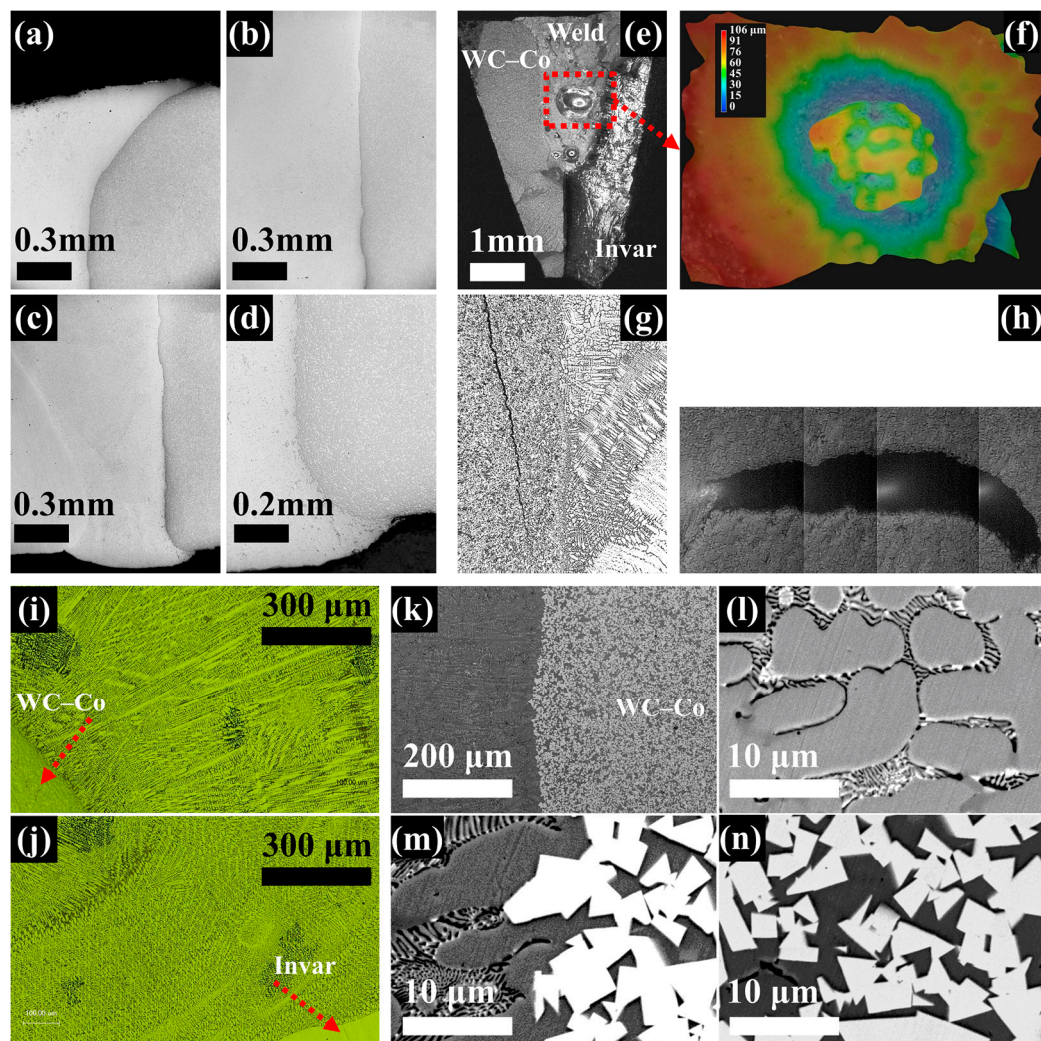


Figure 8. Laser-TIG hybrid welding of cemented carbide and steel without filler metals. As-welded microstructures (cross-sectional view) of the (a) top side, (b) middle, and (c) bottom. (d) Higher-magnification image showing the WC dissolution. (e,f) Porosities. (g,h) Cracks. (i,j) Optical micrographs of the fusion boundary on the WC-Co side and steel side. (k) SEM images of the fusion boundary on the WC-Co side and magnified images of the microstructure (l) in the fusion zone, (m) on the fusion boundary on the WC-Co side, and (n) of the cemented carbide. Adapted from [75], with permission from Elsevier, 2011.

Table 3 provides a summary of previous reports on dissimilar hard metals/steel joints fabricated by laser welding. A disk laser, CO₂ laser, Nd/YAG laser, laser diode (LD) laser, and fiber laser were selected as heat sources. The laser-brazing method can be used for poor-weldability materials (such as graphite [77]).

Table 3. Comparison of previously reported laser welding methods of dissimilar hard metal/steel joints. Nd/YAG, neodymium/yttrium aluminum garnet.

Hard Metals	Counterparts	Inserts	Thickness (mm)	Lasers	Reference
L135 YG15	C45 6542	-	1.5/3	CO ₂ Laser	[59]
K10 K40	1.7182	-	2.5	CO ₂ Laser Nd/YAG	[60]
YG20	C45	-	2/3/4	Fiber Laser	[62]
K10 K40	Hypoeutectoid steel	-	2.5–2.9	CO ₂ Laser Nd/YAG	[63]
HM1-4	1.1231	-	1	Fiber Laser Nd/YAG	[64]
YG20	C45	Invar	3/4	Fiber Laser	[68]
H10S G10	C45	Cu–Ag–Ni	/	Disk Laser	[69]
YG30	C45	-	6	CO ₂ Laser	[71]
K10	Graphite	Cu–Ag–Ti	10 × 10 × 2 5 × 5 × 3.5 3 × 3 × 0.1	YAG Laser Laser Diode (LD) laser	[75]
					[76]
					[77]

2.4. TIG Welding

TIG welding is a process used to join metals by heating them with an arc between a tungsten electrode and the metals [78]. On the basis of previous investigations [79,80], Zhao et al. [81] and her group [82–84] patented novel Ni–Fe–C filler metals. TIG welding was observed to have the bridge ability to increase the metallurgical bonding. The addition of carbon to nickel–iron filler metals is helpful for inhibiting the formation of mixed carbides and improving the bend strength of dissimilar joints. When the plate is smaller or equal to 2 mm thick, single-side welding is adaptable. However, if the thickness is increased to 4 mm, double-side welding is recommended.

2.5. Diffusion Bonding

Diffusion bonding [85] is a process that produces solid-state coalescence between two materials under the following conditions:

1. Temperature is below the melting point.
2. Loads producing coalescence of contacting surfaces are below those that would lead to macroscopic deformation.
3. Interlayer (foil or coating) can be used as a bonding aid.

During the fabrication of diffusion-bonded joints of cemented carbide and steel, interlayers such as copper, silver, or nickel are used as a buffer. Cottenden et al. [86] investigated the metallurgical structure and strength of dissimilar joints of cemented carbide and copper, cobalt, and nickel. Diffusion-bonded joints exhibit superior mechanical properties compared with those of brazed joints [87]. These results lay a foundation for revealing the mechanism of η -phase formation. The effect of a diffusion layer on the interface strength depends on the following factors: the mechanical properties of the diffusion layer [88], surface roughness and thickness of the interlayer [89], strength of the interfacial bond, and mode of loading at the interface.

In addition to a nickel interlayer, silver [90], copper–nickel [91], and nickel–titanium [92] are also used as the interlayer for diffusion-bonded joints. Functionally gradient cemented carbide [93] and interlayers with “sandwich” structures [94] are helpful for reducing the difference in physical and chemical properties between cemented carbide and steel.

2.6. Electron-Beam Welding

Electron-beam welding refers to electron processes in which a beam of electrons can be focused to power densities high enough to melt and vaporize the metals being joined [95]. Vacuum is helpful to avoid contamination of the weld by interstitials, and tungsten and its alloys can be successful joined

by electron-beam welding [27]. Electron-beam welding of cemented carbides without interlayers was confirmed to be effective [96]. The η phases were observed to be formed when a small beam current and low welding speed were used. Carbon depletion in the W–C–Co system and elemental (carbon and iron) diffusion between the cemented carbide and the fusion zones were the main influencing factors for mixed carbide formation. When using Ni–Fe [97] or Ag–Cu–Ti [98] as the interlayer/solder, electron-beam welding–brazing can be used to join cemented carbide and steel. The electron-beam-welded joints were confirmed to be satisfactory fusion welds and brazing joints.

2.7. MIG Welding

Metal-inert gas (MIG) welding employs an electric arc established between a consumable wire electrode and the workpiece to be joined. As illustrated in Figure 9, using robotic MIG welding, Yin et al. [99] welded WC–TiC–Ni cemented carbide to 304 stainless steel using nickel as the filler metal. The results indicated that the weld formation strongly depended on the groove angles. The transitional layers of the WC–8Co were smaller than those of the WC–20Co. The results also indicated that the dissolving behaviors of tungsten carbide occurred not only in the fusion zone, but also in the HAZ, especially near the top surface. The fusion zone has the ability to cure the cracks itself during MIG welding.

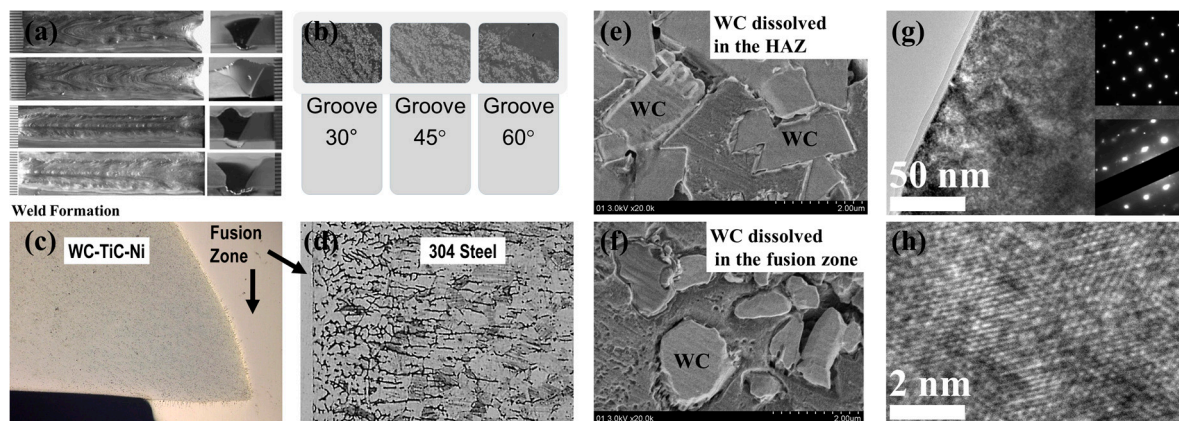


Figure 9. Metal-inert gas (MIG) welding of cemented carbide and 304 stainless steel. (a,b) Front-side weld formation and cross-sectional view of the corresponding joint (groove angles of 30°, 45°, and 60°). (c,d) Optical microstructures of the fusion boundary on the WC–Co side and 304 side. WC dissolving behaviors occurred (e) at the HAZ and (f) in the fusion zone. (g) TEM image and selected area electron diffraction (SAED) patterns showing the typical austenitic structure and (h) high-resolution TEM image showing the corresponding crystal lattice. Reproduced or adapted from [99], with permission from Springer, 2018.

Tungsten was observed in the fusion zone near the WC–Co side because of the dissolution and chemical decomposition of tungsten carbides. In the fusion zone, a gradient layer was helpful for metallurgical bonding and had the ability to cure the cracks.

2.8. Friction Welding

Friction welding is a solid-state, hot-shear joining process [100] in which the heat for welding is produced by the relative motion of the two interfaces being joined [101]. Okita et al. [102] produced dissimilar joints of cemented carbide and tool steel with an intermediate layer by friction welding. The interlayer contained tungsten carbide and a nickel matrix. The joints exhibited high bend strength after optimization of the forge pressure and friction pressure. The fracture occurred in the vicinity of the fusion zone interface in the intermediate layer. It was concluded that the cracks were formed because of the difference in the deformation rates of the Ni matrix and WC. Subsequently, the crack

was thought to be the main factor for the deterioration of the tensile strength. Avettand-Fénoël et al. explored the dissimilar friction stir welding of WC–Co cermet to steels without an interlayer [103] and with a nickel interlayer [104]. More retained austenite was observed in the nugget with increased Ni interlayer thickness. The results indicated that the interface of the Ni bearing joints had better quality than that of the Ni-free joint. A complex materials processing method has been provided to reduce time consumption by selecting the larger radial feed combined with a proper burnishing pressure to ensure the desired quality and compressive residual stress at the surface [105,106], which are indices of enhancing the fatigue strength at the nugget zone of the friction stir welded area.

2.9. Others

Besides the above welding and joining processes, many attempts have been made to realize the coalescence between dissimilar metals, such as adhesive bonding [107] and ultrasonic welding [32,108]. The coatings on the cermet prepared by the Chemical Vapor Deposition (CVD) technique exhibits excellent wear resistance, superior hardness, and excellent adhesion strength [109]. The ultrasonic vibrations assisted processes can improve the mechanical properties, and result in better material mixing and enhanced turbulence, which lead to the remarkable and globalized applications in the welding and joining of dissimilar metals [110].

3. Existing Main Issues, Corresponding Solutions, and Future Work

Although cemented carbides and their joining techniques have been patented for approximately 90 years [1,25], many unanswered questions remain, including the following.

The fundamental theory of the joining of dissimilar cemented carbide has been established. However, relatively few specific situations have been studied. In the future, extensive experiments and different types of theoretical modeling should be performed in this area.

The conventional processes are particularly well suited for welding thicker cemented carbides, but less suited when high-efficiency welding is needed and without vacuum situations. Some processes to join cemented carbides have been attempted. However, relatively few high-efficiency welding processes have been confirmed to be satisfactory for production. With the development of laser science and technology, laser welding will become an important potential process for the joining of cemented carbides and steels. Ultrafast laser welding shows potential for achieving a lower heat input [111].

Welding is a process of achieving complete coalescence (at the atomic level) of two or more materials through melting and re-solidification of the base metals and filler metal. However, only relatively few studies have focused on atomic-level research. In the future, extensive tests and microstructural characterization at the atomic level should be performed.

4. Conclusions

Dissimilar welding of cemented carbides is a key promising component in modern manufacturing. The joining processes of cemented carbides and steel were reviewed.

- (1) Conventional methods, such as sinter-bonding and vacuum brazing, are the most common approaches because of the vacuum environment, good wettability, and low cost, which are particularly well suited for thicker cemented carbides with lower cobalt contents.
- (2) High-efficiency welding processes, such as laser welding, are the most promising joining methods and are better suited for thinner cemented carbides with increased cobalt contents ($\geq 20\%$). A laser beam can be positioned on the steel side and follows the dip soldering mechanism. A laser beam can also be placed on the cemented carbide side and on top of the middle interlayer.
- (3) The coalescence of the melting steels, “soft” binders, and hard carbides (dissolved) is believed to be a possible factor controlling the metallurgical joining. At the collapsed and recreated interface, the higher- and lower-temperature borders of the fusion boundary region are defined by the melting point of the hard carbide and the melting point of the binders.

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