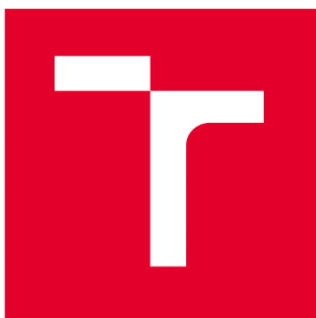




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ÚSTAV KONSTRUOVÁNÍ

ADDITIVE MANUFACTURING OF STRUCTURED COMPONENTS FROM MAGNESIUM ALLOYS

ADITIVNÍ VÝROBA STRUKTUROVANÝCH KOMPONENT ZE SLITIN HOŘČÍKU

DOCTORAL THESIS

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STATEMENT

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ABSTRACT

Wire arc directed energy deposition (WADED) represents a promising additive manufacturing technology for the fabrication of large-scale lightweight structural components. This potential arises from the low density of magnesium alloys combined with their high strength-to-weight ratio. However, their processing by arc-based additive manufacturing remains challenging due to their high chemical reactivity, thermal sensitivity, and susceptibility to defect formation. A limited understanding of deposition strategies, process stability, and thermal management has restricted the wider application of WADED for magnesium alloys. This dissertation focuses on WADED processing of the AZ61 magnesium alloy to improve geometrical accuracy, process stability, and material microstructural while mitigating residual stresses. The research systematically investigates the influence of deposition path planning, cold metal transfer (CMT) characteristic parameters, and thermal management strategies on bead geometry, microstructural evolution, and residual stress development. Particular attention is paid to characteristic structural features, including thin walls, overhangs, and volumetric components. The results demonstrate that appropriate path-planning strategies can significantly reduce geometric deviations in thin walls and volumetric components. Furthermore, the study shows that individual CMT characteristic parameters strongly influence metal transfer behaviour and process stability, highlighting the need for their systematic optimisation. Thermal management strategies, including preheating and interpass temperature control, are shown to reduce residual stresses while simultaneously promoting grain coarsening. Therefore, it is necessary to identify an optimal compromise between microstructural refinement and residual stress reduction. The outcomes of this work contribute to a deeper understanding of WADED processing of AZ magnesium alloys and provide practical guidelines for improving process robustness and component quality. The findings support the broader application of WADED for magnesium alloy structural parts and establish a foundation for future development of arc-based additive manufacturing of lightweight materials.

KEYWORDS

Wire arc directed energy deposition (WADED or WAAM), magnesium alloys, wire arc additive manufacturing, thin walls, overhangs, volumetric parts, deposition strategies, process parameters, residual stresses, microstructure, porosity, mechanical properties

ABSTRAKT

Technologie WADED (wire arc directed energy deposition) představuje perspektivní metodu aditivní výroby pro produkci rozměrných lehkých konstrukčních součástí. Potenciál použití hořčkových slitin pro tyto aplikace je založen hlavně na jejich nízké hustotě a vysokému poměru pevnosti k hmotnosti. Jejich zpracování pomocí technologie obloukového svařování je však náročné vzhledem k vysoké chemické reaktivitě, výrazné tepelné citlivosti a náchylnosti ke vzniku vad. Omezené porozumění plánování trajektorií, stability procesu a řízení tepelné distribuce dosud bránilo širšímu uplatnění technologie WADED pro hořčkové slitiny. Tato disertační práce se zaměřuje na zpracování hořčkové slitiny AZ61 technologií WADED s cílem zlepšit geometrickou přesnost, stabilitu procesu a mikrostrukturu materiálu při současném omezení vzniku zbytkových napětí. Výzkum systematicky analyzuje vliv plánování trajektorií navařování, charakteristických parametrů procesu CMT (cold metal transfer), strategií řízení tepelné distribuce na geometrii návaru, vývoj mikrostruktury a vznik zbytkových napětí. Zvláštní pozornost je věnována charakteristickým dílčím prvkům strojních součástí, mezi něž patří tenkostěnné struktury, převisy a objemové komponenty. Výsledky ukazují, že vhodně zvolené strategie plánování trajektorie mohou významně snížit geometrické odchylky u tenkostěnných i objemových dílů. Práce dále prokazuje, že jednotlivé charakteristické parametry CMT mají zásadní vliv na chování přenosu kovu a stabilitu procesu, což zdůrazňuje potřebu jejich systematické optimalizace. Strategie řízení tepelné distribuce, včetně předehřevu a kontroly teploty mezi jednotlivými vrstvami, vedou ke snížení zbytkových napětí, avšak současně podporují hrubnutí zrna. Z tohoto důvodu je nutné nalézt vhodný kompromis mezi zjemněním mikrostruktury a redukcí zbytkových napětí. Výsledky této práce přispívají k hlubšímu porozumění zpracování hořčkových slitin AZ série technologií WADED a poskytují praktická doporučení pro zlepšení robustnosti procesu a kvality vyráběných komponent. Závěry práce podporují širší využití technologie WADED pro konstrukční díly z hořčkových slitin a vytvářejí základ pro další rozvoj aditivních technologií pro lehké konstrukční materiály.

KLÍČOVÁ SLOVA

Wire arc directed energy deposition (WADED nebo WAAM), hořčkové slitiny, obloukové svařování, tenkostěnné struktury, převisy, objemové díly, depoziční strategie, procesní parametry, zbytková napětí, mikrostruktura, porozita, mechanické vlastnosti

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1 INTRODUCTION

In recent decades, additive manufacturing (AM) has gained increasing importance as a manufacturing technology in modern engineering, offering new possibilities for fabricating machine components that cannot be achieved using conventional manufacturing processes. Based on the layer-by-layer deposition of material, AM enables high material efficiency together with a high degree of geometrical freedom [1]. One of its key advantages lies in the ability to produce components with complex geometries that would otherwise require multiple manufacturing steps or the assembly of several simpler parts when manufactured using traditional methods. As a result, additive manufacturing has been increasingly adopted across a wide range of industrial applications, including the aviation, automotive, and energy industries [2].

Among existing metal additive manufacturing technologies, wire-based metal additive manufacturing has attracted interest for its ability to achieve high deposition rates and fabricate large-scale components. An example of these methods is wire arc directed energy deposition (WADED), based on arc welding processes in which metallic wire is supplied as the feedstock and melted by an electric arc. The WADED setup typically consists of a welding power source (e.g., MIG, MAG, TIG, or their modifications), a wire feeding system, and a robotic or CNC-based positioning unit to deposit material in a layer-by-layer manner [1]. Compared to other metal additive manufacturing approaches, WADED offers lower material costs, higher productivity, and reduced manufacturing times, making it particularly suitable for large structural parts [3].

The low density and high strength-to-weight ratio of magnesium alloys make them suitable for structural applications where weight reduction is required. According to a recent study [4], the use of magnesium alloys as substitutes for aluminium and steel components in automotive allows for weight reductions of 20–25% and 45–50%, respectively, while maintaining structural strength. However, the use of magnesium components is not limited to automotive applications but also includes aerospace, contributing to reduced fuel consumption and more environmentally friendly transportation [4, 5]. As a result, recent reviews [5, 6] consider magnesium alloys to be the most promising green engineering materials in the 21st century.

Despite these advantages, the broader industrial application of magnesium alloys remains limited by manufacturing-related challenges. Poor corrosion resistance can reduce component service life, but various approaches aimed at improving corrosion performance have been extensively investigated [7]. In addition, limited plastic formability resulting from the hexagonal crystal structure restricts the applicability of several conventional manufacturing processes. Therefore, most magnesium components are produced by casting, which is prone to casting-related defects [4]. These limitations motivate the exploration of additive manufacturing as an alternative processing route for magnesium alloys.

Motivation for further technological development is not driven only by material-related challenges. In the aerospace industry, increasing attention is paid to the efficiency of material utilisation during the manufacturing of aircraft components. This efficiency is commonly described by the buy-to-fly (BTF) ratio, defined as the ratio between the weight of input material and the weight of the final component. In conventional manufacturing routes, BTF ratios often exceed 10:1, indicating that more than 90% of the original material is removed as waste [8]. Therefore, the aerospace industry faces a strong demand for manufacturing processes capable of replacing current production routes for large structural components, such as cruciforms, stiffened panels, and wing ribs, which are typically machined from large semi-finished products [1].

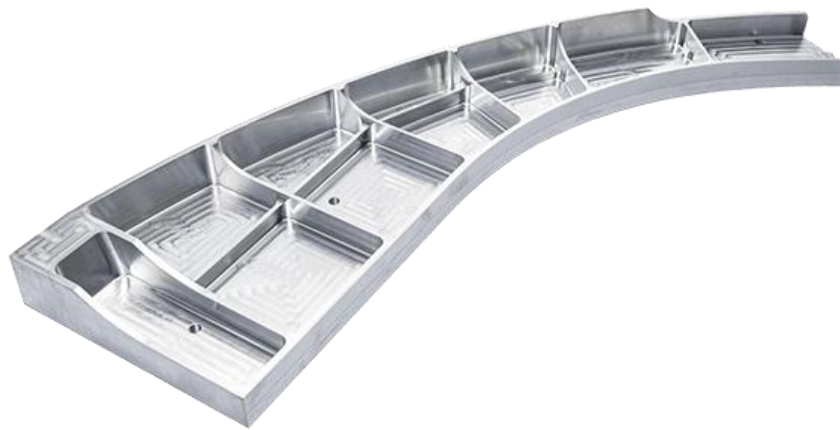


Fig. 1-1 An example of a structured component - Aircraft C-Shape Frame [9]

Another example of such components is the Aircraft C-Shape Frame [9], shown in Fig. 1-1 which will be discussed in more detail later in this dissertation. From a manufacturing perspective, aircraft components of this type typically consist of a combination of thin walls, volumetric regions, inclined surfaces, and other geometrically constrained features. For this work, the term structured components refer to these parts whose geometry is intentionally designed to minimise weight and material consumption while maintaining structural performance. In conventional manufacturing, the production of these components is often associated with high buy-to-fly ratios due to the extensive material removal required for their manufacture. By contrast, the near-net-shape capability of additive manufacturing significantly reduces material waste, making wire-based processes, such as WADED, particularly attractive for large-scale, structured components [1]. For this reason, this dissertation focuses on strategies for fabricating structured components from magnesium alloys using the WADED process. The investigated approaches include deposition path planning for thin-walled, inclined, and volumetric parts, as well as the optimisation of welding parameters. Additionally, the influence of other process-related factors, such as base material preheating and welding wire preparation, on porosity formation, material microstructure, and mechanical properties is considered.

2 STATE OF THE ART

Additive manufacturing (AM), also known as three-dimensional (3D) printing, is a computer-controlled fabrication process for producing objects usually by layer-by-layer deposition of material [10]. In contrast to conventional subtractive manufacturing technologies, which rely on material removal, AM is based on successive material addition during part fabrication. A principal advantage of additive manufacturing is ability to produce geometrically complex components, often enabling substantial weight reduction and design flexibility that cannot be readily achieved with traditional manufacturing methods [11]. According to ISO ASTM 52900, AM processes are classified into seven categories based on their manufacturing principles, as illustrated in Fig. 2-1 [12]. Within the scope of this work, the most significant attention is devoted to directed energy deposition (DED). DED is an additive manufacturing process in which focused thermal energy is used to melt feedstock material supplied in the form of wire or powder [13]. Based on the energy source used for melting, DED processes are commonly classified into laser-based, electron beam, and arc systems [13].

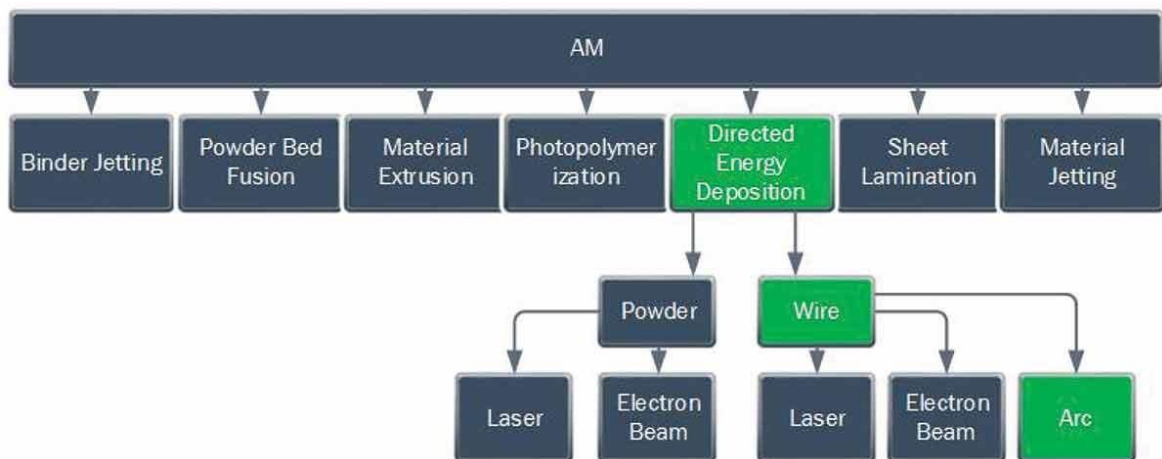


Fig. 2-1 AM processes classification [12]

Among the currently available additive manufacturing technologies, laser powder bed fusion (LPBF), wire arc directed energy deposition (WADED), binder jetting (BJ), electron beam directed energy deposition (EBDED), and friction stir-based additive manufacturing have been investigated for the fabrication of metallic materials, including magnesium alloys [2, 14, 15]. LPBF is characterised by high geometric precision and fine microstructural control. However, its application is limited by relatively low deposition rates, restricted build volumes, and high system and feedstock costs [2].

Binder jetting enables the production of complex geometries with high dimensional accuracy. The thermal input during printing is relatively low. However, the process requires subsequent sintering, which may introduce distortion and residual porosity [15]. Furthermore, component size is constrained by the build chamber. Electron beam directed energy deposition benefits from the high energy density of the electron beam. As a result, material reflectivity is less critical than in laser-based processes [16]. Fabrication is performed under vacuum, which reduces oxidation, a factor particularly relevant for magnesium alloys. These advantages are counterbalanced by the high capital cost of EBDED systems and limitations in build chamber size. In contrast, WADED offers high deposition rates and relatively low equipment cost [1, 2, 15]. The process is highly scalable and enables the fabrication of large components. These characteristics make WADED attractive for structural applications where component size and production efficiency are critical. Challenges remain in dimensional accuracy, surface quality, thermal management, and mechanical properties. Nevertheless, WADED represents one of the most promising additive manufacturing routes for large-scale magnesium alloy components.

2.1 Wire arc directed energy deposition

Wire arc directed energy deposition (WADED) is a variant of directed energy deposition in which the feedstock material is supplied as a wire and melted using an electric arc as the heat source [15]. In the literature, this manufacturing approach is also commonly referred to as wire arc additive manufacturing (WAAM). A typical WADED setup consists of a positioning system, most commonly a robotic manipulator, a welding power source, a welding table or additional positioning system, a shielding gas supply, a gas exhaust system, and a monitoring system. Depending on the level of process control, the monitoring system may include accuracy monitoring, thermal imaging, or melt pool monitoring [17]. A schematic of a typical WADED workplace is shown in Fig. 2-2 [18].

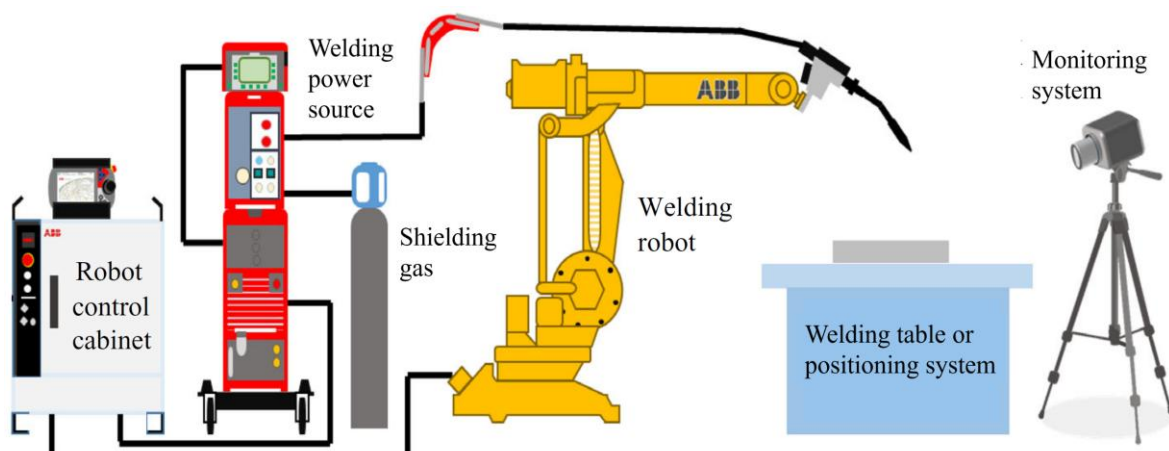


Fig. 2-2 Diagram of a typical WADED workplace (edited) [18]

2.1.1 Welding power sources

The WADED process includes three main types that differ by welding principle: gas metal arc welding (GMAW), gas tungsten arc welding (GTAW), and plasma arc welding (PAW) [13, 19]. Several designations are used for the same welding processes, reflecting differences in terminology across regions. Tungsten inert gas welding is commonly referred to as TIG in English literature, WIG in German literature (Wolfram inert gas), and GTAW in American standards. The term gas metal arc welding (GMAW) covers two closely related processes that differ in the type of shielding gas employed, namely metal inert gas (MIG) and metal active gas (MAG) welding. Active shielding gases include those that chemically interact with the molten pool, such as argon-carbon dioxide mixtures. In contrast, inert gases such as argon and helium do not react with molten metal.

In WADED, different welding power sources exhibit various characteristics that influence process complexity, productivity, and thermal input. Systems based on gas GTAW and PAW require torch rotation to maintain a consistent wire feeding direction, which increases the complexity of path planning and process control [2, 13]. In contrast, GMAW systems are generally associated with lower setup costs and more straightforward integration into robotic WADED platforms. GMAW also offers substantially higher productivity, with deposition rates commonly reported to be two to three times higher than those achieved using GTAW [2, 20]. These advantages are accompanied by higher heat input, which can lead to increased residual stresses, distortion, and microstructural degradation [13]. To address these limitations, modified GMAW variants such as cold metal transfer (CMT) have been developed. By employing controlled metal transfer, CMT significantly reduces heat input while enabling spatter-free metal transfer and a highly stable electric arc, making it particularly suitable for WADED applications [15].

2.1.2 Cold metal transfer

Cold metal transfer (CMT) is a modified gas metal arc welding (GMAW) process based on controlled short-circuit metal transfer, developed by Fronius. The process relies on synchronising electrical current control with active retraction of the filler wire during short-circuiting, enabling droplet detachment primarily through surface tension [21].

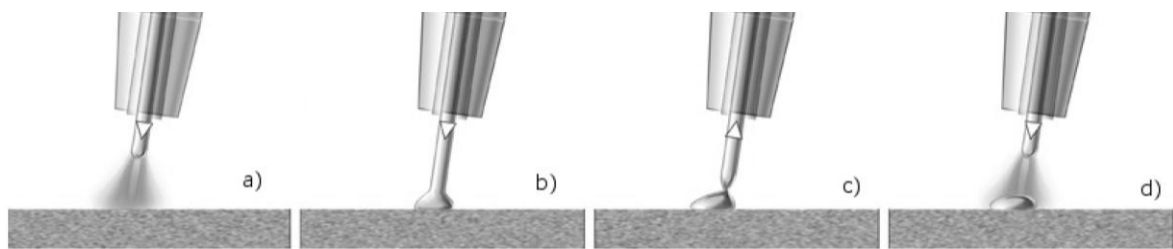


Fig. 2-3 Wire movement during the CMT cycle [22]

The CMT process consists of a periodically repeating transfer cycle. Initially, the end of the wire is melted by a stable electric arc, see Fig. 2-3a. The wire is then moved toward the molten pool until short-circuit contact is established, see Fig. 2-3b. At this point, the electric arc is interrupted, and a molten droplet is transferred to the substrate. After droplet detachment, the wire is moved away from the molten pool and the cycle is repeated, see Fig. 2-3c, d. The key advantage of the CMT technique is the interruption of the arc, which significantly reduces heat input.

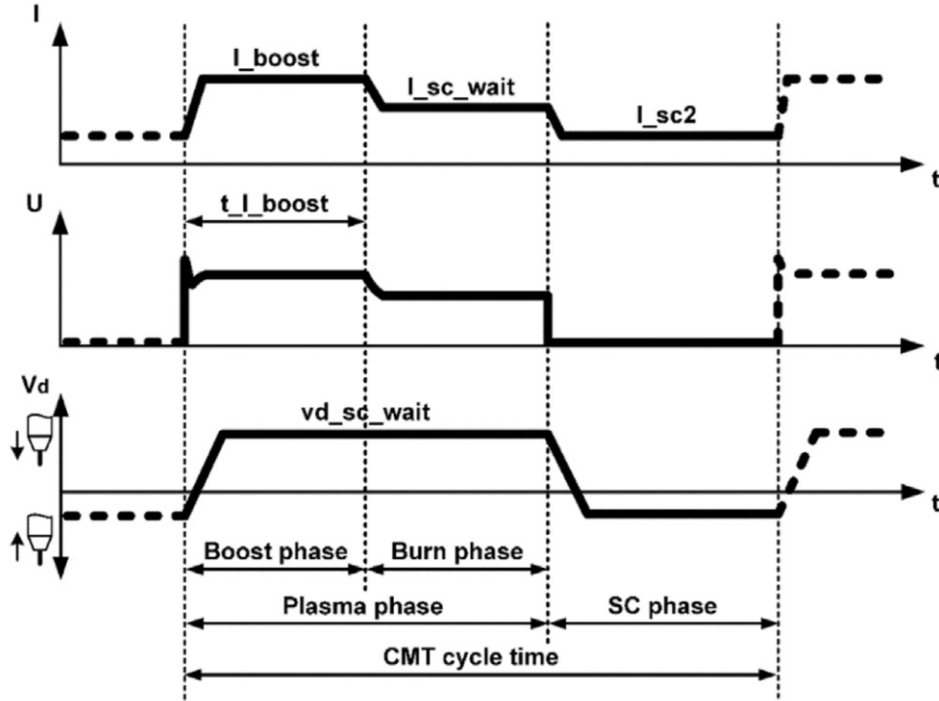


Fig. 2-4 Characteristic of CMT welding process [23]

The transfer cycle can be divided into several characteristic phases as shown in Fig. 2-4 [23]. During the Boost phase, current and voltage reach their peak values, and the wire tip melts while remaining separated from the substrate. This is followed by the Burn phase, in which the wire moves toward the molten pool and the current and voltage are progressively reduced. Upon contact between the wire, the molten droplet, and the substrate, the Short-circuit phase begins. In this phase, current and voltage are minimised, and after a short time interval, controlled wire retraction initiates droplet detachment [23].

Available studies [23, 24] identify several primary CMT parameters that influence the individual phases of the metal transfer cycle described above. These parameters include the boost phase current – I_{boost} (A), boost phase duration – t_{I_boost} (ms), burn phase current – I_{sc_wait} (A), the speed of the electrode during the boost and burn phase – vd_{sc_wait} (m·min⁻¹) and SC phase current I_{sc2} (A).

In practical applications, operators typically do not set these parameters individually. Instead, CMT power sources are commonly equipped with synergic welding systems, in which the operator adjusts a single primary parameter, such as welding arc current, or wire feed speed (WFS). The remaining process parameters are then automatically adapted according to a predefined synergic characteristic line. However, such synergic lines are not available for all filler materials, including magnesium alloys. Consequently, no comprehensive study of magnesium alloys has yet been reported that systematically evaluates the influence of individual CMT parameters on metal transfer behaviour, process stability, and weld bead formation.

In addition to the conventional CMT mode, advanced variants such as CMT Pulse and CMT Advanced have been developed further to expand the control of metal transfer and heat input. CMT Pulse combines controlled short-circuit transfer with pulsed phases, which can improve penetration behaviour and may enable higher deposition rates [25]. CMT Advanced introduces alternating polarity cycles, allowing additional reduction in heat input while maintaining stable metal transfer and potentially increased deposition rates [26]. However, the application and systematic evaluation of these advanced CMT modes for WADED remain largely unexplored in the available literature.

2.1.3 Process parameters

In addition to the parameters governing the CMT transfer cycle itself, several other process parameters directly influence metal transfer behaviour and weld bead geometry. These include the wire material and diameter, the type and flow rate of shielding gas, the travel speed of the welding torch, the wire feed rate, the distance between the contact tip and the workpiece, weld start and end correction, as well as thermal management [2, 19, 27].

The electrode diameter represents one of the fundamental parameters in welding processes. Depending on the filler material and wire diameter, an appropriate predefined CMT characteristic is selected. An increase in wire diameter generally requires higher energy input to achieve stable melting, sufficient penetration and metal transfer.

The travel speed, also known as the welding speed (TS or WS), significantly influences the geometry of the deposited weld bead. An increase in travel speed generally reduces both bead height and bead width. Travel speed also affects the microstructure of the deposited material. Higher travel speeds are commonly associated with finer grain structures, which can be attributed to increased cooling rates resulting from reduced heat input per unit length. However, at very high speeds, inconsistent deposits may form.

Arc current significantly influences weld-bead geometry. An increase in arc current is generally associated with a deterioration in surface roughness, while parameters such as penetration depth, contact angle, and bead width tend to increase [2].

The wire stick-out defines the distance between the contact tip and the molten pool or substrate during WADED [19]. An increase in electrode stick-out results in higher electrical resistance, which affects the wire's melting behaviour and, consequently, the geometry of the deposited weld bead [19]. If the stick-out is not appropriately selected, it may lead to instability or interruption of the welding arc.

The wire feed rate (WFR or WFS) determines the amount of filler material supplied to the molten pool per unit time [28]. An increase in WFR is usually associated with larger weld bead dimensions and a reduced contact angle [2].

The shielding gas flow protects the molten pool from oxidation due to atmospheric exposure [2]. For reactive materials such as magnesium alloys, inert shielding gases are essential to minimise oxidation and ensure stable deposition. Commonly used inert gases include argon, helium, and their mixtures. In addition to providing adequate protection of the molten pool, the composition of the shielding gas can influence the contact angle and penetration depth [29].

Thermal management in WADED production represents a key factor influencing the final properties of the manufactured component. Process parameters affecting thermal management include the preheating temperature and the interpass temperature, defined as the part temperature between successive layers. Appropriate thermal management can reduce thermal gradients associated with geometric instability, mitigate residual stresses, and improve microstructural features and mechanical properties [27, 30–32].

2.1.4 Deposition strategies and path planning

Machine components typically consist of several characteristic feature types that require specific WADED deposition approaches. Common features include thin walls, overhangs, and volumetric regions.

Thin-walled parts in the context of WADED are geometries formed by the vertical stacking of single tracks without transverse torch oscillation [33]. During the fabrication of these structures, only one perimeter is deposited per layer and the resulting wall thickness corresponds to the width of a single track. The final fabricated geometry does not adjoin or overlap the surrounding material. In contrast, volumetric regions are formed by multiple adjacent weld beads deposited within a single layer. In this case, sufficient fusion must be ensured not only with the underlying layer but also between neighbouring beads to achieve adequate bonding and geometric integrity. Overhangs are geometric regions of a part that extend beyond previously deposited layers without direct support from the underlying material. These unsupported features pose challenges during fabrication due to the lack of mechanical support and thermal conduction. When overhangs cannot be eliminated by adjusting the build orientation, additive support structures are often required.

Thin walls and overhangs

The fabrication of thin-walled structures is one of the most common applications of wire arc directed energy deposition. Previous studies have demonstrated that the dimensional accuracy and stability of thin walls are strongly influenced by single-bead geometry, thermal management, process stability, and the applied path-planning strategy [34, 35]. The start and the end of individual weld beads are frequently associated with local variations in thermal conditions and material deposition, which can result in either excessive or insufficient bead height [34, 36]. These local irregularities may accumulate across successive layers, leading to pronounced geometric deviations.

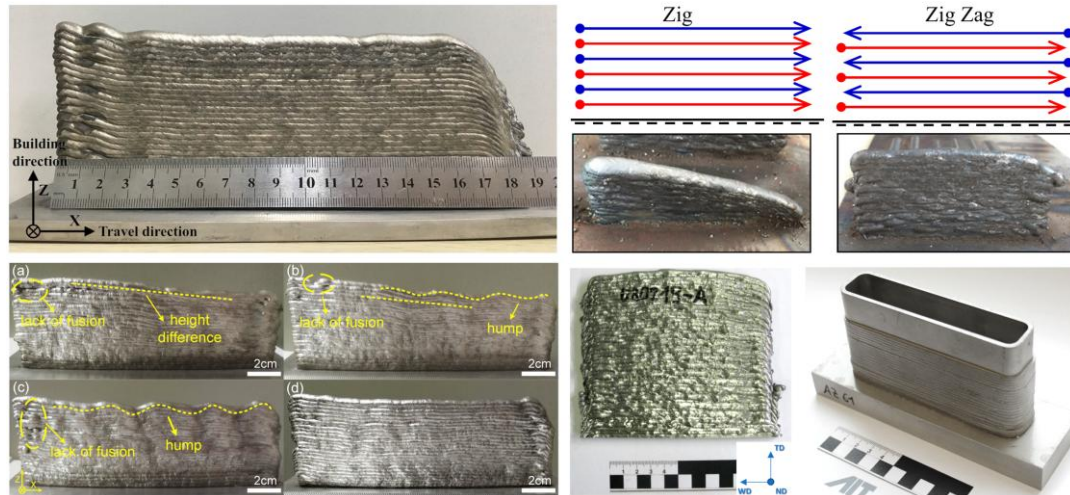


Fig. 2-5 Deposition irregularities in thin walls [34, 35, 37, 38]

As shown in Fig. 2-5, several authors have investigated the production of thin-walled structures [34, 35, 37, 38]. However, most studies have focused primarily on process parameter selection and material properties rather than on deposition path-planning strategies. One of the rare works addressing the influence of path orientation in typical thin-wall geometries is the study by Hu et al. [34], which examined bead start and end abnormalities in GMAW-based deposition. For large thin walls, it is necessary to consider deformations induced by residual stresses, which are typically most pronounced in the direction of deposition [2]. To suppress these deformations, conventional welding practice divides long beads into shorter segments deposited in controlled sequences to reduce heat accumulation and residual stresses [39, 40].

Overhanging structures pose challenges across a wide range of AM technologies, and WADED is no exception. Due to the absence of underlying support material and reduced thermal conduction, the fabrication of overhangs requires specific manufacturing considerations. In WADED, two principal approaches can be employed to produce overhanging features [41–43]. The first approach [42] uses a positioning device to maintain a constant orientation of the melt pool with respect to gravity. The second approach [43] relies on a fixed orientation of the fabricated part, combined with adaptive melt-pool orientation and robot kinematic adaptation to align the weld torch with the overhang.

The use of a positioning device increases process stability and reduces the adverse effects of gravity-induced material flow. However, this approach is associated with significant limitations, particularly regarding the maximum size of the manufactured component, which is constrained by the positioning system's dimensions and the reduced effective build volume. In addition to gravitational effects, the stability of material deposition in overhang regions is strongly influenced by heat input, bead shape control, and the selected path-planning strategy.

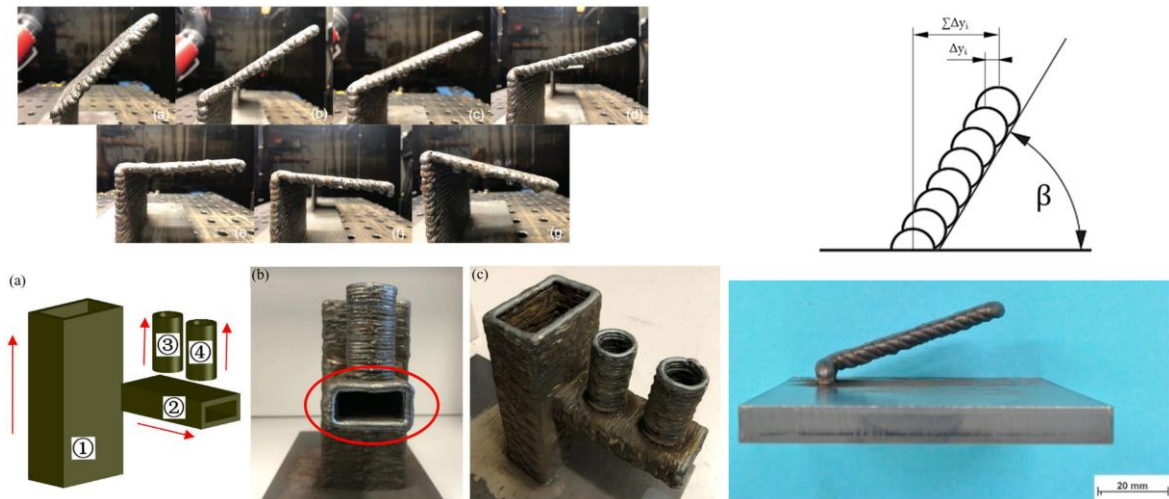


Fig. 2-6 WADED overhang fabrication [42–44]

Several studies on WADED overhang fabrication [41–44] have shown that inappropriate deposition orientation or excessive heat accumulation can result in bead sagging, humping, lack of fusion, or collapse of unsupported material. To mitigate these issues, deposition strategies and recommendations have been proposed, including reduced heat input, modified torch orientation or travel speeds. Despite these advances, the reliable fabrication of thin-walled structures incorporating overhanging features remains an active area of research, particularly for materials with high thermal sensitivity.

Volumetric components

Unlike thin-walled structures, volumetric parts are formed by depositing multiple adjacent weld beads within each layer. The fabrication of these components requires selecting an appropriate deposition strategy and suitable process parameters to ensure adequate fusion between neighbouring beads and overall geometric accuracy. According to Sharifi [45] and Müller [46], effective parameter selection for volumetric deposition is characterised by wide weld beads, a large contact angle, and sufficient penetration.

Several toolpath trajectories and deposition strategies have been proposed for the manufacture of volumetric parts. For all approaches, it is essential to determine the appropriate distance between adjacent weld beads to ensure sufficient inter-bead penetration while achieving the desired top-surface flatness and dimensional accuracy.

The calculation of bead distances is commonly simplified by approximating the weld-bead cross-sectional geometry using analytical functions, such as a parabolic, arc, or cosine model, see Tab. 2.1 [47]. The specific function is selected based on the experimentally observed bead shape.

Tab. 2.1 Different models for distance calculations [47]

Models	Model function	Bead height, h	Bead width, w
Parabola model	$y = a \cdot x^2 + c$	c	$2 \cdot \sqrt{\frac{c}{-a}}$
Cosine model	$y = a \cdot \cos(b \cdot x)$	a	$\frac{\pi}{b}$
Arc model	$y = \sqrt{a^2 - x^2} + b$	$a - b$	$2 \cdot \sqrt{a^2 - b^2}$

The calculations of distances between adjacent welds aim to produce a flat upper surface for each deposited layer. It is based on the assumption that the overlap area (Fig. 2-7 area CDE) between two adjacent beads equals the unfilled area (Fig. 2-7 area ABC) between their respective bead peaks [48]. For computational simplicity, the weld bead cross-section is often approximated as a parabola, although the same procedure can be applied using alternative functional representations shown in Tab. 2.1.

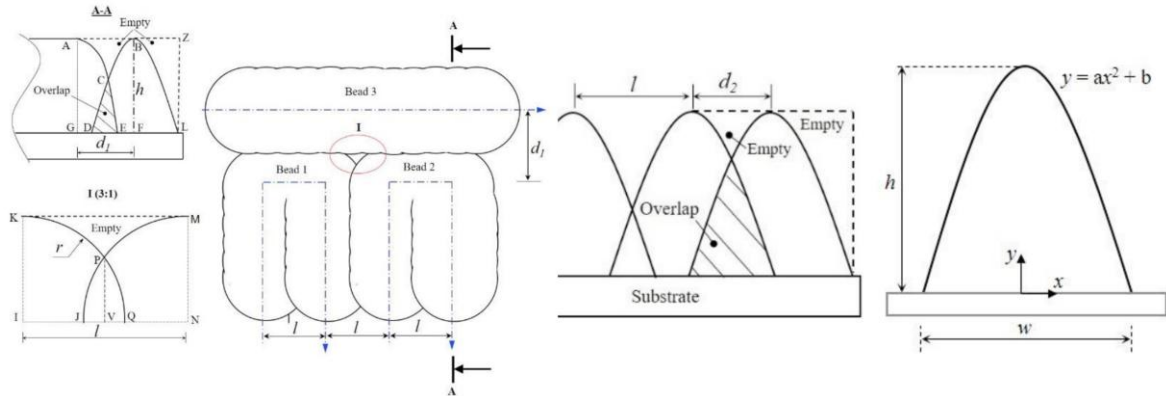


Fig. 2-7 A schematic of multi-bead profile for zigzag and contour pattern using parabolic model [48]

The distances between adjacent welds can be divided into three categories: the distance between inner weld beads, the distance of edge weld beads parallel to the inner beads, and the distance of edge weld beads perpendicular to the inner beads' trajectory. For these three categories, authors Ding [47] and Nguyen [48] performed a series of calculations that define suitable distances for the production of volumetric components. The derived approximations for the parabolic system are as follows: $l \approx 0.738 \cdot w$, $d_1 \approx 0.47 \cdot w$, $d_2 \approx 0.5 \cdot w$, where w is the weld width, l is the distance between internal welds, d_1 is the distance of edge weld beads perpendicular to the inner beads' trajectory and d_2 is distance of edge weld beads parallel to the inner beads.

The above-mentioned approximations determine the theoretically ideal distance between adjacent weld beads to achieve a geometrically accurate part. In practice, however, deviations may occur because newly deposited material tends to be influenced by previously deposited welds. This interaction can shift the weld bead centre toward the adjacent bead, resulting in local geometric deviations within the part [49]. Such inaccuracies are particularly pronounced in trajectory corners, where previously deposited welds often remain at elevated temperatures.

When planning the production of volumetric parts, the initial planning principle is similar to that used for thin-walled structures. The component is first divided into layers. For thin-walled parts, the resulting two-dimensional contour directly defines the welding trajectory for each layer. For volumetric parts, this step is more complex, as multiple deposition strategies can be used to achieve uniform material distribution across the entire layer. Depending on the selected strategy, one or several toolpaths may be generated within a single layer. Upon completion of the welding torch movement along these paths, the corresponding layer is deposited. Typically, a waiting time is introduced between successive layers to allow partial cooling of the deposited material. Commonly used deposition strategies for volumetric WADED parts include raster, zigzag, contour, or spiral trajectories, as well as hybrid approaches, such as, for example, a zigzag infill combined with a single perimeter contour, see Fig. 2-8 [50].

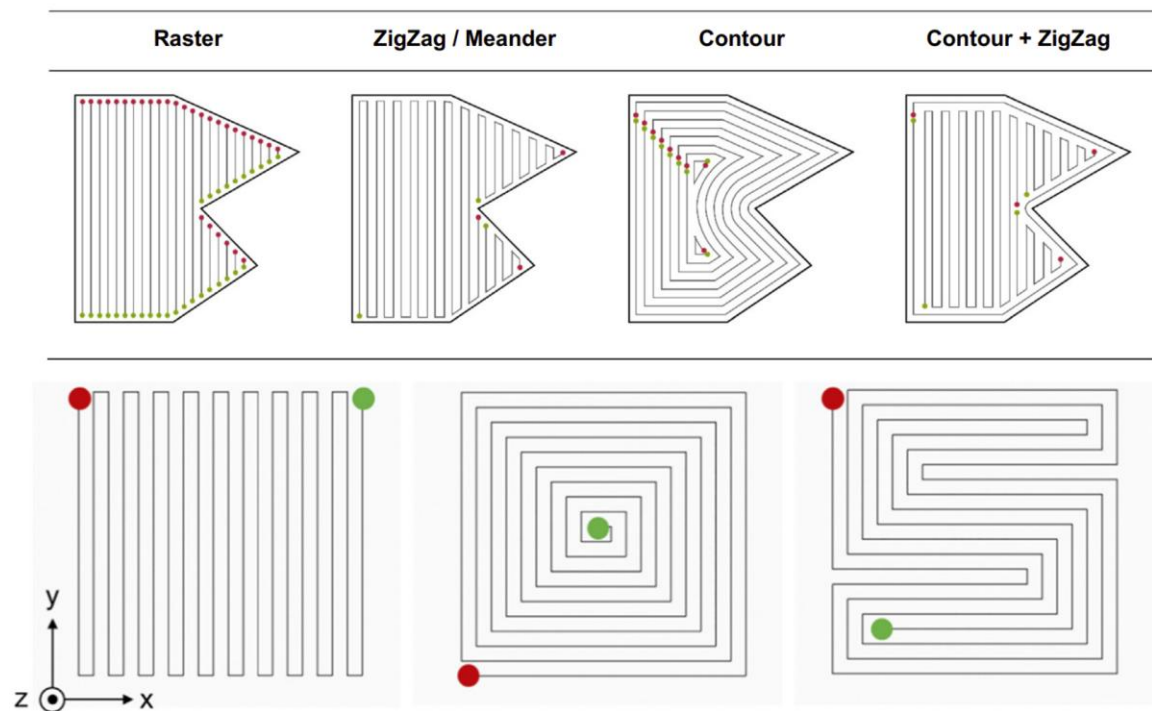


Fig. 2-8 Example of commonly used deposition trajectories for WADED [50]

When selecting a suitable deposition trajectory for WADED, it is necessary to consider that weld starts and ends are generally associated with geometric deviations [34, 50]. Consequently, trajectories with many weld interruptions, such as raster, may be less suitable for WADED. For example, conventional contour trajectories are often replaced by spiral, which enable uninterrupted material deposition [50]. A more recent approach to continuous path planning is the S-pattern trajectory, developed to reduce residual stresses [51].

However, the geometric accuracy of the top surface is not the only one criterion for assessing the suitability of a given deposition strategy. Additional considerations include the resulting microstructure and residual stresses, which represent key challenges associated with WADED manufacturing. Both microstructural evolution and residual stress formation are influenced by the distribution of heat within the manufactured component. Residual stresses arise primarily from local temperature gradients induced by the moving heat source, and more uniform thermal gradients generally lead to lower residual stress levels [52].

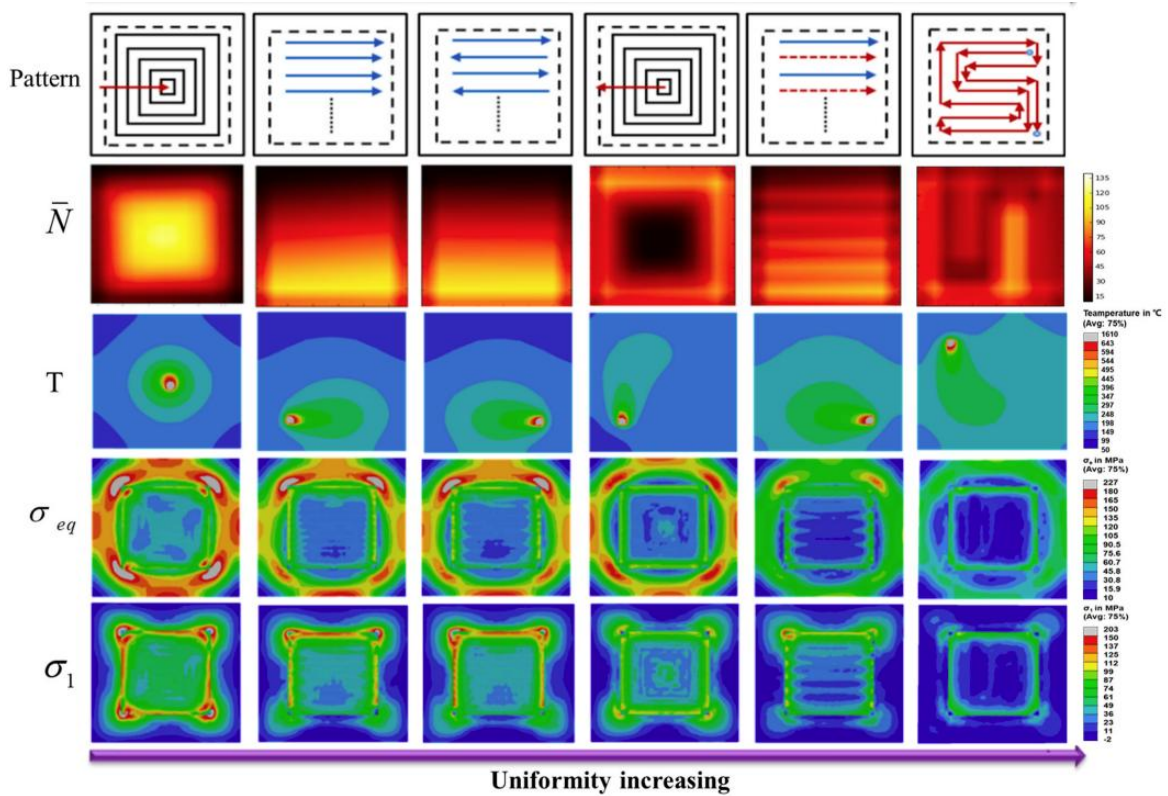


Fig. 2-9 Deposition strategies - temperature distributions and residual stresses [52]

To evaluate heat distribution and residual stress development, Li et al. [52] proposed a bead sequence-driven deposition pattern evaluation method (see Fig. 2-9). This approach enables the comparative assessment of different deposition strategies with respect to their thermal behaviour and the resulting material properties. The bead sequence-driven evaluation method was used to compare ZigZag, Spiral, and S-pattern deposition strategies. The S-pattern exhibited the most uniform heat distribution, followed by Spiral trajectory, while the ZigZag strategy showed the least favourable thermal behaviour.

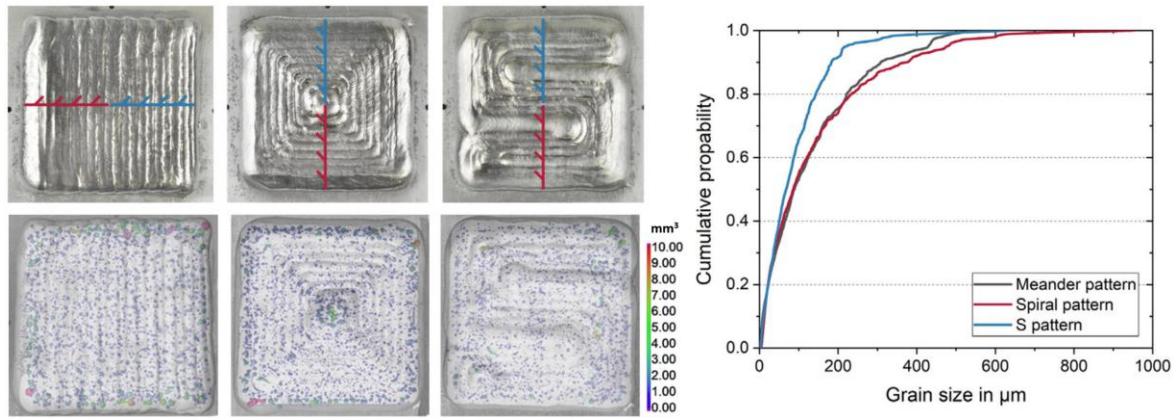


Fig. 2-10 Comparison of WADED deposition strategies using Al-4046 [50]

All trajectories were experimentally manufactured by Köhler using the aluminium alloy Al-4046 [50]. Grain size analysis was performed within a selected region of interest for all strategies (Fig. 2-10, marked in red). Experimental results indicated that, although overall residual stress levels were comparable, local tensile stress peaks occurred at arc start and end locations. Among the evaluated strategies, the S-pattern resulted in reduced porosity, lower residual stresses, and a finer grain structure due to a more homogeneous temperature distribution [50].

2.1.5 Process quality assessment and defects

The quality of a component manufactured using WADED technology reflects a wide range of process parameters and deposition conditions. The fundamental building block of WADED production is the individual weld bead, often referred to as a single deposit or single track. Geometric and metallurgical characteristics of a single deposit directly influence the quality of the final additively manufactured part. Several fundamental parameters are evaluated on a single deposit, including the weld bead width, contact angle, penetration depth, the area of weld penetration below the reference horizontal line corresponding to the bead width (A_{wp}), and the area of weld reinforcement above this line (A_{wr}) [53]. A fundamental requirement for a single deposit is that, in order to ensure adequate interlayer bonding, the underlying layer must be sufficiently remelted along the entire interlayer interface to prevent lack-of-fusion defects [54]. The main measured parameters of a single deposit are shown in Fig. 2-11.

The contact angle represents one of the key geometric parameters of a weld bead. It is influenced by process variables, such as the torch travel speed (TS) and the wire feed speed (WFS), as well as the associated CMT process parameters. From the perspective of layer-by-layer deposition, contact angles greater than 90° are generally considered favourable to ensure stable stacking of successive layers. When depositing adjacent weld beads, as is typical in the fabrication of volumetric parts, larger contact angles, commonly exceeding approximately 115° , are often recommended to promote adequate inter-bead fusion and geometric stability [53].

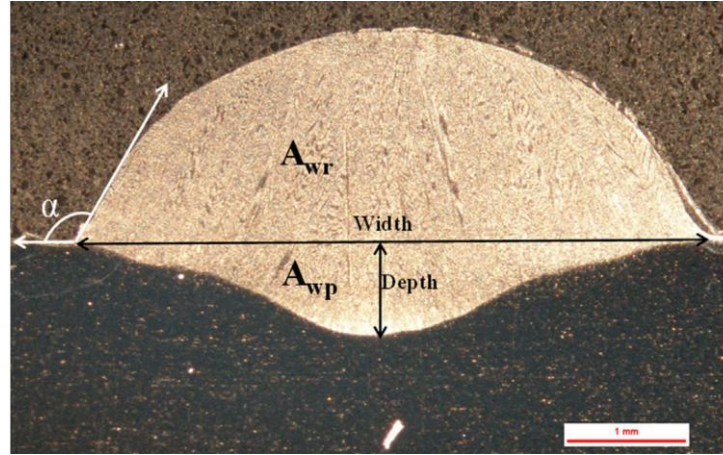


Fig. 2-11 Single deposit geometry characteristics [53]

In addition to directly measurable parameters, calculated parameters are used to describe weld-bead geometry. These include, for example, dilution. Dilution is defined as the ratio of the cross-sectional area of the melted base material to the total cross-sectional area of the fusion zone [55]:

$$Dw = \frac{A_{wp}}{A_{wr} + A_{wp}} \cdot 100 (\%) \quad (2-1)$$

where A_{wp} (mm^2) is the area of weld penetration and A_{wr} is the area of weld reinforcement (mm^2).

The description of the welding process is not defined solely by dimensional variables. The average power of a single CMT cycle, P_c , is therefore commonly used as an indicator of the overall cycle performance [24]:

$$P_c = \frac{1}{n} \sum_{i=0}^n U_{(i)} \cdot I_{(i)} (W) \quad (2-2)$$

where n (-) is the number of cycles, $U(i)$ is the arc voltage (V), and $I(i)$ is the welding current (A).

Based on the average power calculated for a single CMT cycle another key parameter, namely the heat input, can be determined. Heat input represents the amount of thermal energy delivered per unit length of deposition [56]. It is widely used to correlate process parameters with weld bead geometry and to compare different parameter settings. The heat input T_w can be calculated as:

$$T_w = \eta \cdot \frac{P_c}{WS} (J \cdot mm^{-1}) \quad (2-3)$$

where η (-) is the heat transfer efficiency, P_c is the average power (W), and WS is the welding speed or torch travel speed ($mm \cdot s^{-1}$).

The most common defects observed in WADED processes include residual stresses and distortions, porosity, lack of fusion, cracking, humping, and surface irregularities [2, 20, 57–59]. These defects are primarily associated with high thermal gradients, dynamic heat input, and the layer-by-layer nature of arc-based deposition. Their occurrence can significantly affect the dimensional accuracy, mechanical performance, and fatigue behaviour of additively manufactured components. Consequently, ensuring reliable production requires effective process monitoring and systematic parameter optimisation. Svetlizky et al. [57] focused on identifying primary defects in materials processed using the DED method and techniques for characterising them and compiled an overview in Tab. 2.2.

Tab. 2.2 Common defects in DED-processed materials and their characterisation techniques [57]

Defect	Defect's origin	Defect's selected effects on the deposited material properties	Defect's characterization techniques	Mitigation
Residual stresses and distortions	High cooling rates and thermal gradients	Induced phase transformations, loss of geometric tolerance, cracking, delamination, degradation in fatigue behavior, premature failure	Hole drilling, serial sectioning, ring-core drilling, X-ray and neutron diffraction, speed of sound, Barkhausen noise, computational models	Substrate preheating, preheating the chamber and printed part during deposition, optimization of the scan strategy, post-print heat and surface treatments
Porosity	High laser energy density, feedstock porosity, selective evaporation, entrapment of shielding gas, insufficient energy input (LoF)	Degradation of mechanical properties (mainly fatigue life), facilitated crack propagation, anisotropy, reduced corrosion resistance	Archimedes method, ultrasonic pulse-echo velocity measurements, image analysis of metallographic cross-sections, X-ray micro-computed tomography (μ -CT), hard X-rays in synchrotron facilities, gas pycnometry	Process optimization, control of the powder feedstock's composition and quality, HIP post-print treatment
Cracking and delamination	High cooling rates and thermal gradients, either too high or too low energy input, physical and metallurgical properties of the deposited material	Degradation of both static and dynamic mechanical properties, reduced corrosion resistance, premature failure	Metallographic cross-sections, crack opening and fractography in SEM; NDT: magnetic particles, radiography, μ -CT, ultrasonic, computational modeling	Process and geometric optimization, e.g., substrate and chamber preheating, materials compatibility when printing jointly multi-materials
High surface roughness	Low heat input, large powder particles, high laser scanning speeds; a variety of material feedstock, part design, processing, and post-processing conditions and variables	Adverse effects on dimensions and geometric tolerances, degradation of mechanical properties (particularly fatigue)	Contact or non-contact profilometry (including AFM and CLSM), SEM	Increase of heat input, small layer thickness, finer powder particles, post-processing (e.g., HIP and chemical/electrochemical polishing)

A graphical representation of typical defects in WADED is shown in Fig. 2-12.

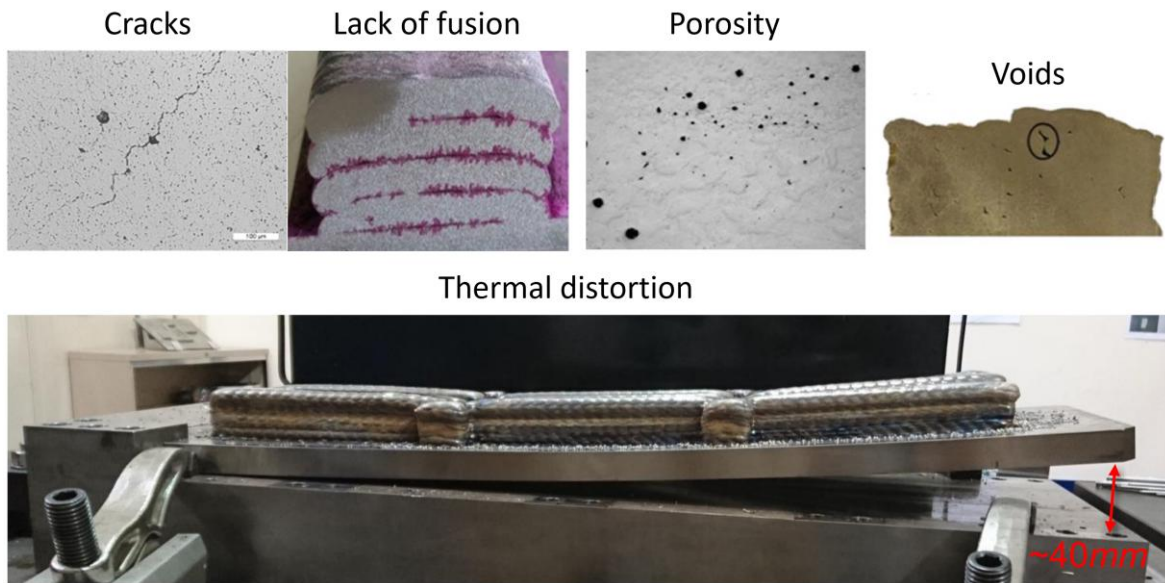


Fig. 2-12 Common defects in WADED [58]

Magnesium is a heat-sensitive and highly reactive metal, which makes WADED processing particularly prone to defect formation. Excessive heat input and local heat accumulation can lead to porosity and microcracking, typically observed at the bottom of weld beads or at interlayer interfaces. While reducing heat input can limit defect formation, it often results in insufficient penetration for reliable WADED fabrication. Comparative studies have shown that more uniform heat distribution can be achieved through appropriate path planning strategies, leading to reduced grain coarsening and porosity [60]. Consequently, effective thermal management is critical for controlling defects and ensuring reliable WADED of magnesium alloys.

2.2 Magnesium alloys

Due to its high chemical reactivity, magnesium is rarely found in its elemental form in nature. It is estimated that magnesium constitutes approximately 2.7% of the Earth's crust and about 0.13% of the world's oceans [61]. Consequently, seawater represents a virtually inexhaustible source of magnesium. Although magnesium does not occur naturally as a metal, its compounds are widely distributed in mineral ores. The most common magnesium-bearing minerals include magnesite (MgCO_3), dolomite ($\text{MgCO}_3 \cdot \text{CaCO}_3$), and carnallite ($\text{MgCl}_2 \cdot \text{KCl} \cdot \text{H}_2\text{O}$) [61]. Metallic magnesium is primarily produced by the electrolysis of molten anhydrous magnesium chloride (MgCl_2), followed by thermal reduction of magnesium oxide (MgO). Magnesium is the lightest of all structural metals and has therefore found widespread application in a broad range of engineering fields [61].

Magnesium is employed across a wide range of industrial sectors, including pyrotechnic, metallurgical, chemical, electrochemical, and structural applications [61]. A significant portion of primary magnesium production is consumed as an alloying or microalloying element in ferrous and non-ferrous metal systems. Approximately one-third of the global magnesium output is used for the production of magnesium alloys [61]. In recent decades, magnesium production has increased steadily, reflecting the growing demand for lightweight materials, as illustrated in Fig. 2-13 [62].

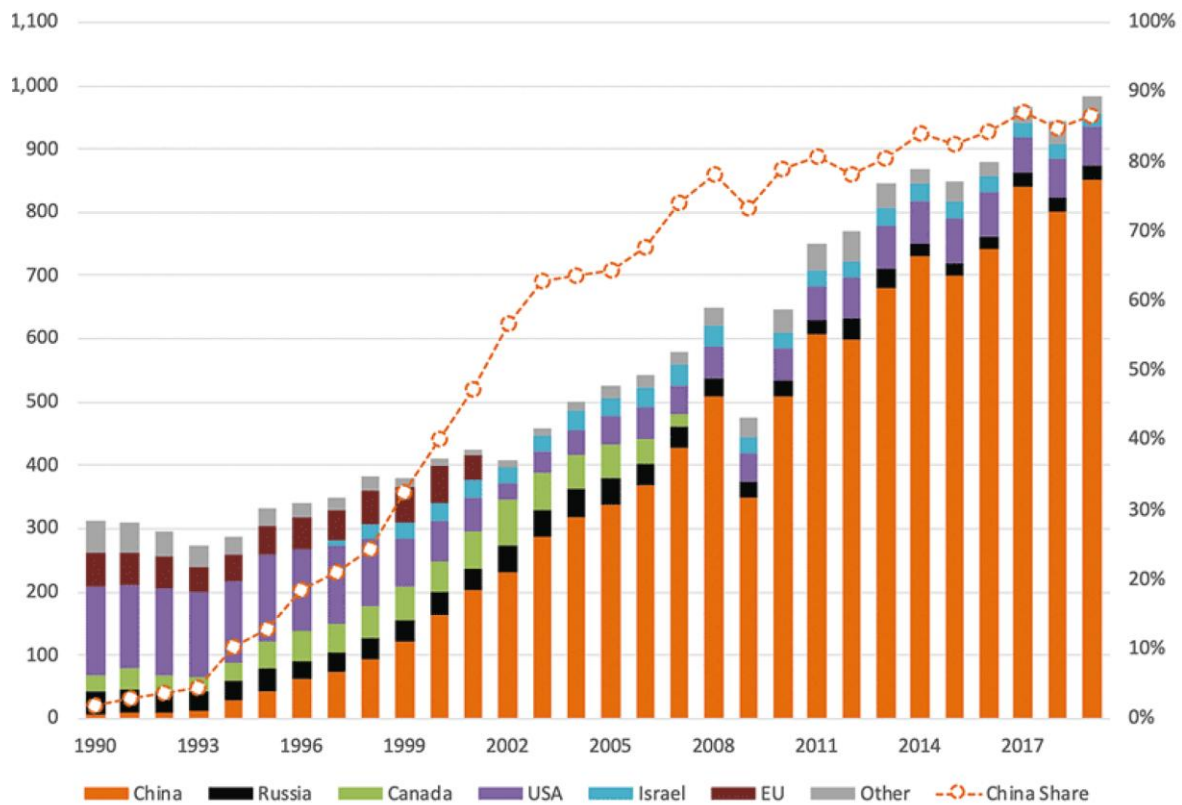


Fig. 2-13 Worldwide magnesium production by country, 1990–2019 (kt) [62]

The basic properties of pure magnesium at room temperature and atmospheric pressure are: solid state, hexagonal (six-sided) crystallographic system, density $1738 \text{ kg}\cdot\text{m}^{-3}$, melting point $650 \text{ }^{\circ}\text{C}$, boiling point $1090 \text{ }^{\circ}\text{C}$, thermal conductivity $156 \text{ W}\cdot\text{m}^{-1}\cdot\text{K}^{-1}$, specific heat capacity $26.19 \text{ J}\cdot\text{mol}^{-1}\cdot\text{K}^{-1}$ [61].

The low density and high strength-to-weight ratio of magnesium alloys make them attractive for weight-critical structural applications. Recent studies report that substituting aluminium and steel components with magnesium alloys in automotive applications can achieve weight reductions of approximately 20–25% and 45–50%, respectively, while maintaining structural performance [4]. Beyond the automotive sector, magnesium alloys are increasingly considered for aerospace applications due to their potential to reduce fuel consumption [4, 5]. They are consequently regarded as promising materials for environmentally sustainable engineering [5, 6].

Magnesium alloys are designated according to ASTM B951, in which the first one or two letters identify the principal alloying elements and the subsequent numerals indicate their approximate weight percentages. Based on their dominant alloying elements, magnesium alloys are commonly grouped into series, including:

- AZ series (aluminium, zinc)
- AM series (aluminium, manganese)
- ZK series (zinc, zirconium)
- AE series (aluminium, rare earth elements)
- WE series (yttrium, rare earth elements)

The final properties of magnesium alloys depend strongly on their alloying elements. Rare-earth alloying enhances strength and creep resistance, aluminium improves strength, ductility, and corrosion resistance, and zinc promotes grain refinement and solid-solution strengthening [63]. Excessive additions of alloying elements may lead to the formation of brittle phases and reduced corrosion resistance.

2.2.1 Welding of magnesium alloys

Reliable welding of magnesium alloys is a prerequisite for WADED manufacturing and can be achieved using conventional welding processes, such as MIG and TIG, operated in a protective atmosphere. Most commonly, shielding gases are argon, helium, or their mixtures. Due to magnesium's low specific heat capacity, welding is typically associated with a relatively low heat input [61]. Weldability is strongly influenced by alloy composition, particularly the content of alloying elements. Rapid cooling during welding may promote crack formation, which can be mitigated by reducing thermal gradients through preheating of the base material. An overview of the weldability of selected magnesium alloys and the corresponding recommended preheating temperatures is provided in Tab. 2.3 [61].

Tab. 2.3 Weldability, heat treatment and preheating temperatures for welding of magnesium alloys [61]

Alloy	Weldability	Heat treatment	Preheating temperature(°C)
AZ81A	B+	T4	175-400
AZ91C	B+	T4	175-400
AZ61A	B	-	-
AZ63A	C	T4	175-380
AZ31C	A	-	-
WE43	B-	T4 nebo T6	300

2.2.2 Mechanical properties

The alloying elements content, grain size, and the phase distribution mainly determine the mechanical properties of magnesium alloys. Grain size and phase distribution are strongly influenced by thermal history and processing conditions.

Tab. 2.4 Chemical composition and mechanical properties of selected cast magnesium alloys [61]

Alloy	Chemical composition	Tensile strength (MPa)	Yield strength (MPa)	Ductility (%)
AM60A-F	6,0Al-0,13Mn	205	115	6
AZ91A-F	9,0Al-0,13Mn-0,7Zn	230	150	3
AZ63A-T6	6,0Al-3,0Zn-0,15Mn	275	130	5
AZ91C-T6	8,7Al-0,13Mn-0,7Zn	275	145	6
AZ92A-T6	9Al-2Zn-0,1Mn	275	150	3
ZK61A-T6	6,0Zn-0,7Zr	310	195	10
EZ33A-T5	3,3RE-2,7Zn-0,6Zr	160	110	2

The typical chemical composition and mechanical properties of magnesium alloys produced by casting are shown in Tab. 2.4 and the properties of wrought alloys in Tab. 2.5.

Tab. 2.5 Chemical composition and mechanical properties of selected wrought magnesium alloys [61]

Alloy	Chemical composition	Tensile strength (MPa)	Yield strength (MPa)	Ductility (%)
AZ31B-F	3,0Al-1,0Zn	260	200	15
AZ61A-F	6,5Al-1,0Zn	310	230	16
AZ80A-T5	8,5Al-0,5Zn	380	240	7
AZ82A-T5	8,5Al-0,5Zn	380	275	7
AZ31B-H24	3,0Al-1,0Zn	290	220	15

Commercial magnesium welding wires are typically available in AZ31, AZ61, AZ91 or WE43 compositions. AZ61 was selected for this work because it offers a favourable combination of mechanical properties and weldability. Although its weldability is slightly lower than that of AZ31 and AZ91, it should remain sufficient for WADED processing while providing improved mechanical performance.

2.2.3 Microstructure AZ-series magnesium alloys

The microstructure of a material plays a key role in determining its mechanical properties. In general, finer microstructures are associated with improved mechanical performance. The relationship between grain size and strength is commonly described by the Hall–Petch relationship, which implies that smaller grains act as barriers to dislocation motion, thereby increasing strength [64]. Faster cooling of the melt pool increases nucleation rate, resulting in finer microstructures [65].

In addition to grain size, the type and distribution of precipitates within the microstructure are also important. During the solidification and cooling of AZ-series magnesium alloys, eutectic phases are most commonly formed in the form of the intermetallic compound $Mg_{17}Al_{12}$. The volume fraction of these precipitates increases with aluminium content, as observed in the microstructures of AZ31 (3 wt.% Al), AZ61 (6 wt.% Al), and AZ91 (9 wt.% Al) alloys, as shown in Fig. 2-14 [66]. These precipitates contribute to strengthening by impeding dislocation motion and restricting slip along grain boundaries.

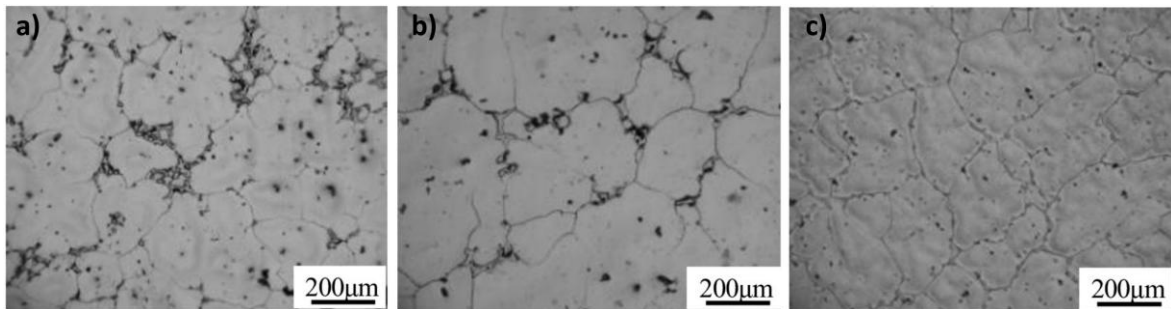


Fig. 2-14 Microstructure of AZ magnesium alloys: a) AZ91, b) AZ61, c) AZ31 [66]

Grain size in AZ-series alloys is strongly affected by cooling rate and thermal history. Higher cooling rates usually promote finer microstructures, as observed on AZ91 alloy [67]. WADED is a process in which cooling rates are very high, reaching several hundred degrees per second [68], which is advantageous in terms of microstructure formation. However, repeated thermal cycling and prolonged exposure to elevated temperatures may lead to grain coarsening.

The resulting microstructure is also significantly influenced by post-processing, such as heat treatment. A higher grain growth rate leads to faster microstructural coarsening during thermal exposure, which results in reduced mechanical strength [69]. Fig. 2-15 illustrates the dependence of grain size on heat-treatment temperature and holding time for AZ31, AZ61, and AZ91 alloys.

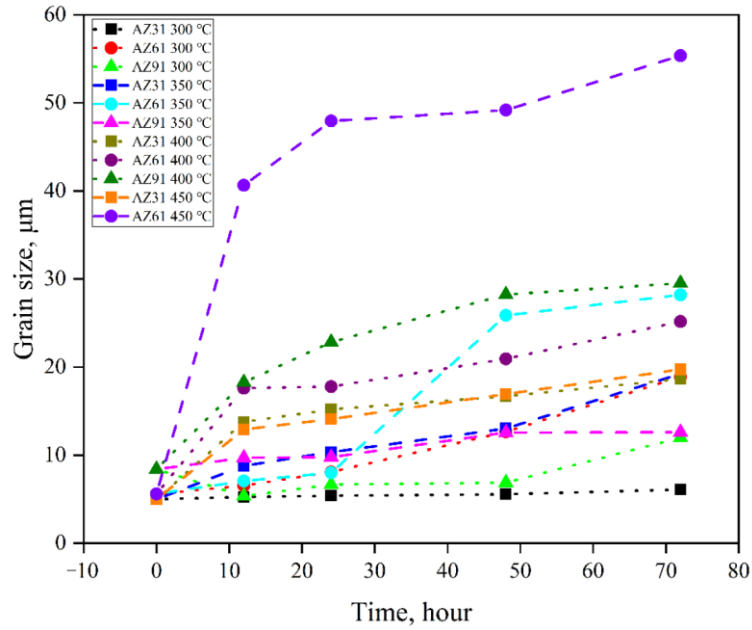


Fig. 2-15 Effect of heat treatment temperature and time on grain size of AZ31, AZ61, and AZ91 alloys [69]

2.2.4 Thermal and physical properties

The thermophysical properties of magnesium alloys are strongly influenced by their hexagonal close-packed (HCP) crystal structure, leading to significant anisotropy [70]. The key thermophysical properties include thermal conductivity, which describes a material's ability to conduct heat, and specific heat capacity, which is the amount of heat required to raise the temperature of a unit mass of material by 1 degree. Magnesium alloys are characterised by relatively high thermal conductivity and low specific heat capacity. These properties significantly affect heat transfer rates, melt pool behaviour, and the development of thermal gradients during additive manufacturing.

Grain size influences the thermal conductivity of magnesium alloys primarily through the presence of precipitates and grain boundaries. Grain boundaries act as barriers to heat transport. Therefore, a higher grain boundary density, typical of fine-grained microstructures, reduces the effective thermal conductivity [71]. The thermophysical properties of magnesium alloys are further affected by chemical composition. In Mg–Al alloys, aluminium content plays an important role in the temperature dependence of thermal conductivity. For AZ31, AZ61, and AZ91 alloys, both thermal conductivity and specific heat capacity increase with increasing temperature [70, 72, 73].

2.3 WADED of magnesium alloys

The application of WADED to magnesium alloys represents a relatively new and still limited area of research. Nevertheless, a growing number of publications have appeared in recent years, indicating increasing interest in this technology and its potential for processing magnesium alloys, as illustrated in Fig. 2-16 [15]. This chapter presents the key publications in this field, based on which the research gaps have been identified.

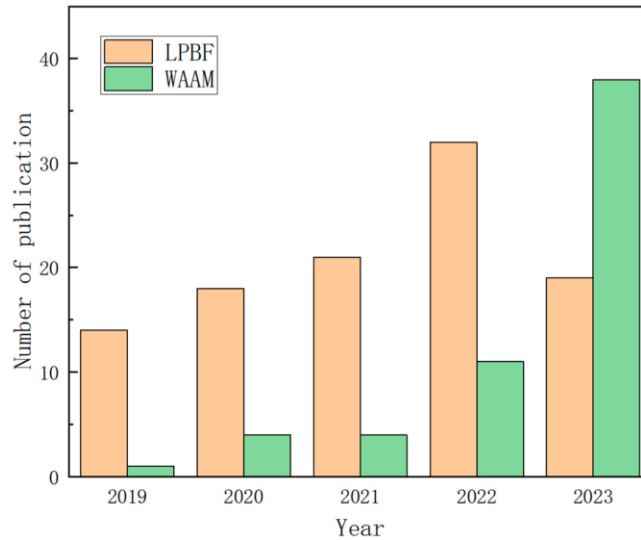


Fig. 2-16 Publication trends in additive manufacturing of Mg alloys [15]

One of the earliest studies in this field was conducted by Guo et al. [11], who investigated grain refinement in WADED of magnesium alloys by adjusting the pulse frequency of a GTAW heat source. The study shown that optimisation of the pulse frequency within the range of 1–500 Hz can influence microstructure and mechanical properties. Tensile strength values of 221–263 MPa and grain sizes of 21–39 μm were reported. The finest microstructure and the highest mechanical strength were obtained at a pulse frequency of 10 Hz. The microstructure of the material at different frequencies is shown in Fig. 2-17.

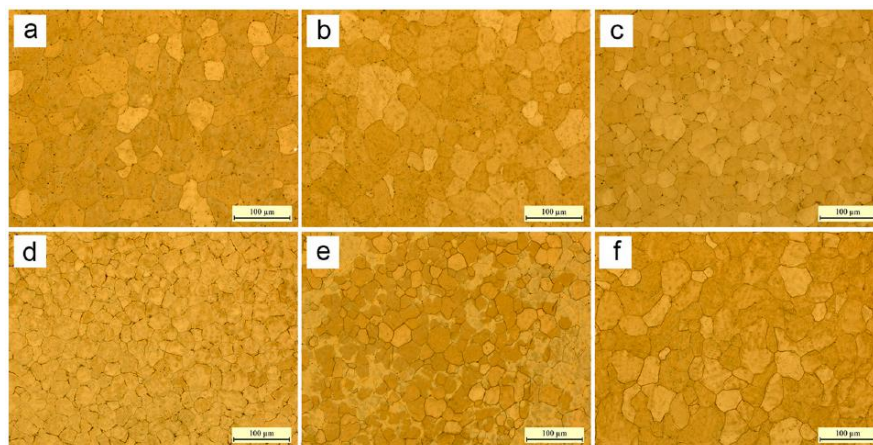


Fig. 2-17 Microstructure for pulse frequency: a) 500 Hz, b) 100 Hz, c) 10 Hz, d) 5 Hz, e) 2 Hz, f) 1 Hz [11]

In addition to Guo et al., several other authors have investigated GTAW-based WADED processing of magnesium alloys. Studies on the AZ31 alloy have shown that appropriate control of process parameters and thermal conditions promotes the formation of fine, predominantly equiaxed grains, thereby improving mechanical properties and corrosion performance [74]. Under optimised conditions, porosity as low as 0.01 % and an average grain size of approximately 24.7 μm were reported [74].

Further work focused on optimising GTAW parameters for the AZ61 alloy demonstrated the formation of a fine-grained microstructure with a low defect density [75]. Reported ultimate tensile strength values reached approximately 260 MPa, while elongation values of up to 16 % were achieved, indicating a favourable balance between strength and ductility [75].

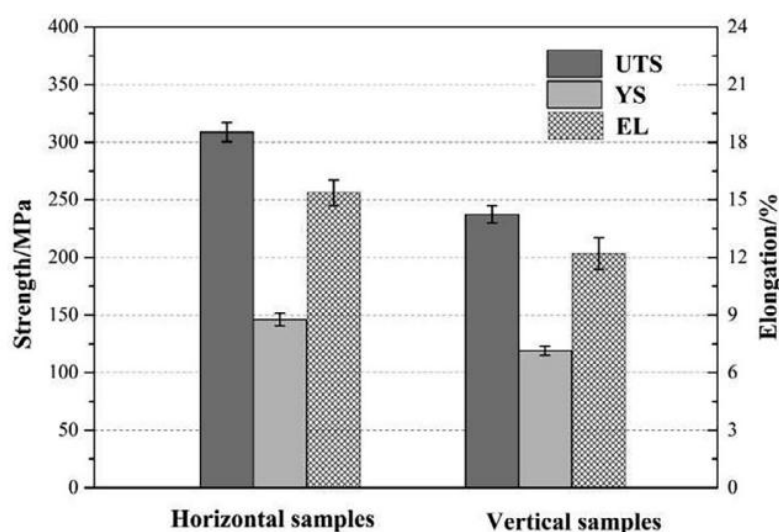


Fig. 2-18 Tensile properties for the AZ80M alloys fabricated by WADED [76]

Investigations of higher-strength alloys such as AZ80M fabricated using GTAW-WADED demonstrated ultimate tensile strengths exceeding 300 MPa [76]. However, these samples exhibited significant anisotropy, see Fig. 2-18. The mechanical performance was strongly linked to microstructural heterogeneity and defect distribution, highlighting the increasing challenges associated with WADED processing of magnesium alloys with higher aluminium content [76].

GTAW-based WADED processes have demonstrated favourable microstructural and mechanical properties. However, the requirement for torch rotation to maintain a consistent wire-feeding direction reduces their attractiveness for large-scale manufacturing [2, 13]. Consequently, the following studies focus on GMAW or CMT-based WADED approaches for the production of magnesium alloys.

Wang et al. [3] investigated WADED processing of AZ31 magnesium alloy using a cold metal transfer (CMT) heat source. Three different CMT characteristic settings were examined, and arc current and voltage signals were recorded to calculate the average power and effective heat input for each cycle. The geometry of weld beads and thin-walled deposits was analysed, revealing a strong dependence on the selected CMT characteristic. Lower heat input produced narrower beads and reduced penetration [3]. Mechanical testing revealed significant anisotropy as shown in Fig. 2-19d, e. The observed anisotropy was attributed to directional grain growth, porosity, and the layer-by-layer thermal history inherent to the CMT-based WADED process.

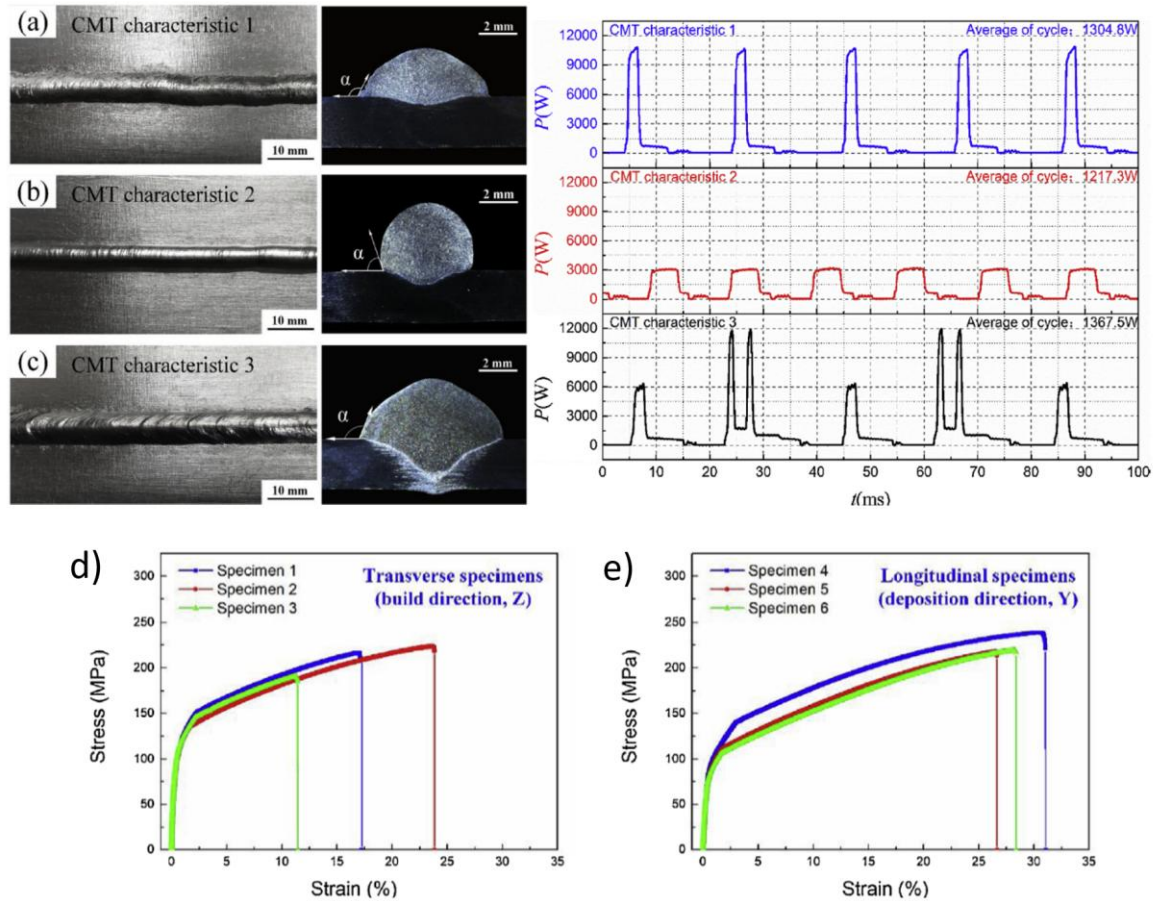


Fig. 2-19 CMT-based WADED of AZ31: a,b,c) CMT characteristic energy power, d,e) mechanical prop. [3]

Zhang et al. [77] investigated CMT-based WADED processing of the AZ31 magnesium alloy, focusing on the influence of welding speed ($8\text{--}14\text{ mm}\cdot\text{s}^{-1}$) on microstructural evolution. The fusion zone was characterised by fine dendritic α -Mg and dispersed β -Mg₁₇Al₁₂ precipitates, while increased porosity was observed at the highest welding speed of $14\text{ mm}\cdot\text{s}^{-1}$ [77]. In the heat-affected zone, grain coarsening and β -phase precipitation were detected. At a welding speed of $8\text{ mm}\cdot\text{s}^{-1}$, the β -phase formed continuous bands along grain boundaries, leading to liquation cracking. The higher welding speeds resulted in a more dispersed β -phase morphology and the suppression of liquation cracks [77].

A further investigation addressed CMT-based WADED processing of the AZ31 magnesium alloy, focusing on the influence of process parameters (TS, WFS, current, voltage) on bead geometry, microstructure, and mechanical properties [78]. The results showed that wire feed rate and travel speed strongly affect bead width, height, and penetration, thereby governing the stability of layer formation. Microstructural analysis revealed predominantly columnar α -Mg grains with β -Mg₁₇Al₁₂ precipitates distributed along grain boundaries. The average grain size at the interface zone, centre, and top portion of single tracks was 35, 42, and 60 μ m, respectively. Mechanical testing indicated that tensile properties in the travel direction were UTS 222 MPa, YS 102 MPa, EL 18%, and in the build direction were UTS 202 MPa, YS 110 MPa, EL 14% [78]. The influence of wire feed speed and travel speed on the geometry of the single deposit is shown in Fig. 2-20.

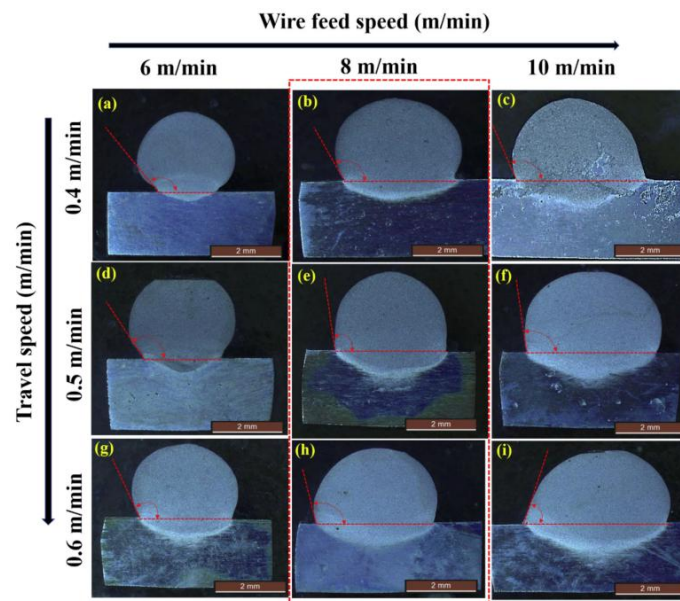


Fig. 2-20 Cross section of single deposit beads at different of WFS and TS [78]

Yang et al. examined the fabrication of AZ31 magnesium alloy by CMT-based WADED with emphasis on microstructural characteristics and mechanical performance [35]. The deposited material exhibited a layered microstructure composed mainly of columnar α -Mg grains, with β -Mg₁₇Al₁₂ precipitates and some dispersive η -Al₈Mn₅ phases. Mechanical testing revealed tensile strengths comparable to wrought AZ31 (see Tab. 2.6), while clear anisotropy between the build and deposition directions was observed [35].

Tab. 2.6 Mechanical properties AZ31 - CMT, wrought, cast [35]

Conditions	Ultimate tensile strength (MPa)	Yield strength (MPa)	Elongation (%)
WAAM-CMT AZ31 (Building direction)	210.5 $\pm$ 3.5	131.6 $\pm$ 4.2	10.55 $\pm$ 1.61
WAAM-CMT AZ31 (Travel direction)	151.9 $\pm$ 12.9	71.2 $\pm$ 4.5	7.54 $\pm$ 1.32
Wrought AZ31 (ASTM standard B91-12)	234	131	6
Cast AZ31 [34]	152	71	4.8

This anisotropic behaviour was attributed to directional solidification, heterogeneous grain morphology, and the thermal cycling inherent to the layer-by-layer CMT-based WADED process. The in-plane anisotropies (%) of the UTS, YS, and EL for the AZ31 deposit are 0.278, 0.459, and 0.285, respectively [35]. The mechanical properties of the manufactured thin-walled object are shown in Fig. 2-21.

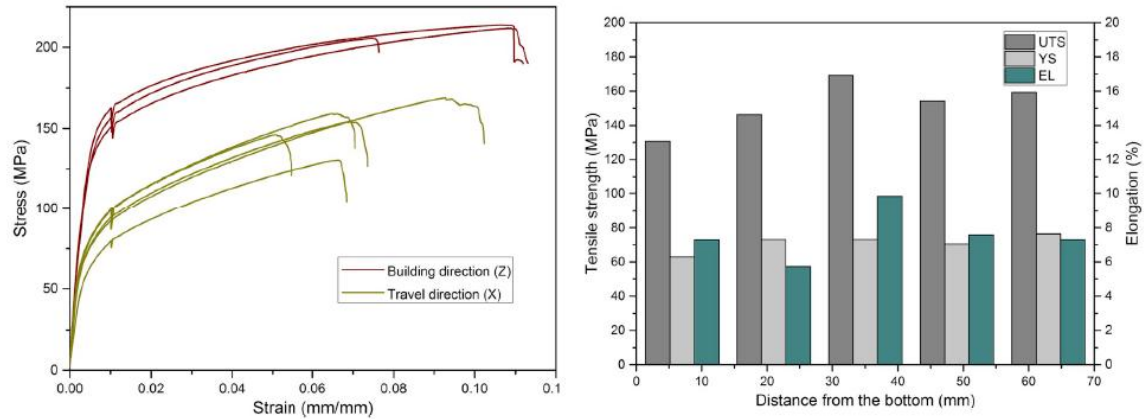


Fig. 2-21 Mechanical properties CMT-WADED magnesium alloy AZ31 [35]

Another commercially available magnesium alloy is AZ91, whose WADED fabrication was investigated by Bi et al. [79]. The study focused on the production of thick-walled components using a weaving deposition strategy to improve material distribution and geometric stability. Mechanical testing revealed only minor anisotropy, with ultimate tensile strengths of approximately 245 MPa in the transverse direction and 250 MPa in the longitudinal direction, corresponding to a difference of about 2 % [79]. The low degree of anisotropy was attributed to the deposition strategy and the resulting homogenisation of the microstructure.

Another contribution to this field was made by the LKR research group [38, 80]. Klein's pilot study [80] investigated CMT-based WADED fabrication of AZ61, achieving yield strength (YS) of 99.2 MPa (longitudinal) and 104.4 MPa (transverse), and ultimate tensile strength (UTS) of 256.4 MPa and 264.1 MPa, respectively. The results showed crack-free deposits with a fine, predominantly equiaxed, and homogeneous microstructure, leading to only weak mechanical anisotropy and properties superior to those of the cast material. Gneiger et al. [38] focused on developing a magnesium alloy specifically for wire arc additive manufacturing. The alloy was engineered to achieve high strength while maintaining sufficient process stability and resistance to cracking during WAAM processing. Compared to parts made with commercially available filler wires, the newly developed alloy exhibited higher mechanical performance, increasing YS by approximately 9 MPa and UTS by about 25 MPa, reaching UTS 299 MPa in the welding direction and 298 MPa in the transversal direction [38].

Yang et al. [81] investigated the WADED processing of a novel ATZM31 magnesium alloy with emphasis on microstructural evolution and mechanical performance. The deposited material exhibited a refined, more homogeneous grain structure compared to conventional AZ-series alloys, attributed to the modified alloy composition's effect on solidification behaviour [81]. Tensile testing showed ultimate tensile strengths of 225 MPa and 214 MPa along the deposition and build directions, respectively, while the elongations are 15.8% and 18.8%, respectively [81]. These results demonstrate that targeted alloy design can substantially enhance both the processability and mechanical performance of magnesium alloys in WADED. The detailed grain-size distribution of the investigated samples is shown in Fig. 2-22.

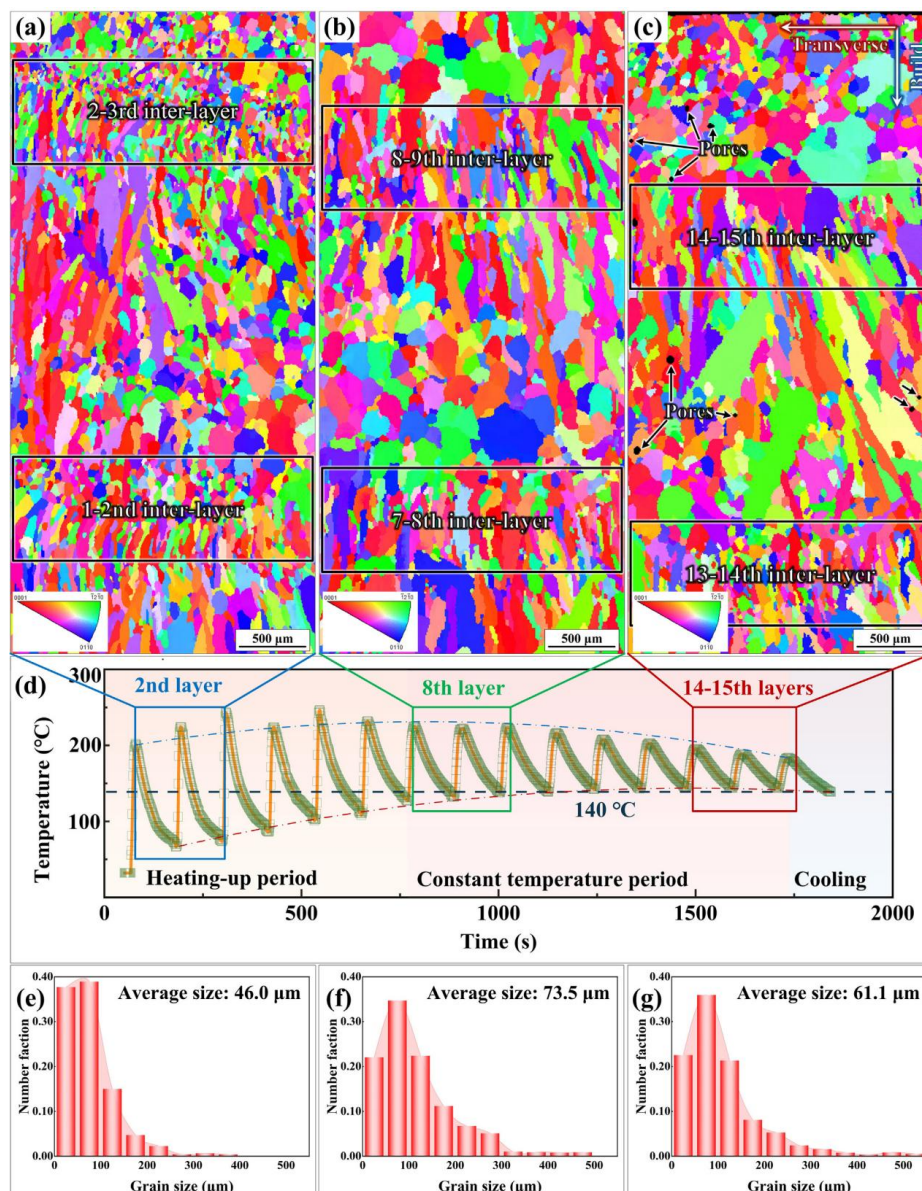


Fig. 2-22 Detailed grain-size distribution in ATZM31 magnesium alloy [81]

A study summarising findings from major publications in the field, focusing on WADED production of magnesium alloys, was presented by Hao et al. [59]. Fig 2-23 summarises the mechanical properties reported across these studies and compares them with those of cast and wrought magnesium alloys.

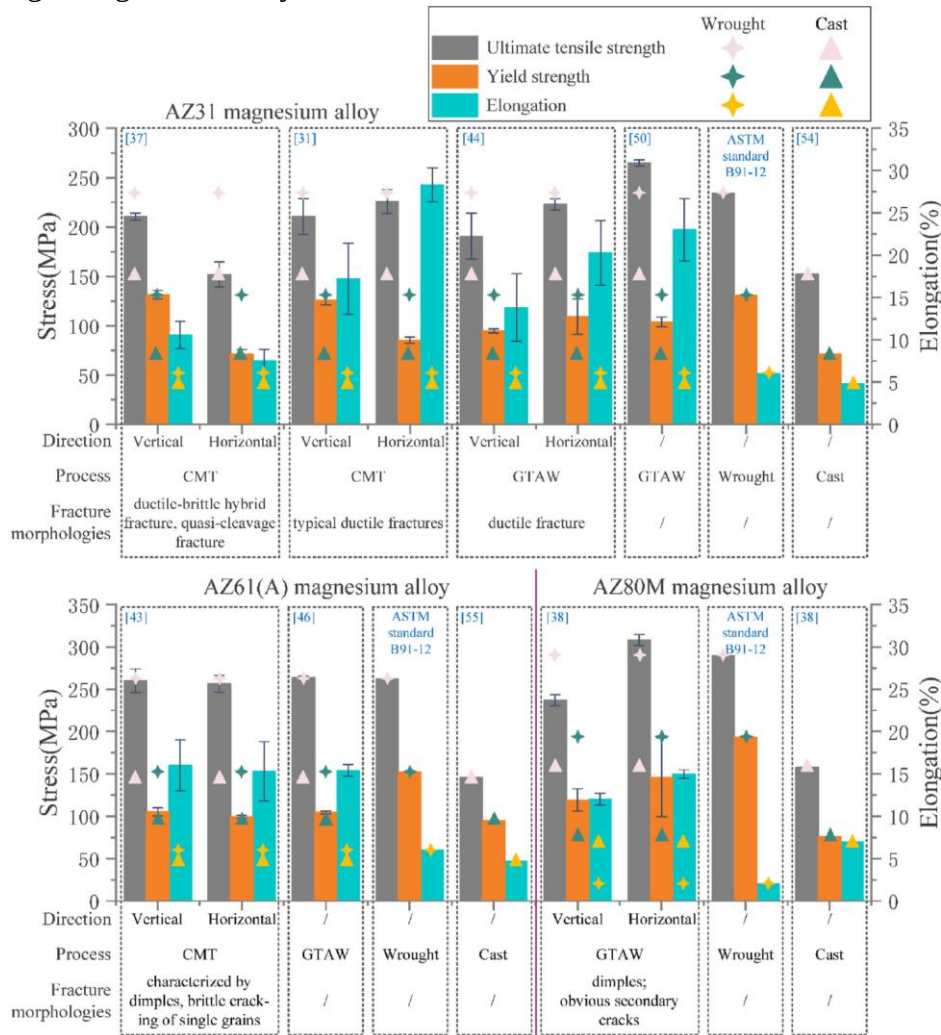


Fig. 2-23 WADED mechanical properties comparison [59]

In addition to the studies presented above, a few studies in the literature deviate from the mainstream focus on WADED of AZ-series magnesium alloys using GTAW or CMT. Fukang et al. [59] investigated CMT-based processing of the WE43 alloy, comparing conventional CMT with advanced variants including CMT Pulse, CMT Advanced, and CMT Pulse Advanced. Cai et al. [82] focused on the development of CMT process parameters for the repair of surface defects in ZM6 magnesium alloy. Beyond optimising conventional parameters such as WFS and TS, particular attention was paid to thermal management via substrate preheating, resulting in improved process stability. Han et al. [83] examined the optimisation of process parameters and corrosion resistance of AZ91 alloy using plasma arc welding. While these studies provide valuable insights into alternative alloys, processes, and applications, their scope and objectives differ from the present work's focus, and therefore, their direct applicability is limited.

3 ANALYSIS AND CONCLUSION OF LITERATURE REVIEW

The following section provides a critical analysis of the current state of knowledge derived from research publications in the field of wire arc directed energy deposition of magnesium alloys. It synthesises key findings on deposition path-planning strategies and process parameters, with particular emphasis on their influence on process stability, geometrical accuracy, and material behaviour. For each thematic area, the main conclusions drawn from the literature are summarised, followed by the identification of unresolved issues that define the research gaps addressed in this dissertation.

3.1 Deposition path planning

Reliable fabrication of complex WADED components requires controlled production of characteristic features such as thin walls, overhangs, and volumetric regions. In all cases, the weld bead represents the fundamental building block, and its geometry directly governs dimensional accuracy. The literature reports that weld starts and ends are critical locations where local thermal variations cause irregularities in bead height [34, 36]. These defects may accumulate across successive layers, leading to excessive geometric deviations. Overhang fabrication further amplifies these challenges due to the lack of underlying support, requiring reduced heat input or modified process parameters to prevent bead sagging or collapse [41–44]. Volumetric parts are mainly affected by heat accumulation and slowly cooling regions, which promote grain coarsening and residual stress development. Although several deposition strategies have been proposed to improve heat distribution, their effects on fabrication accuracy have been evaluated primarily for aluminium alloys [52].

Residual stresses represent a key challenge across most additive manufacturing processes. To suppress residual-stress-induced deformations conventional welding practice divides long weld beads into shorter segments deposited in controlled sequences to reduce heat accumulation and residual stresses [39, 40]. In the context of WADED of volumetric parts, Li et al. [51] examined the influence of deposition strategies on residual-stress evolution using numerical simulation. Their results demonstrated that deposition strategy can significantly homogenise heat distribution and mitigate stress development in volumetric components [51].

Knowledge gap - Deposition path planning:

Despite the demonstrated influence of deposition path planning on heat distribution and geometric accuracy, no study has systematically investigated bead sequencing strategies for thin-walled WADED components to eliminate weld start and end irregularities while simultaneously evaluating their effects on residual stress development.

3.2 Key process-related factors for magnesium alloy

Process stability and defect formation during WADED of magnesium alloys are primarily influenced by metal transfer behaviour, heat input control, and thermal management [82]. Magnesium's high reactivity and thermal sensitivity make the process prone to porosity, cracking, and lack of fusion when heat input is either excessive or insufficient [59, 84]. Stable deposition requires a narrow and carefully balanced processing window that ensures sufficient penetration while limiting local heat accumulation. Most studies on CMT-based WADED of magnesium alloys focus on the influence of fundamental parameters such as wire feed rate and travel speed, or on the development of specialised alloys [38, 78, 82, 85]. Only a limited number of works address CMT parameters at the characteristic level (e.g. boost phase current, boost phase duration), and these parameters are typically reported rather than systematically analysed [3]. Nevertheless, parameters such as boost-phase current, boost duration, and electrode motion during the CMT cycle are known to strongly influence metal transfer behaviour, bead geometry, and penetration, as demonstrated for aluminium alloys [86]. In addition, process parameters, including wire feed rate, travel speed, shielding gas, stick-out, and thermal management, also significantly affect arc stability and defect formation [78, 82].

Knowledge gap - Key process-related factors for magnesium alloy:

Despite the growing number of publications on CMT-based WADED, no study has yet provided a systematic analysis of how individual CMT characteristic parameters control bead geometry, metal transfer behaviour and process stability, which currently limits the ability to optimise WADED processing of magnesium alloy.

3.3 Thermal management

Thermal management is identified as a key factor influencing microstructural evolution and residual stress development during WADED [2, 19, 87]. Due to the high thermal conductivity and low specific heat capacity of magnesium alloys, heat is rapidly redistributed during deposition. These conditions strongly affect solidification behaviour, grain size, and defect formation [84], particularly in multi-layer and volumetric components, where heat accumulation becomes significant.

Excessive heat accumulation and regions with reduced cooling rate are reported to promote grain coarsening and increase the probability of defect formation, such as porosity and microcracking, especially at interlayer interfaces [60, 88]. In contrast, the higher cooling rates typical of thin-walled structures favour finer microstructures. Preheating and increased interpass temperatures are considered effective methods for reducing thermal gradients and residual stresses [82]. However, they simultaneously decrease cooling rates and promote grain growth. Residual stresses arise primarily from local temperature gradients induced by the moving heat source [50, 52]. A more uniform temperature distribution is generally associated with lower residual stress levels. Therefore, special path-planning strategies, such as the S-pattern, aimed at equalising thermal loads, were developed for the production of volumetric parts [50, 52].

Knowledge gap – Thermal management:

Although thermal management is recognised as crucial, no systematic study has evaluated the combined influence of preheating and interpass temperature on microstructural development and residual stress evolution during WADED of AZ61 magnesium alloy components.

4 AIM OF THE THESIS

This dissertation aims to investigate wire arc directed energy deposition (WADED) as a manufacturing approach for structured components fabricated from magnesium alloys. The work focuses on the development and evaluation of WADED processing strategies, including deposition path planning, thermal management, and key process parameters. The objective is to enhance the geometrical stability, microstructure, and mechanical properties, while mitigating the formation of porosity and residual stresses in WADED-fabricated magnesium alloy parts. To achieve the main goal of this thesis, the necessary sub-objectives are as follows:

- Development and implementation of an experimental WADED setup suitable for processing magnesium alloys.
- Analysis of the influence of deposition path planning on component geometry, residual stress development, and the buildability of inclined and thin-walled structures.
- Identification of key process-related factors governing the WADED fabrication of magnesium alloy components.
- Evaluation of the effect of welding parameters and thermal conditions, including base material preheating and interpass temperature, on porosity formation and material properties.
- Characterisation of the resulting microstructural features and mechanical properties of WADED-fabricated magnesium alloy components.
- Assessment of the suitability of WADED for the fabrication of structured magnesium alloy components with improved material efficiency.

4.1 Scientific questions & Hypothesis

Based on a critical analysis of the available literature, the following scientific questions were identified. Each scientific question was subsequently tested through working hypotheses formulated based on the current state of the art.

Q1: How does deposition path planning influence the geometrical accuracy and residual stress of components fabricated by WADED?

H1a: Available studies show that during WADED production of thin walls, material accumulation and unacceptable height irregularities in the manufactured object may occur [34, 35, 89]. Weld starts and ends are often associated with local variations in thermal conditions and material deposition, resulting in either excessive or insufficient bead height [34, 36]. These local geometric deviations can accumulate during layer-by-layer fabrication, especially when the weld start and end locations are repeatedly situated in the same regions across the component's height [34, 36]. Therefore, deposition strategies that redistribute or alternate weld start and end locations along the component length are expected to promote a more uniform layer height and improved geometrical stability.

H1b: Residual stresses are a significant source of undesirable effects in welded structures, including excessive distortions, delamination or reduced fatigue properties and fracture resistance [2, 20, 90]. In conventional welding practice, long weld beads are often divided into shorter ones and welded in specific sequences to redistribute heat input and reduce residual stress accumulation [39, 40]. Therefore, it is assumed that a similar mechanism implemented in WADED path planning for a 120 mm-long thin wall could lead to a more uniform distribution of heat across the entire object. Consequently, strategies that promote more uniform heat distribution during deposition are expected to reduce the residual stresses in thin-walled WADED-fabricated components.

Q2: What are key process-related factors and how they affect process stability and defect formation during WADED of AZ61 magnesium alloy?

H2: The stability of the welding process is key to achieving WADED production without excessive defects [12, 20]. The low specific heat capacity and high thermal conductivity of magnesium alloys lead to rapid heat dissipation and steep thermal gradients during welding, which complicates melt pool control and increases process instability [82, 84]. At the same time, magnesium alloys exhibit a narrow processing window, where insufficient heat input results in lack of fusion, while excessive heat input promotes porosity, grain coarsening, and cracking. An appropriate adjustment of welding parameters is essential for stable metal transfer across different welding methods [12, 20]. In the CMT process, parameters governing the boost, burn and SC phase are expected to have a dominant influence on melt pool behaviour and process stability [3, 24]. Furthermore, preheating the base material, commonly used in magnesium welding to mitigate cracking, has also been reported to enhance weld-pool stability [82]. For this reason, key parameter settings, and appropriate preheating were identified as the main process-related factors affecting the stability of the WADED process. Proper control of these factors is expected to stabilise the melt pool, reduce porosity, and enable the fabrication of thin-walled components with a relative material density exceeding 99%.

Q3: How do thermal management strategies affect the development of microstructure and residual stresses during WADED of AZ61 magnesium alloy components?

H3: Fast cooling of the melt pool promotes high nucleation rates, resulting in a fine microstructure [65]. Deposition strategies that provide a more uniform heat distribution without the formation of locally overheated, slowly cooling regions are therefore expected to reduce grain coarsening [50, 52]. On the contrary, preheating the base material and increasing interpass temperatures lead to slower cooling and promote grain growth. Residual stresses are assumed to originate primarily from large thermal gradients [91]. Consequently, lower temperature gradients, achievable through preheating and interpass optimisation, are expected to reduce residual stress formation [31, 92]. It is assumed that increasing preheating and interpass temperatures will reduce residual stresses while simultaneously promoting grain coarsening. Therefore, it is necessary to identify an optimal compromise between microstructural refinement and residual stress reduction.

4.2 Thesis layout

This dissertation is based on four scientific publications that address the scientific questions and hypotheses defined in Chapter 4.1. The individual studies are ordered to reflect the chronological development of the WADED process for magnesium alloys, from basic path planning to the development of custom devices and suitable process settings for magnesium alloy, to the fabrication of a structured component demonstrator.

Research paper I focused on deposition path-planning strategies for thin-walled and overhanging structures fabricated by WADED. The study investigated how the orientation of deposition paths and the placement of weld start and end points affect geometrical accuracy and the formation of residual stresses. The results address scientific question Q1 and hypotheses H1a and H1b.

SLAVÍČEK, J.; FRANKE, J.; JAROŠ, J.; KOUTNÝ, D.; Strategies for Wire Arc Additive Manufacturing of thin walls and overhangs, *Journal of Mechanical Science and Technology*, 2023, vol. 37, no. 1, p. 5529-5534, ISSN: 1738-494X.

Journal impact factor = 1.7, Quartile Q3, CiteScore = 3.0
Author's contribution: 70%



Research paper II addressed the identification of key process-related factors influencing process stability during WADED of AZ61 magnesium alloy using the CMT process. The study investigated the influence of characteristic CMT parameters on weld bead geometry, arc stability, metal transfer behaviour, and resulting material properties. The findings correspond to scientific question Q2 and its associated hypothesis H2.

SLAVÍČEK, J.; ŠVÁB, Č.; GNEIGER, S.; HURNÍK, J.; PROCHÁZKA, P.; KOUTNÝ, D.; Wire-arc directed energy deposition of AZ61 magnesium alloy fabricated by cold metal transfer: Processing, microstructure, and mechanical properties, *Scientific Reports*, 2026, p. 1-22, ISSN 2045-2322.

Journal impact factor = 3.9, Quartile Q1, CiteScore = 6.7
Author's contribution: 53%



Research paper III examined the role of thermal management, specifically base material preheating and interpass temperature control, during WADED of AZ61 thin-walled components. The study analysed their influence on single-track geometry, wall stability, residual stress development, porosity formation, and microstructural evolution. These results partially address scientific questions Q2 and Q3 and corresponding hypotheses H2 and H3.

SLAVÍČEK, J.; ZEMAN, S.; HORYNOVÁ, M.; SENCK, S.; KOUTNÝ, D.; Optimisation of preheating and interpass temperature in WADED magnesium alloy AZ61: A comparative study on microstructure, residual stresses, and internal defects, (PREPRINT – under review)

Author's contribution: 58%



Research paper IV extended the investigation to volumetric AZ61 components and focused on the effect of deposition strategy on thermal history, microstructural evolution, microhardness distribution, and geometrical accuracy. Thermal numerical simulations were employed to predict cooling behaviour and identify critical regions prone to grain coarsening, which were subsequently verified experimentally. This study contributes to answering scientific questions Q1 and Q3 and evaluating hypotheses H1a and H3.

SEIDLER, F.; SLAVÍČEK, J.; KOUTNÝ, D.; Influence of deposition strategy on the microstructure of volumetric WADED-fabricated AZ61 magnesium alloy components, *Additive manufacturing letters*, 2025, vol. 16, p. 1-10, ISSN: 2772-3690

Journal impact factor = 4.7, Quartile Q2, CiteScore = 6.3

Author's contribution: 46%



5 MATERIALS AND METHODS

To test the hypotheses formulated based on the defined scientific questions, a series of experimental investigations and analyses was conducted. An overview of the sample types, applied analyses, and required outputs for each published article is provided in Fig. 5-1. This chapter describes the experimental equipment, materials, processing methods, and characterisation techniques used to develop and validate the additive manufacturing of structured magnesium alloy components by wire arc directed energy deposition.

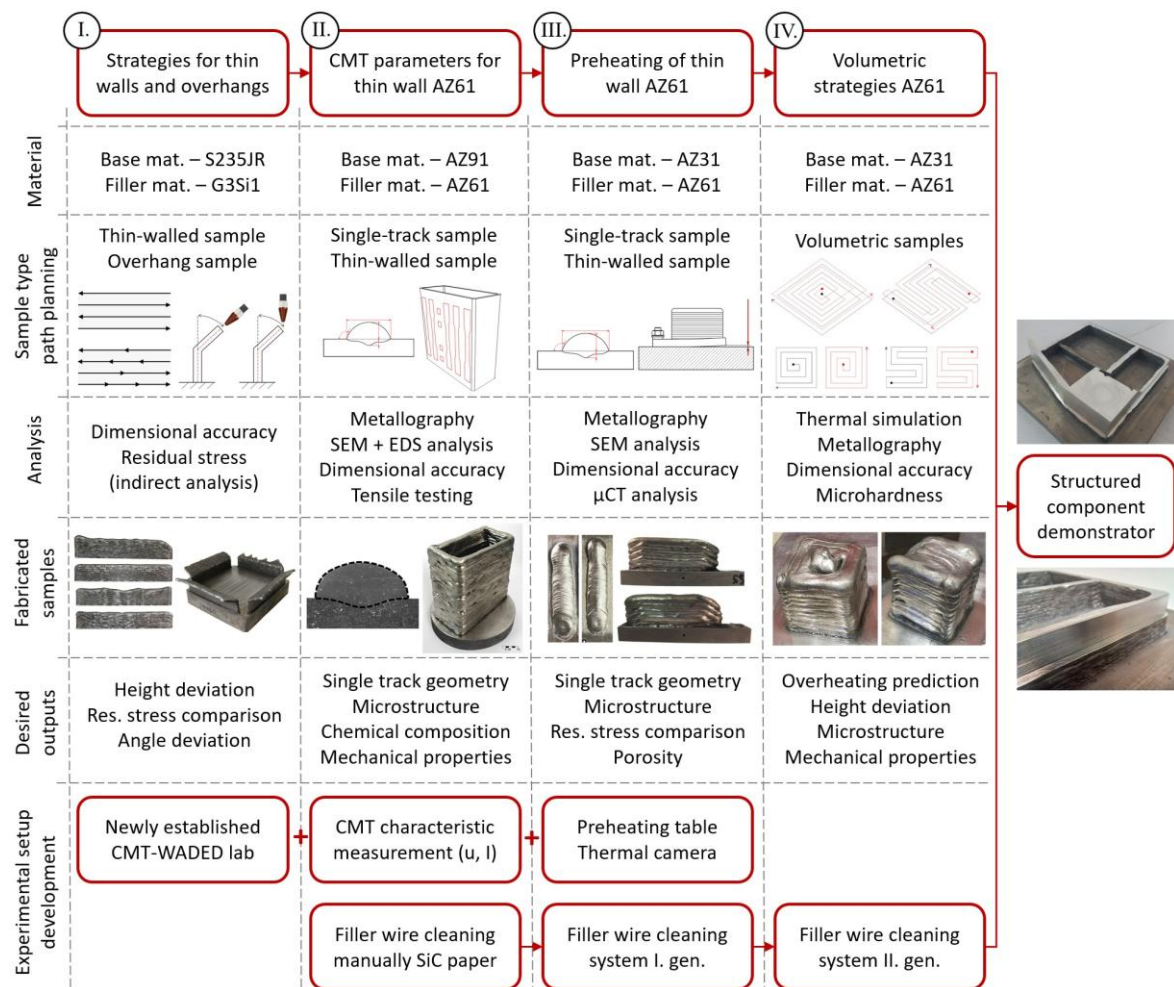


Fig. 5-1 Overview of experimental methods and analyses applied in this dissertation

The experimental workflow followed a chronological sequence corresponding to Research Papers I–IV. Initially, deposition strategies for thin-walled and overhanging structures with a focus on geometrical accuracy and residual stress development were investigated. As these experiments represented the first research conducted at the newly established CMT-based WADED facility, the proposed path-planning strategies were evaluated using the more easily processable G3Si1 steel.

Subsequently, suitable process parameters for WADED of AZ61 magnesium alloy were developed, while key process-related factors influencing process stability and defect formation were identified. Further experiments focused on the production of AZ61 thin-walled structures with preheated base material. The influence of base material preheating and interpass temperature on single-track geometry, thin-wall stability, residual stresses, and internal defects was systematically investigated. The final stage of the experiments addressed the influence of deposition strategies on microstructural evolution in volumetric components. The knowledge gained from all partial studies was applied to the fabrication of a structured component demonstrator.

5.1 Wire arc directed energy deposition

5.1.1 WADED Experimental Setup

As a part of this dissertation, the WADED experimental workplace was gradually developed in parallel with the conducted research activities, as schematically illustrated in Fig. 5-1. The basic workplace configuration consists of a robotic positioning system, a welding power source, a gas exhaust system, and a welding table, as shown in Fig. 5-2.

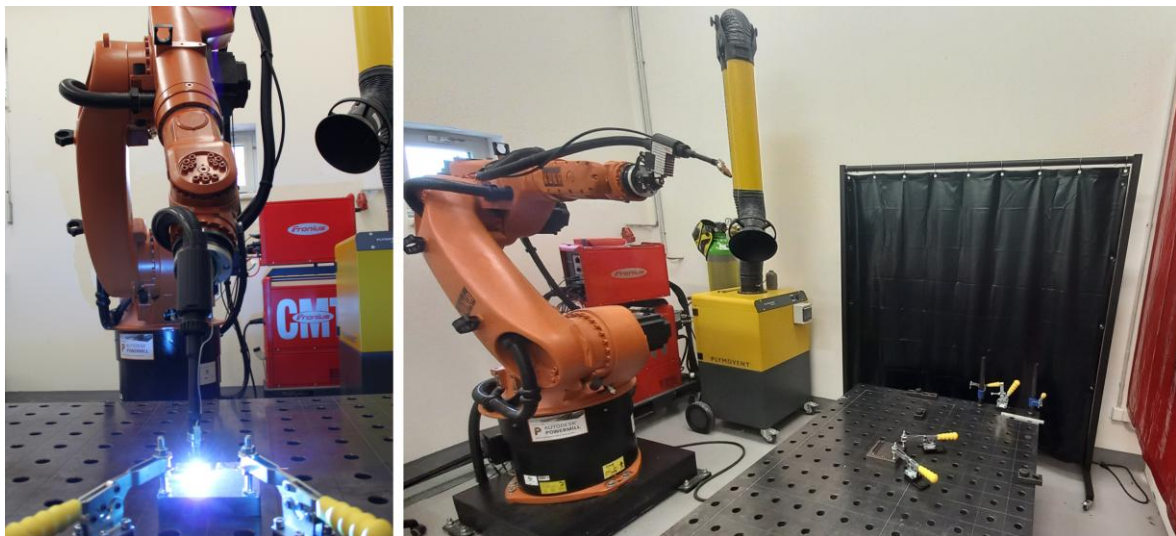


Fig. 5-2 WADED experimental setup at Brno University of Technology

All movements of the welding torch were controlled by a KUKA KR 60 HA robotic arm. The robot is equipped with six axes and offers a payload capacity of 60 kg, a maximum reach of 2033 mm, a rotational speed of 120-322°/s, and a pose repeatability of ± 0.05 mm. The combination of positioning accuracy and dynamic performance makes this robotic system suitable for WADED fabrication of large-scale components. The control code for the robotic arm and the welding power source was generated using a custom slicer developed in the Rhino Grasshopper environment and implemented via the KUKA|prc plugin.

The material deposition was performed using a Fronius TPS 3200 CMT welding power source. The device supports conventional MIG and MAG welding processes with Standard, Pulse and CMT modes. Especially, the CMT mode is suitable for WADED due to its reduced heat input and controlled metal transfer. The power source provides a maximum welding current of 550 A and allows wire feed speeds up to 16 m/min. The system enables adjustment of synergic characteristics, allowing precise control of the essential CMT process parameters. In addition, the power source automatically regulates the shielding gas flow to the welding area.

5.1.2 Process parameters

Welding process parameters were set using the RCU 5000i controller. The Fronius TPS 3200 CMT power source operates according to synergic characteristic lines predefined by the manufacturer for selected filler materials. When working with a synergic welding system, the operator adjusts one of the primary parameters (such as welding current, voltage, or wire feed speed), while the remaining process parameters are automatically adapted according to the selected synergic line.

However, synergic characteristic lines are not available for all filler materials. Therefore, for this dissertation, it was necessary to adjust the parameters at the level of the synergic characteristic itself. In the CMT welding mode, these parameters include the boost phase current – I_{boost} (A), boost phase duration – $t_{I_{\text{boost}}}$ (ms), burn phase current – $I_{\text{sc_wait}}$ (A), the speed of the electrode during the boost and burn phase – $v_{d_{\text{sc_wait}}}$ (m·min⁻¹) and SC phase current – I_{sc2} (A).

These parameters define the characteristic CMT electrical waveform shown in Fig. 2-4. The influence of individual CMT parameters on metal transfer behaviour, process stability, and weld bead geometry is discussed in detail in *Research Paper II*.

5.1.3 Thermal management

Thermal management was identified as a key factor affecting the reliability of WADED processing of magnesium alloys. Due to their specific thermal properties, inadequate control of heat input and temperature distribution resulted in process instabilities and increased defect formation. Consequently, the WADED workplace was upgraded with equipment for base-material preheating and temperature control, enabling stable, reproducible deposition conditions.

A custom preheating table was developed to ensure stable thermal conditions during WADED fabrication. The device was designed to preheat the base material in order to reduce thermal gradients, particularly during the initial stages of the deposition process. The preheating table was developed as part of Zeman's master's thesis [93] and subsequently integrated into the WADED experimental workplace. The overall dimensions of the heated table are 450×550 mm. Eight cartridge heaters provide heating with a total installed power of 3.2 kW, divided into two independently controlled heating zones. The system allows base material preheating to a maximum temperature of $400\text{ }^{\circ}\text{C}$, with accuracy of $\pm 1\text{ }^{\circ}\text{C}$.

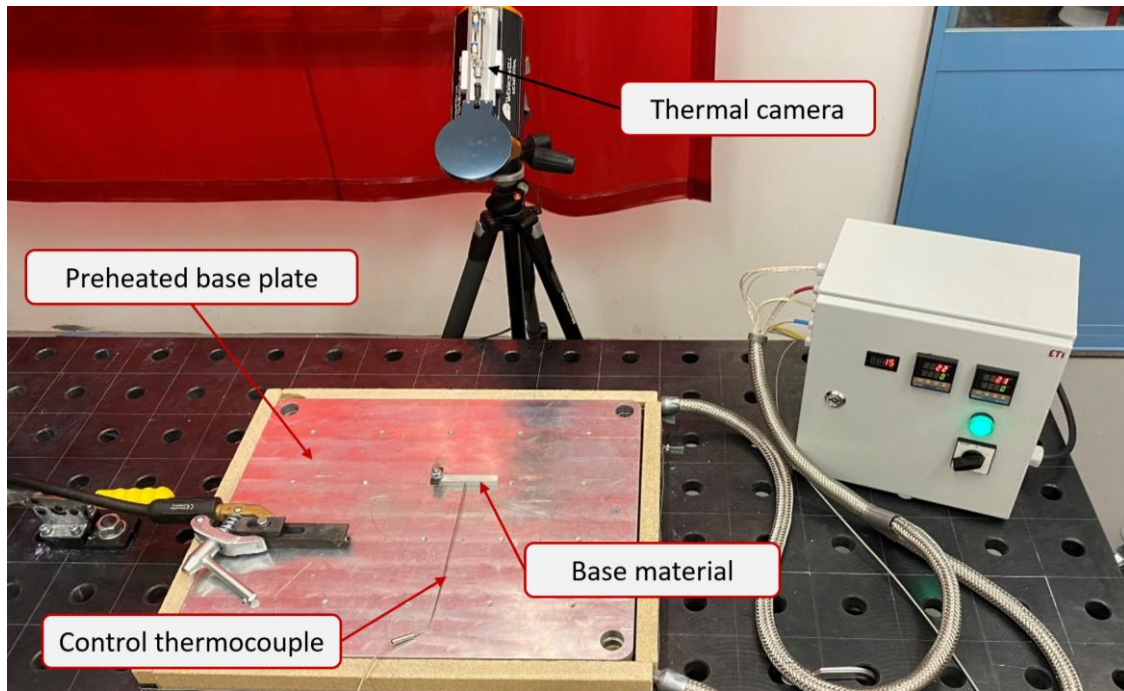


Fig. 5-3 WADED thermal management setup at Brno University of Technology [93]

A Workswell WIC SC industrial thermal camera with a resolution of 336×256 pixels, a temperature measurement range of -40 to $550\text{ }^{\circ}\text{C}$, and an accuracy of $\pm 2\text{ }^{\circ}\text{C}$ was used to monitor and control the preheating and interpass temperatures during deposition. In addition, an Easytherm M2-K1.5D250L thermocouple with a measurement range of -40 to $980\text{ }^{\circ}\text{C}$ and an accuracy of $\pm 2.5\text{ }^{\circ}\text{C}$ was used to control the preheating temperature of the base material and to recalibrate the thermal imaging camera's emissivity settings.

5.1.4 Materials

Two main material types were used in the WADED experiments: steel and magnesium alloys. For *Research Paper I*, a G3Si1 steel welding wire and an S235JR steel base material were used. In *Research Papers II–IV*, as well as for the fabrication of the structured component demonstrator, AZ61 magnesium alloy wire with diameter 1.6 mm was used as the filler material. For base materials, two types of magnesium alloy were used: AZ31 or AZ91, depending on the specific experimental setup.

The exact chemical composition and application of the individual base materials are summarised in Tab. 5.1. The dimensions of the base materials are described in detail in the corresponding *Research papers II–IV*. For the fabrication of the structured component demonstrator, AZ31 base material with a thickness of 20 mm was used.

Tab. 5.1 Chemical composition of filler and base materials

Alloy	Elements	Al	Zn	Mn	Si	Fe	Cu	Ni	Paper
AZ61 (filler wire)	Min (%)	5.50	0.50	0.15					Paper II-IV.
	Max (%)	6.50	1.50	0.40	0.10	0.005	0.05	0.005	
AZ91 (base material)	Min (%)	8.30	0.35						Paper II.
	Max (%)	9.70	1.00	0.13	0.10	0.005	0.03	0.002	
AZ31B (base material)	Min (%)	2.50	0.60	0.20					Paper III-IV.
	Max (%)	3.50	1.40	1.00	0.10	0.005	0.05	0.005	

5.1.5 Filler wire preparation

Magnesium is a highly reactive metal that readily reacts with atmospheric moisture, forming oxide and hydroxide layers on the surface of the filler wire [96]. It was observed that the surface cleanliness of the filler material has a key role in achieving stable WADED processing without significant defect formation. However, the available quality options for commercially supplied magnesium filler wires are limited, and even newly purchased material often shows considerable surface oxidation.

In this dissertation, several different methods for cleaning the welding wire were investigated, as schematically shown in Fig. 5-1. The first method involved manual surface cleaning using SiC abrasive paper. This approach was effective in removing surface impurities. However, it is time-consuming and therefore unsuitable for large-scale production. This cleaning method was applied in *Research Paper I*.

Further cleaning approaches were developed within Seidler's thesis [94] and are illustrated in Fig. 5-4. The filler wire cleaning system I. generation is based on a custom-designed automatic rotary device intended for the mechanical removal of surface contaminants. Following mechanical cleaning, the wire was annealed at 380 °C for 11 minutes to convert residual magnesium hydroxide into magnesium oxide, which is known to have a less harmful effect on the welding process [96]. However, it was observed in *Research Paper III* that this treatment did not ensure complete removal of surface impurities or prevent rapid recontamination of the wire surface before feeding into the welding system, leading to excessive porosity in the deposited material.

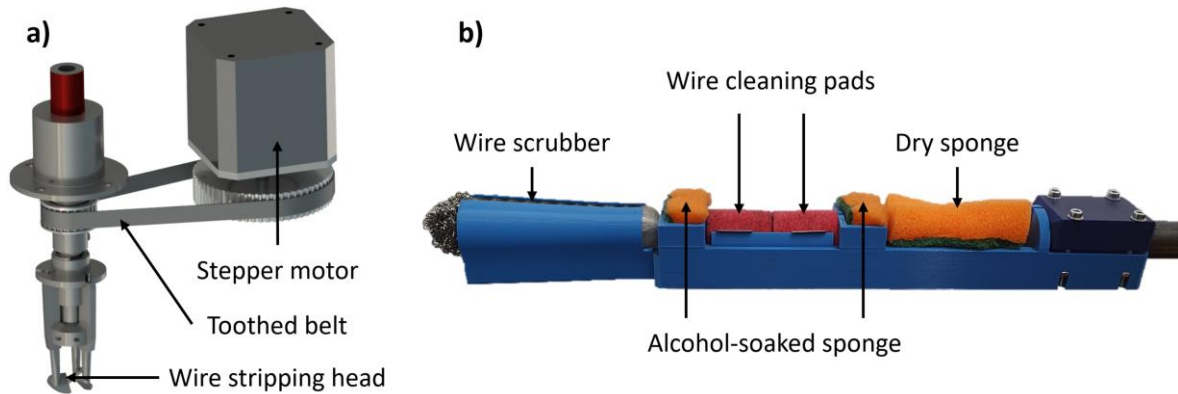


Fig. 5-4 Filler wire cleaning system a) I. generation, b) II. generation [94]

The last tested method was the filler wire cleaning system II. generation, which guides the welding wire through a sequence of cleaning elements immediately before it enters the welding torch. Although this approach is less effective for heavily contaminated filler materials, it proved to be highly efficient for wires with moderate surface contamination. This method was employed in *Research Paper IV* and subsequently used to fabricate the structured component demonstrator.

5.1.6 Samples design and deposition strategies

The design of experimental samples and the selection of deposition strategies were adapted to the objectives of individual research papers. Thin-walled and overhangs were used to investigate deposition path-planning strategies, geometrical accuracy, and buildability (*Research Paper I*). Subsequently, thin-walled AZ61 magnesium alloy samples and single-tracks were employed to study process stability, defect formation, and the influence of CMT parameters (*Research Paper II*), as well as the effects of thermal management on geometrical stability, residual stresses, porosity, and microstructural evolution (*Research Paper III*). Finally, volumetric samples were fabricated to assess influence of deposition strategies on heat distribution, grain size evolution, microhardness, and geometrical accuracy (*Research Paper IV*). The combined knowledge gained from these sample designs was applied to the fabrication of a structured component demonstrator.

5.2 Measurement methods

5.2.1 Material properties

Material properties and microstructural analyses were performed to characterise the material phases, grain size and defects of WADED-fabricated samples across different geometries and processing conditions. Standard metallographic preparation procedures, including sectioning, mounting, grinding, polishing, and etching, were applied to reveal grain boundaries, and phase distribution. Optical light microscopy (OLM) was primarily used to evaluate grain size, layered microstructural features, and geometrical characteristics of deposits and walls, while scanning electron microscopy (SEM) combined with energy-dispersive spectroscopy (EDS) enabled detailed analysis of phase composition and intermetallic particle distribution. These methods were applied throughout *Research Papers II–IV* to assess the influence of deposition strategy, thermal management, and process parameters on microstructural development. The specific equipment and measurement settings varied slightly across individual publications and are therefore not listed here in detail, as they are described in the corresponding *Research papers II–IV*.

5.2.2 Characteristic waveforms measurement

Characteristic electrical waveforms were measured to analyse the behaviour of the cold metal transfer (CMT) process and to assess the influence of individual process parameters on arc stability and metal transfer. Electrical signals were recorded using a custom experimental setup consisting of a Tektronix DPO 2024B oscilloscope with a sampling frequency of 312.5 kHz and voltage and current sensors. This measurement approach was applied in *Research Paper II*.

5.2.3 Mechanical properties characterization

Tensile testing

Tensile testing was performed in *Research Paper II* to evaluate the mechanical properties of WADED-fabricated thin-walled components. Tensile specimens were machined from the deposited walls according to DIN 50125 (type E) and extracted in both the welding and transverse directions to assess mechanical anisotropy. The tensile tests were conducted using a Shimadzu AGX-V 100 kN testing machine at a deformation rate of $0.5 \text{ mm} \cdot \text{min}^{-1}$. Ductility was measured using a stereo 3D digital image correlation (DIC) system, Dantec Dynamics, equipped with two 5 MPx cameras ($2448 \times 2048 \text{ px}$) and 50 mm lenses.

Microhardness testing

Microhardness measurements were performed in the same regions as the grain-size analysis in *Research Paper IV* using a Qness 60 A+ microhardness tester with a 0.1 kgf load (HV0.1). The volumetric WADED-fabricated samples were not sufficiently large to prepare standard tensile specimens. Therefore, microhardness testing was used to assess local variations in mechanical behaviour associated with different deposition strategies and microstructure.

5.2.4 Geometrical accuracy and defect analysis

Dimensional accuracy

The geometric accuracy of the manufactured samples in *Research Papers I, II and IV* was evaluated using an ATOS Triple Scan III optical 3D scanner. The acquired scan data were processed in the GOM Inspect 2018 software to assess deviations from the desired geometry.

Indirect residual stress assessment

In addition to geometric evaluation, optical 3D scanning was also used for the indirect assessment of residual stress. These measurements were performed in *Research Papers I and III*. In both cases, the sample design was inspired by the principles of the Bridge Curvature Method commonly used in LPBF, which enables rapid evaluation of the influence of process parameters on residual stress development. The sample designs for indirect residual stress measurement are shown in Fig. 5-5 Indirect residual stress assessment: a)

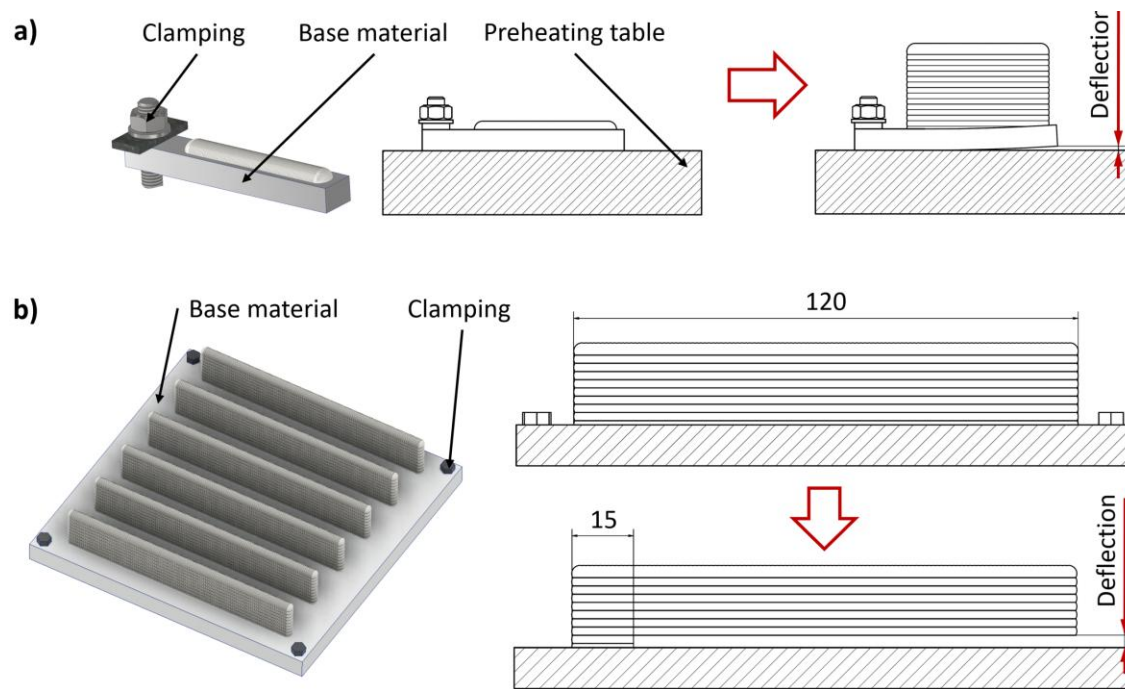


Fig. 5-5 Indirect residual stress assessment: a) Research paper III, b) Research paper I

Research paper III, b) Research paper I.

In *Research Paper I*, a base material with a thickness of 30 mm was rigidly clamped to the welding table on both sides to suppress deformation during deposition. Thin walls were fabricated using different deposition strategies, allowed to cool, and subsequently scanned. After first scanning, the walls were cut on one side, leaving only a 15 mm section attached to the base material from the original wall length of 120 mm. The samples were then scanned again, and the resulting deflections were compared to evaluate the influence of deposition path-planning strategies on residual stress formation.

A different sample design was applied in *Research Paper III*. In this case, the base material thickness was intentionally reduced to 10 mm and clamped on only one side, allowing free deformation during cooling. Initial scanning was performed directly on the welding table prior to thin-wall fabrication using a handheld optical scanner, Scantech SIMSCAN 22. After deposition, the sample was allowed to cool to room temperature and scanned again. Both scans were aligned and analysed in GOM Inspect 2018 software to quantify deflection in the direction normal to the preheating table surface. Using this approach, it was possible to compare the influence of different preheating conditions on residual stress development in thin-walled structures.

Micro-computed tomography analysis

In addition to deflection-based residual stress assessment, the influence of preheating on internal defect formation was investigated in *Research Paper III* using micro-computed tomography (μ CT). A Procon CT-Alpha μ CT system was employed to analyse the relative material density of the deposited thin-walled samples. The specimens were scanned using a cone-beam μ CT device with an isotropic voxel resolution of 45 μ m. The acquired μ CT data were evaluated in VGStudioMax 2023.1, where pore distribution, morphology, and size were quantified using an ISO50 threshold and a minimum pore size of 8 voxels.

5.2.5 Thermal simulations

Thermal numerical simulation was employed in *Research Paper IV* to analyse heat distribution and cooling behaviour during WADED fabrication of volumetric AZ61 magnesium alloy components. The simulations were used to identify regions with reduced cooling rates, which are susceptible to grain coarsening. The predicted thermal behaviour served as a basis for selecting locations for subsequent microstructural and microhardness analyses, which were used to compare different deposition strategies. Thermal modelling was carried out using Ansys 2025 R1 with the DED Process add-on to simulate the temperature evolution during deposition. The model accounted for transient heat transfer, including temperature-dependent thermal conductivity and specific heat capacity. The exact simulation setup is described in detail in *Research Paper IV*. In addition, a validation experiment focused on monitoring the cooling behaviour of a single deposit was performed to verify the input parameters used in the simulation.

6 RESULTS AND DISCUSSION

The following section presents the dissertation's key results, focusing on additive manufacturing of structured magnesium alloy components using WADED. The first part of this chapter is organised into four subsections corresponding to the individual research papers, each addressing one or more of the above-mentioned scientific questions. The second part of this chapter contains the full versions of the research papers, in which the results are presented in detail.

6.1 Research paper I

The key results of *Research Paper I* concern deposition path-planning strategies for thin-walled and overhanging structures fabricated by wire arc directed energy deposition. As this study represents the first experimental investigation conducted at the newly established WADED workplace, the experiments were performed using the easier-to-process G3Si1 steel filler material rather than the magnesium alloy, which is the main subject of this dissertation. The findings primarily address geometrical and process-related aspects of deposition path planning and are therefore, to a certain extent, applicable to other filler materials. These results were subsequently applied and further examined in the following research papers focused on magnesium alloy processing.

The experimental results demonstrated that deposition path planning significantly influences the geometrical accuracy of thin-walled components fabricated by WADED. Although all investigated deposition strategies enabled the formation of continuous walls, notable differences in wall height and material distribution were observed. Excessive material deposition occurred at the weld bead starting points, while insufficient material deposition was observed at the weld bead ends. Similar behaviour of weld beads has been previously reported by Zhang et al. [36]. This phenomenon is attributed to rapid heat dissipation at the starting point of the weld bead, where the penetration depth is typically shallow. As a result, the weld pool solidifies with a greater layer height than in the central region of the weld bead, where deeper penetration occurs. When the weld bead start points and end points were repeatedly located at the same positions along the wall length, these local geometric deviations accumulated, leading to pronounced asymmetry of the fabricated walls.

Deposition strategies that alternated the welding direction between successive layers or redistributed weld bead start points and end points along the wall length resulted in a more uniform material distribution and improved geometrical accuracy of the thin-walled components.

The influence of deposition path planning on residual stress formation was also investigated. In conventional welding practice, welding sequences are commonly used for long welds to improve heat distribution and reduce residual stresses [39, 40]. However, under the investigated conditions, no clear systematic dependence between the applied deposition strategies and the magnitude of residual stresses was identified. A likely explanation is that the use of the cold metal transfer welding mode inherently reduces heat input and residual stresses to such an extent that additional sequencing of weld beads does not lead to further stress reduction for thin walls 120 mm long. Moreover, it was observed that an increased number of weld bead starts and ends can even contribute to higher residual stresses if the deposition strategy is not appropriately designed. For substantially longer wall lengths, deposition sequencing may still positively influence residual stress formation. However, within the investigated wall length, no such effect was observed.

Although sequential welding did not significantly affect residual stress levels in thin-walled structures, it did improve the manufacturability of inclined walls. Overhanging structures with inclination angles ranging from 30° to 90° were successfully fabricated without support structures. Among the investigated approaches, sequential welding deposition strategies achieved higher geometrical accuracy and surface quality. In contrast, continuous deposition strategies led to overheating, melt pool collapse, greater angular deviations, and surface irregularities, particularly at higher overhang angles.

The results further showed that increasing the overhang inclination inherently reduces geometrical accuracy, humping effect occurs (same as observed in [44]), even when optimised deposition strategies are applied. This behaviour was attributed to gravity-induced weld pool flow and limited heat dissipation, which became more pronounced with increasing inclination. Nevertheless, the study demonstrated that suitable deposition path planning significantly extends the geometric limits of inclined walls fabricated by WADED and enables their production without changing the component's orientation, which is particularly important for large-scale applications.

6.2 Research paper II

The key findings of *Research Paper II* relate to wire arc directed energy deposition of AZ61 magnesium alloy using the cold metal transfer (CMT) welding process. The study focuses on process stability, deposition geometry, and the resulting material properties. In particular, the study identifies key process-related factors in magnesium alloy processing and presents suitable CMT parameters for reliable WADED fabrication of thin walls.

Prior to fabricating samples for this study, a series of preliminary experiments was conducted to assess the feasibility of WADED processing of AZ61 magnesium alloy. These experiments revealed that the surface condition of the filler wire has a critical role in process stability. Magnesium alloys tend to form a surface layer of oxides and hydroxides when exposed to atmospheric air [95, 96]. When untreated welding wire was used, the welding arc became unstable, in several cases leading to a complete interruption of the deposition process. To mitigate this effect, the filler wire was manually cleaned using SiC paper prior to deposition. This procedure resulted in more stable arc behaviour and reliable material transfer, identifying the surface quality of the filler material as a key process-related factor influencing process stability during WADED of AZ61 magnesium alloy.

The parametric study further demonstrated that the characteristic parameters of the cold metal transfer process strongly influence deposition geometry and process stability during WADED of AZ61 magnesium alloy. The boost-phase current and boost-phase duration were identified as dominant parameters governing weld bead geometry, heat input, and metal transfer behaviour. Inappropriate settings of these parameters can lead to excessive droplet growth [97], unintended arc behaviour (such as spray transfer [24]), and irregular deposition geometry. With respect to weld bead geometry, increasing the boost current and its duration resulted in an increased bead width, penetration depth, and total deposited area. In contrast, a balanced adjustment of the boost-related parameters enabled stable metal transfer and uniform bead formation.

The electrode speed during the boost and burn phases was found to affect weld pool behaviour and penetration depth significantly. Similar trends have been reported in the research paper focused on CMT-based welding of aluminium alloys [24]. Excessively low or high electrode speeds resulted in shallow penetration, unstable deposition conditions and increased sensitivity to process interruptions. The short-circuit phase current primarily influenced penetration depth, while its effect on bead width and height was less pronounced. These results indicate that stable WADED processing of AZ61 magnesium alloy cannot be achieved by tuning individual parameters independently, but requires a carefully balanced combination of CMT characteristic parameters.

Under stable CMT processing conditions, defect formation was mainly observed as isolated porosity, including globular pores and local shrinkage cavities. The occurrence of these defects was not directly associated with arc instability, but was attributed primarily to the surface condition and quality of the filler wire. These results indicate that, once stable arc behaviour is achieved through appropriate parameter selection, defect formation during WADED of AZ61 magnesium alloy is influenced primarily by material-related factors.

Using the optimised set of parameters, a thin-walled AZ61 magnesium alloy component was successfully fabricated using the CMT-based WADED process. The produced structure exhibited good geometrical accuracy, demonstrating the feasibility of this approach for thin-walled magnesium alloy components. Microstructural analysis revealed a characteristic layered structure with alternating fusion and intralayer regions, similar to that observed by Klein in Al-based materials [98], reflecting cyclic thermal loading during deposition. The microstructure was composed predominantly of an α -Mg matrix with dispersed intermetallic particles and a limited amount of β -phase ($\text{Mg}_{17}\text{Al}_{12}$). Mechanical testing indicated anisotropic behaviour. Possible causes of anisotropy include the alternation of the fusion zone and interlayer zone associated with applying material layer by layer, as described by Kumar [78] or Guo [76]. The mechanical properties reported in this paper are higher than those for castings, lower than forgings, and approximately comparable to those produced by laser powder bed fusion.

6.3 Research paper III

The key results of *Research Paper III* relate to thermal management during CMT-based WADED processing of AZ61 magnesium alloy. The study primarily examines the influence of base material preheating and interpass temperature control on microstructural evolution, residual stress development, and internal defect formation. Thematically, this work builds upon previous research focused on CMT process parameters and the identification of key process-related factors. Previous studies [60, 82] have shown that thermal management has a crucial role in arc-based additive manufacturing processes, motivating the in-depth investigation presented in this paper. To enable controlled thermal conditions, a custom preheating table and filler wire cleaning system were developed and integrated into the WADED system, as described in the Materials and Methods section.

First, the effect of base material preheating on single-track geometry was investigated. Two sets of CMT parameters were evaluated: one developed in a previous study and a second, higher-energy parameter set designed to increase penetration into the AZ31 base material. The results showed that increasing the preheating temperature led to a wider weld bead and deeper penetration. At low or zero preheating temperatures, insufficient penetration and unstable bead geometry were observed. At a preheating temperature of 250 °C, both parameter sets resulted in continuous penetration across the entire interface between the deposited and base material.

Subsequently, the effect of base material preheating and interpass temperature on the geometrical stability of thin-walled AZ61 structures was analysed. The results revealed that thermal conditions strongly affect the evolution of layer width, particularly in the initial layers. A pronounced narrowing of the deposited wall in the first layers, commonly referred to as necking, has been reported in WADED for various materials [32, 99] and was observed in this study as well. Without sufficient preheating, significant necking occurred in the lower layers, resulting in reduced geometric stability. In contrast, the application of controlled preheating combined with an elevated interpass temperature resulted in a more uniform layer width along the wall height. The evaluation of this experiment showed that higher-energy parameters were unsuitable for producing thin walls. Therefore, only parameters from the previous study were used for further experiments.

Residual stress development was evaluated indirectly by measuring base material deflection under different thermal conditions. The results demonstrated a clear dependence of deflection magnitude on the applied preheating temperature. Increasing the preheating temperature led to a significant reduction in base material deflection, indicating lower residual stress levels. This effect was attributed to reduced thermal gradients during deposition, with a deflection reduction of up to 50% observed at a preheating and interpass temperature of 300 °C.

Internal defects in thin-walled samples were analysed using micro-computed tomography. The results showed that both the amount and morphology of porosity strongly depend on thermal conditions. At low preheating temperatures, columnar pores oriented along the build direction were observed. Increasing the preheating temperature led to a substantial reduction in total porosity and suppression of columnar pore formation, resulting predominantly in small, isolated globular pores. The mechanisms governing porosity formation and its reduction by preheating are discussed in detail in *Research Paper III*. In brief, hydroxides present on the surface of the filler wire decompose upon entering the molten pool, releasing hydrogen gas [96]. Higher preheating and interpass temperatures raise the component's overall thermal state, slowing melt-pool cooling and providing sufficient time for dissolved gas to escape before solidification.

Microstructural analysis revealed that thermal management significantly affects grain size and phase distribution within WADED-fabricated AZ61 thin walls. The deposited structures exhibited a characteristic layered microstructure consisting of fusion and intralayer regions. Increasing preheating and interpass temperatures promoted grain coarsening and a more pronounced presence of the β -Mg₁₇Al₁₂ phase, particularly in regions subjected to prolonged thermal exposure.

Evaluation of the above-mentioned thermal management effects identified an optimal preheating and interpass temperature range of 200–250 °C. This temperature window represents a compromise among microstructure, residual stress reduction, and minimising internal defects and provides suitable processing conditions for the WADED of AZ61 alloy.

6.4 Research paper IV

The key results of *Research Paper IV* concern the fabrication of volumetric AZ61 magnesium alloy components using the CMT-based WADED. The primary focus of the work is the influence of deposition strategy on microstructural evolution and geometrical accuracy. Unlike the previous studies, which primarily addressed thin-walled structures, this work extends the investigation to volumetric components, which are often part of large-scale structured components. The study builds on earlier findings on deposition path planning and thermal management and examines their combined effect in volumetric WADED-fabricated geometries.

Three different deposition strategies were proposed and investigated using numerical simulation and subsequently verified by experimental fabrication to analyse heat distribution and microstructural evolution in volumetric WADED-fabricated AZ61 components. Numerical simulations were employed to evaluate the thermal behaviour during deposition and to identify critical regions with reduced cooling rates that are susceptible to microstructural coarsening. Based on these predictions, selected areas were defined for detailed microstructural and microhardness analysis. The simulations revealed that cooling rates decrease with increasing material accumulation and are locally reduced near the outer edges and corners of the deposition trajectories, where heat dissipation is limited. Similar trends have been reported by other authors [100–102] investigating thermal behaviour in arc-based additive manufacturing processes, and were subsequently confirmed in this study through experimental validation.

Optical grain-size measurements confirmed reduced cooling rates in critical regions predominantly located at trajectory corners and along the outer edges of the deposited volume. Among the investigated deposition strategies, the S-pattern exhibited the narrowest grain-size range (26.32–44.71 μm) and the smallest maximum grain size, indicating a more uniform thermal distribution and effective suppression of grain coarsening, consistent with observations reported for aluminium alloy by Sun et al. [52]. In contrast, the Spiral strategy produced the coarsest grains, reaching up to 60.27 μm in the predicted critical regions. The ZigZag strategy showed intermediate behaviour, characterised by a wider grain-size range. These results confirm that the deposition strategy significantly influences the microstructural evolution in volumetric WADED-fabricated AZ61 components.

Microhardness measurements were used to complement microstructural observations and to evaluate variations in mechanical response across different deposition strategies. The results revealed spatial variations in hardness that corresponded well with the observed microstructural heterogeneity. These findings indicate a clear relationship between local cooling rate, grain size, and microhardness.

The deposition strategy also significantly influenced the geometric accuracy of the fabricated volumetric components. Three-dimensional surface scanning revealed that the ZigZag deposition strategy achieved the smallest range of height deviations and the highest overall geometrical precision. In contrast, the Spiral strategy exhibited the largest deviations, including excessive material accumulation at layer start points and the formation of a critical defect at the centre of the component, which may compromise process reliability. The S-pattern strategy provided acceptable geometrical accuracy, characterised by a narrower deviation range than the Spiral strategy and the absence of major surface defects.

Overall, the results demonstrate that deposition strategy represents an important design parameter in WADED of volumetric AZ61 components, influencing both microstructural uniformity and geometrical accuracy.

6.5 Structured component demonstrator

As a final step of this dissertation, a structured component demonstrator was fabricated by combining the knowledge gained from the research presented in *Research Papers I–IV*. The demonstrator geometry was selected to highlight the potential of wire arc additive manufacturing to reduce the buy-to-fly ratio in the production of structured AZ61 magnesium alloy components. The selected design was inspired by an aircraft frame component introduced earlier in the Introduction and incorporates both volumetric and thin-walled sections, as shown in Fig. 6-1a.

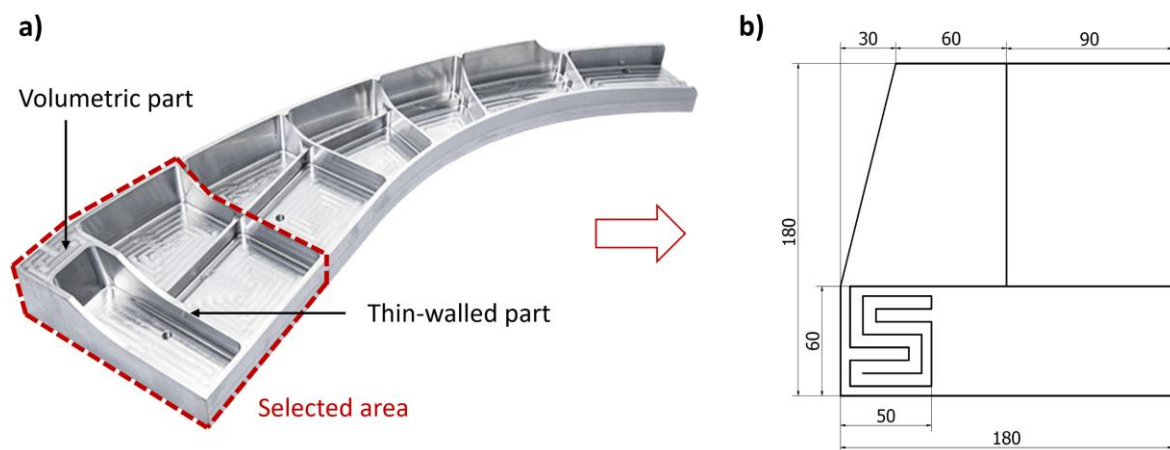


Fig. 6-1 Aircraft C-shape part: a) selected area [9] (edited), b) deposition path sketch

The manufactured demonstrator component had overall dimensions of $180 \times 180 \times 50$ mm and consisted of 20 deposited layers. The geometry was produced according to the deposition path scheme shown in Fig. 6-1b. During fabrication, differences in layer height were observed between the volumetric and thin-walled regions. These differences were compensated for by increasing the number of thin-walled perimeter paths. After deposition, the demonstrator was cleaned, documented, and subsequently machined to its final shape, as shown in Fig. 6-2.

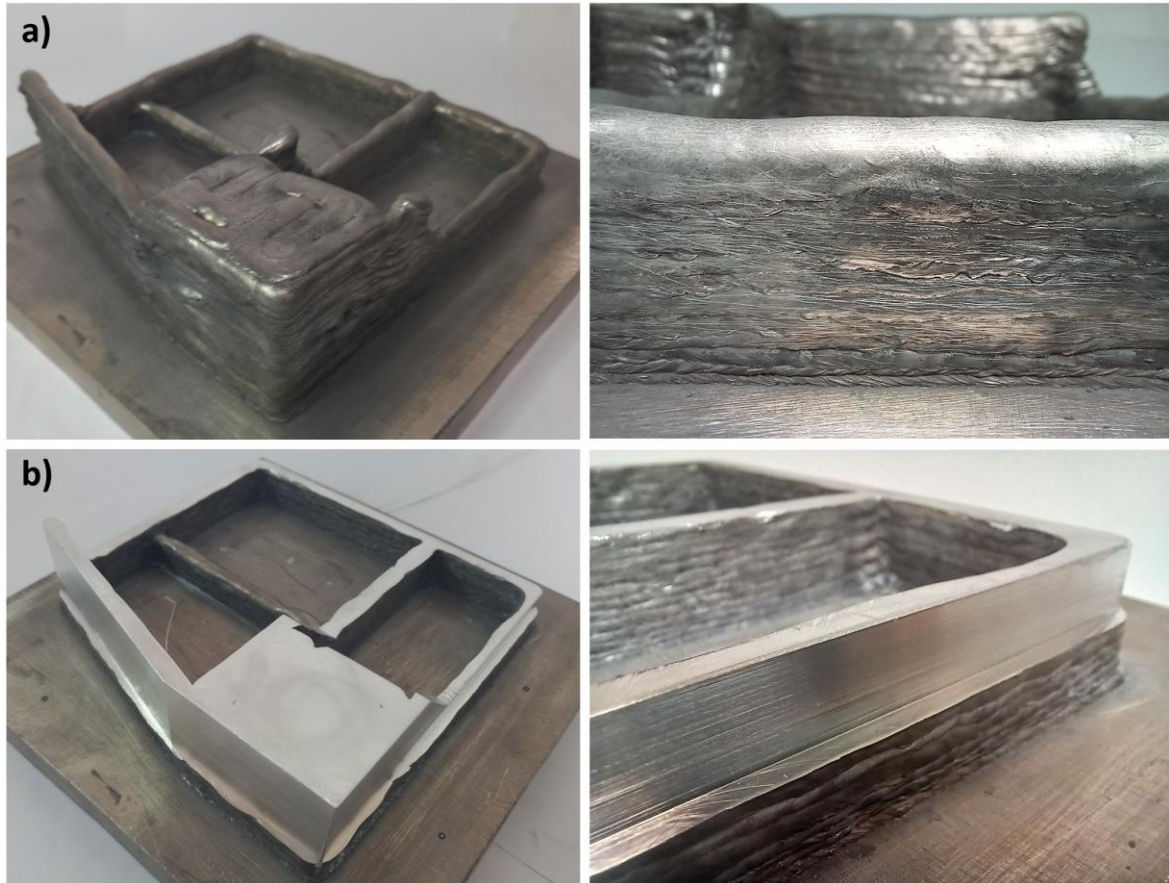


Fig. 6-2 Structured component demonstrator: a) as-fabricated, b) partially machined

The fabrication of this structured component demonstrator successfully integrated the findings from the preceding research papers and applied them to a representative component. The final machined part exhibited no columnar porosity or other unacceptable internal defects, confirming the practical applicability and robustness of the developed WADED strategies for the production of structured magnesium alloy components.

6.6 Original research papers

Original Article

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· Overhangs
· Thin walls
· WAAM

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Strategies for wire arc additive manufacturing of thin walls and overhangs

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Abstract In recent decades, additive manufacturing has become a widely discussed topic. Wire arc additive manufacturing (WAAM) is an example of additive manufacturing based on direct energy deposition. The core principle of this technology is depositing filler material layer by layer using electric arc welding. Most additive technologies face manufacturing limits in the form of overhangs and residual stresses, and WAAM is no exception. This paper deals with wire arc additive manufacturing and focuses on production strategies for thin walls and overhangs. The article describes the influence of production strategies on residual stresses and production accuracy. Five different strategies for thin wall production were proposed and verified, and overhangs were produced. Results show that the position of the beginnings and ends of the weld bead is crucial for production accuracy and residual stresses. Suitable strategies to produce overhangs fabricated with low angular deviation and sufficient surface quality were identified.

1. Introduction

Additive manufacturing is an alternative method for producing near-net-shape parts. It has become a much-discussed topic in the world of manufacturing and engineering over the past few decades [1]. According to the ASTM standard, there are three groups focused on additive manufacturing processes for metals: powder bed fusion, sheet metal lamination and direct energy deposition [2]. The direct energy deposition method can be further subdivided according to the type of energy source (plasma, electron beam, laser and arc) [2].

Wire arc additive manufacturing (WAAM) is an example of a direct energy deposition method. In this process, a filler material is deposited layer by layer by electric arc welding (GMAW or GTAW) [1]. WAAM technology has many advantages, such as low equipment costs, high deposition rates, good structural integrity and low welding wire costs [3]. However, arc welding production can be challenging as several process parameters need to be monitored. These parameters include wire feed speed (WFS), travel speed (TS), voltage, current, contact tip-to-work distance (CTWD), torch angle and shielding gas flow rate. They are crucial for the quality of the finished parts and usually need to be optimised for each material and equipment facility [1].

The material deposition process in WAAM technology is very similar to the classical welding process. Therefore, the occurring defects in wire arc additive manufacturing are similar and include porosity, delamination, hot and cold cracks and spatter. These defects can be caused by poor feedstock quality, gas contamination, poor path planning and excessive heat input, which implies the formation of more residual stresses [1]. However, the deposition of weld beads next to and on top of each other leads to new macroscopic defects. These defects are typical for WAAM and usually caused by the summation of partial defects and include the collapse of start and end points [4], unmelted wire [5] and deformations after the clamp is released [6].

Residual stresses and deformations have accompanied welding technology since its beginnings. The formation of residual stresses and deformations is related to the heat input during

welding. During welding, liquid metal is deposited at a high temperature, and when this material cools, it causes shrinkage and deformation [6]. These residual stresses are highest in the longitudinal direction of the deposit and can lead to fatal production defects. There are several ways to reduce residual stresses in welded parts, such as cold-work-based techniques (CMT), heat treatments, mechanical treatments (rolling, shot peening), reduction of thermal gradients (preheating) and optimisation of path planning strategies.

In addition to residual stresses and process parameters, the production of complex parts can also be negatively affected by the problematical geometry of the manufactured part. WAAM technology, like most additive technologies, faces production limits such as overhanging surfaces, sharp corners, thin walls and difficult-to-process materials. Currently, several studies in the field of WAAM deal with path planning strategies [7], thin walls and overhangs [8] manufacturing. But so far, no publication connects the topic of path planning with residual stresses. This paper deals with WAAM strategies to produce thin walls and overhang fabrication and describes the influence of production strategies on residual stresses and production accuracy.

2. Materials and methods

The production of components with WAAM technology is a specific task, especially in the case of thin walls and overhangs. Therefore, five different printing strategies were tested to reduce manufacturing inaccuracy and residual stress and increase the surface quality of the manufactured components.

2.1 Wire arc additive manufacturing

The robotic arm (KUKA KR60HA, Augsburg, Germany) was equipped with a welding power source (Fronius TPS 3200 CMT, Wels, Austria). A G3Si1 filler wire with a diameter of 0.8 mm and an S235JR steel base plate were used. Production took place in the inert atmosphere of CORGON (82 % Ar, 18 % CO₂) at a flow rate of 15 l/min. Table 1 shows the chemical composition of the G3Si1 filler wire.

Based on the previous experiments, the WAAM process parameters were used, i.e. a wire feed rate (WFR) of 5 m/min and a torch speed (TS) of 0.005 m/min. These parameters were used in all the experiments mentioned below. These process parameters were selected based on several criteria. These criteria include good continuity of the weld bead, a suitable contact angle and an inconspicuous start and end point of the weld bead.

Table 1. G3Si1 composition (declared by KOWAX - Svarmetal s.r.o.).

Fe	Mn	Si	C
Balance	1.53	0.90	0.09

2.2 Thin-wall production strategy

The production of thin walls using WAAM technology is specific because the small volume of the material cannot effectively dissipate the thermal energy when a layer is formed on top of the previous layer. An optimal production strategy can improve heat dissipation and reduce production time.

The five production strategies with different arrangements of torch paths were used (Fig. 1) to observe the formation of a thin wall. The thin-walled samples consist of 20 layers of material with a length of 120 mm. The deformations of the thin wall were observed with a 3D optical scanner.

Based on previous experiments focused on joining weld beads, offsets in the longitudinal direction of the weld bead were used. They were used to achieve a continuous material deposition in the entire length of a thin wall. These offsets ensure compensation of excess or insufficient material deposition at the weld bead joining points. There are two types of joints in these production strategies. The first type is the case where the weld beads are welded one behind the other. This case uses an offset of minus one millimetre (theoretical gap). The second type of joint is the case, where the weld beads are welded against each other. This case works with an offset of plus four millimetres (theoretical overlap).

2.3 Overhang strategy

The combination of the robotic arm with a welding power source allows the production of complex components such as inclined parts. Therefore, the inclined thin walls were fabricated to determine the optimal strategy for overhangs (Fig. 2(c)).

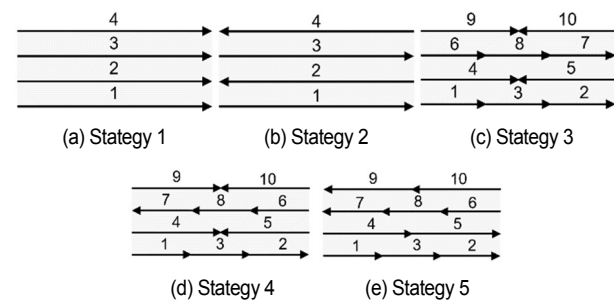


Fig. 1. Thin wall production strategies.

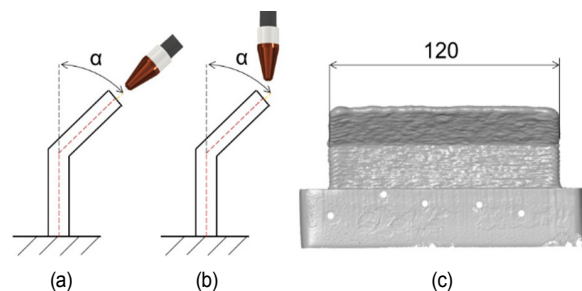


Fig. 2. Inclined thin walls: (a), (b) schemes; (c) produced samples.

Based on the previous experiment, the three production strategies were used to produce inclined thin walls. Strategies 2, 4 and 5 (Fig. 1) were used with the same orientation of the welding torch as the overhang walls. In addition, strategy 2, with the vertical orientation of the welding torch, was used to compare the influence of the welding torch orientation. Four inclinations of the thin walls were selected, namely 30°, 45°, 60° and 90° (Fig. 2(a)). The evaluation criteria for choosing the best strategy for overhang production include the angle deviation and the quality of the surface.

2.4 3D optical scanning

An optical 3D scanner (Atos Triple Scan III, GOM GmbH, Braunschweig, Germany) was used to analyse the geometry of the manufactured samples. The measuring point distance was 0.055 mm. The scanned data were processed in the GOM Inspect software.

In the case of thin walls, the magnitude of residual stress was evaluated based on deformation after a partial cut of the sample. The samples were scanned after production to obtain the production shape. Then the thin walls were cut by a 105 mm horizontal cut near the base plate (Fig. 5). The specimens were scanned again to obtain the deformed shape. The amount of residual stress was evaluated based on the deformation of the samples (Fig. 5). This method compared the influence of the strategies on the residual stresses and did not obtain exact stress values.

In the case of inclined walls, the angle deviation and the surface quality need to be analysed. Resulting analyses help to compare the influence of the welding torch orientation and the strategies on the quality of the specimens. The angle was measured between two auxiliary lines fitted through the scanned surface (Fig. 3(b)). Each sample was measured in three cross-sections (Fig. 3(a)). The standard deviation (σ) was used for the surface quality evaluation. It was calculated as the profile deviation from the auxiliary fitted line (see Fig. 3(b), green - fitted line, black - surface profile). A lower standard deviation indicates less waviness which means better surface quality. The auxiliary lines were created using the Gaussian best-fit method (based on the least squares method) with the three-sigma rule.

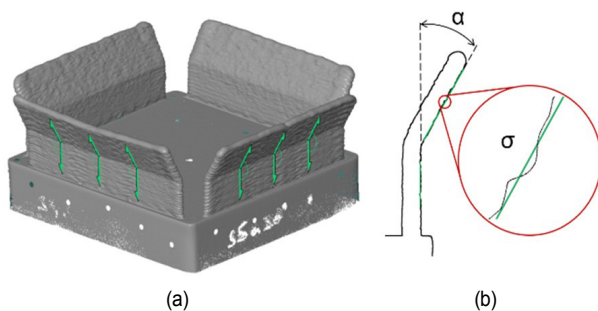


Fig. 3. Down-skin measurement: (a) measurement positions; (b) detail.

3. Results

Two types of strategy-focused experiments were conducted. The first focused on the formation of residual stresses during the fabrication of thin-walled specimens. The second experiment dealt with the influence of the strategies on the production of overhangs.

3.1 Thin walls production

The five production strategies were tested to obtain an optimal geometry of the thin walls without imperfections. Fig. 4 shows the results. All strategies led to the formation of solid walls, but the beginnings and the ends of torch paths influenced the geometry of the thin walls.

3.1.1 Geometry accuracy of thin-walled samples

Thin walls were made by layering weld beads on top of each other. During this process, local defects from the geometry of the individual weld beads were added up. As a result, irregularities occurred at the beginnings and ends of the weld beads where the welding conditions were unstable. An excessive material amount was deposited at the beginning of the weld bead. Conversely, an insufficient amount of material was deposited at the ends of the weld. The appropriate material amount was deposited only in the stable weld bead zone.

Strategy 1 is an example of the accumulation of local defects in the production of thin-walled specimens. This strategy resulted in depositing an excessive amount of material on one side of the sample and an insufficient amount on the other side. Opposite directions of welding paths in every second layer compensated for the deviation at the beginnings and ends of the weld beads. Therefore, production strategy 2 resulted in a solid thin wall with a straight top surface. Strategy 3 contained more weld bead starts on one side of the sample, which again resulted in an area of excess material deposition. This excess material can be seen on the left side of the specimen in Fig. 4(c). Strategies 4 and 5 consist of several partial weld beads welded in a specific order using the offsets described in Subsec. 2.2. This division of the weld beads should provide more uniform cooling of the specimen and, therefore, less re-

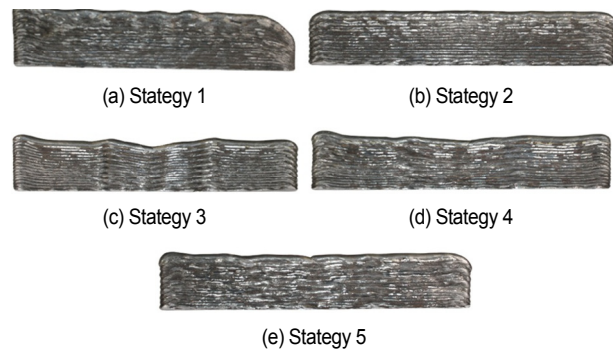


Fig. 4. Fabricated thin-walled samples.

Table 2. Deformation after partial cutting of samples.

Strategy 1	Strategy 2	Strategy 3	Strategy 4	Strategy 5
0.35	0.67	0.91	0.56	0.79

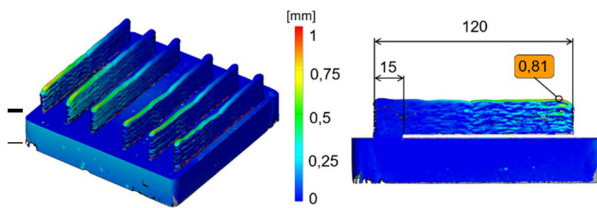


Fig. 5. Deformation after partial cutting of samples.

sidual stress. The number of beginnings and ends of the weld beads is the same on both sides of the thin wall. And that is why it is possible to achieve a straight top surface of a thin wall with these strategies.

3.1.2 Residual stress

Table 2 shows the measured deformations corresponding to the individual production strategies. These deformations are the result of the relaxation of residual stresses after partial cutting off the specimens from a base plate (see Fig. 5).

Strategy 1 showed the least amount of deformation. The direction in which the specimen was cut from the base plate may cause a tiny deformation. The sample was cut off at the side where was the insufficient material amount and low residual stress.

On the contrary, strategy 3 showed the highest deformation of the sample. Strategies 1 and 3 showed considerable geometric inaccuracies and were therefore not used in further experiments. Strategies 2, 3 and 5 showed similar deformations of the samples. Dividing the torch paths into shorter partial weld beads did not significantly reduce the residual stresses. No significant trend was observed in the relationship between the production strategies and the formation of residual stresses.

3.2 Overhangs production

Four different fabrication strategies were tested for different overhang angles. Fig. 6 shows the resulting samples for four different overhang angles, including 30°, 45°, 60° and 90°. The fabrication strategies included strategies 2, 4 and 5 described in previous experiments with the orientation of the torch in the printing direction (see Fig. 2(a)) and strategy 2_z with the orientation of the torch in the vertical position (see Fig. 2(b)).

Almost all overhangs were produced without failure and without the need to interrupt the production process. Only the 90° specimen production with strategy 2_z had to be interrupted during manufacturing due to poor material deposition and the risk of damage to the welding torch. Significant imperfections were found during the 45° and 60° specimens manufacturing with strategy 2_z. Generally, the larger the overhang angle, the more difficult it is to fabricate a sample.

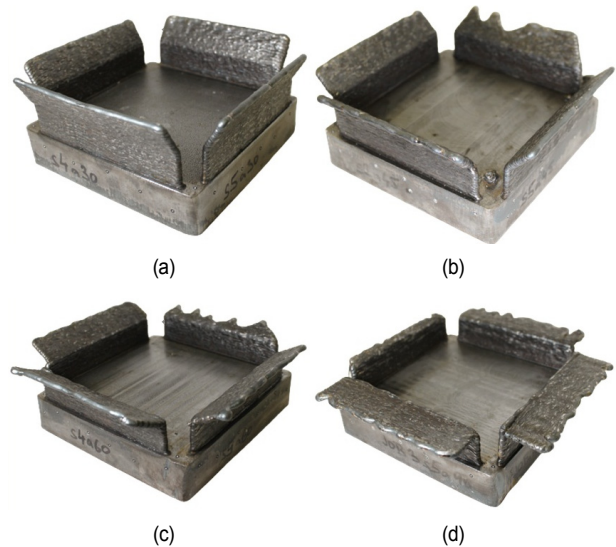


Fig. 6. Overhangs: (a) 30°; (b) 45°; (c) 60°; (d) 90°.

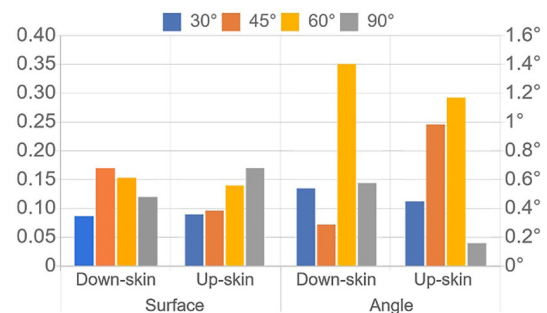


Fig. 7. Strategy 4 - deviations graph (left y-axis - standard deviation, right y-axis - angular deviation).

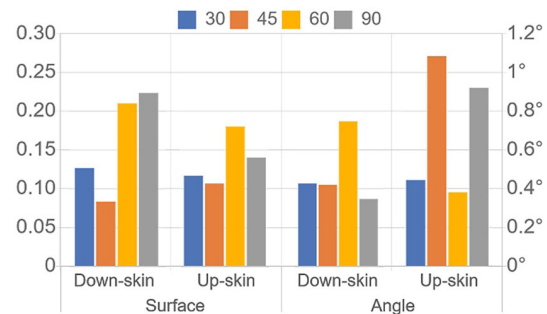


Fig. 8. Strategy 5 - deviations graph (left y-axis - standard deviation, right y-axis - angular deviation).

3.2.1 Accuracy measurement

The best results in terms of dimensional accuracy in the production of overhangs, i.e. the lowest deviations, were achieved with strategies 4 and 5. All overhang samples produced with these strategies have an angular deviation of less than 1.4° and a standard surface deviation of less than 0.22. Figs. 7 and 8 show these deviation results. With these strategies there were no significant differences between the down-skin and up-

skin deviation values. With strategies 4 and 5, all overhanging samples could be produced with a low angular deviation and sufficient surface quality. However, the production of more inclined walls was demanding even with these strategies. The more inclined the specimen, the more significant the irregularities at the weld bead joints. Therefore, the upper surface was not so straight as with simple walls without an inclined surface.

An overhang of 30° was produced without problems, also with strategy 2 and strategy 2_z. Problems in production with strategy 2_z already occurred with a 45° overhang. There was a collapse of the weld pool on the overhanging surface, and it caused the formation of craters on the sample. Therefore, this production strategy is not suitable for overhangs greater than 30°.

Strategy 2 did not result in the formation of craters and collapse of the weld pool, but the production was significantly less accurate than strategies 4 and 5. The angular deviation reached a value of 4.5°, and the standard deviation was 0.3 when using strategy 2. These values are significantly higher than those fabricated with strategies 4 and 5. This reduction in production accuracy could be due to insufficient cooling of the object between layers.

4. Discussion

For the production of complex parts using WAAM technology, it is necessary to master the production of partial characteristic shapes. Examples of such geometry features are thin walls and overhangs. A common problem with welded parts is the formation of residual stresses. There are several ways to influence the level of residual stresses. One of them is the usage of appropriate welding strategies. This chapter discusses the influence of production strategies on geometric accuracy and the formation of residual stresses in thin-walled and overhanging parts.

4.1 Thin walls

During the production of thin walls with the proposed strategies, the correct arrangement of the starting and ending points of the weld bead proved to be essential. An excessive amount of material was applied at the starting points and insufficient at the endpoints. Previously, Zhang et al. [9] described this behaviour of weld beads. This phenomenon is caused by the rapid heat dissipation at the starting point of the weld bead, where the penetration depth of previous welds is shallow. The low penetration depth leads to the solidification of the weld pool with a greater layer height than in the middle of the weld bead, where the penetration is greater. An insufficient amount of material is deposited at the end of the weld bead. It is due to the flow of the molten metal, and skewing occurs. After several layers of material are produced, there is a significant deviation from the desired shape. This problem can be solved in several ways, for instance, by compensating the speed of the welding torch at the start and end points or by using offsets as de-

scribed in this paper. In this paper, compensation of the start and end points by placing the same number of start and end points in layers above each other was presented. The offsets were used to compensate for excess and insufficient material in the joints of the weld beads inside the fabricated sample.

A comparison of the residual stresses of thin-walled specimens produced using different strategies was realized. The original hypothesis assumed that dividing long weld beads into several shorter ones should lead to more uniform cooling and the generation of fewer residual stresses. However, the CMT welding mode was used in specimen production. This mode significantly reduced both the energy input to the weld pool and the residual stresses. The CMT mode reduced residual stresses to such an extent that splitting the long weld beads into shorter ones did not lead to a further significant reduction in residual stresses. On the other side, adding more weld starts and ends can increase residual stresses if the strategy is inappropriately designed (see strategy 3). The effect of significantly reducing residual stresses by using strategies was not observed. To describe this effect more precisely and possibly confirm the original hypothesis, it would be necessary to change the design of the samples. The desired result may appear when using strategies with longer partial weld beads in a larger object overall.

4.2 Overhangs

With strategy 2_z, in which the welding torch was oriented vertically during the entire production, only 30° overhangs could be built. The production of more inclined surfaces resulted in the burn-off and collapse of the weld pool. The collapse of the weld pool resulted in the formation of craters and large deviations. Turning the welding torch in the printing direction can effectively suppress this effect. All other tested strategies used the print head orientation in the print direction. The best results were obtained with strategies 4 and 5. The partial weld beads were welded in a specific sequence, which resulted in more uniform cooling and better accuracy and surface quality.

To clearly state that the improved accuracy and surface quality are due to these strategies, it would be necessary to use an interpass temperature monitoring system. The poorer accuracy and surface quality with strategy 2 may be due to insufficient material cooling between layers. During production, offsets were added in the z-direction at each weld start and end point. These offsets were always 100 mm at a speed of 0.005 m/s. This fact led to the strategy with more weld start and end points having a longer overall cooling time for each layer.

This study confirms that the fabrication of overhanging structures using WAAM technology is feasible even without changing the orientation of the fabricated part. The production strategy can influence the possibilities of overhang fabrication. However, during the production of inclined surfaces, the weld pool is affected by gravity forces. As a result, the humping effect occurs. Several publications have dealt with the forma-

tion of humping during down-hand welding. From these publications, it appears that the humping increases with increasing TS. Yuan et al. [10] focused specifically on wire arc additive manufacturing of inclined surfaces. This publication shows that the backward flow of the metal under the influence of gravity is responsible for the occurrence of humping. The greater the overhang angle, the easier the humping effect occurs, even at lower speeds. Therefore, productivity decreases when printing overhangs in this way. Thus, it is necessary to consider the humping effect, and when printing overhanging surfaces at a large angle, it is advisable to use a positioning device.

5. Conclusions

The study deals with strategies for wire arc additive manufacturing of thin walls and overhangs, where the influence of the manufacturing strategies on geometric accuracy and residual stress was described. The conclusions from the achieved results are as follows:

- Five different strategies for thin wall production were designed and compared. All production strategies resulted in the formation of solid walls. Strategies with a different number of starts and ends in layers above each other were evaluated as unsuitable for thin wall production. It was due to the asymmetry of the produced samples.
- The weld bead beginnings and ends location was found to be critical from the point of view of production accuracy and residual stress. No significant trend was observed in the relationship between the production strategies and the formation of residual stresses.
- It was possible to produce all types of inclined specimens (30°, 45°, 60°, and 90°) using strategies in which long weld beads were divided into several short partial welds. The produced overhangs show only a low angular deviation and sufficient surface quality. All overhang samples produced with these strategies have an angular deviation of less than 1.4° and a standard surface deviation of less than 0.22.

Follow-up research will focus on production strategies for printing large-scale thin-walled components. During the production of large parts, the influence of these strategies on residual stress formation and production time reduction could be more visible. Furthermore, optimization of heat dissipation by using production strategies is essential within large-scale manufacturing due to base plate deformation occurrence. A suitable application of mentioned strategies is the production of thin ribs welded on a thin sheet base material. An example of such a lightweight part can be aircraft components, where WAAM technology can save material and manufacturing costs.

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sign of process parameters and strategies for WAAM of thin-walled components.



OPEN Wire arc directed energy deposition of AZ61 magnesium alloy fabricated using cold metal transfer

Jakub Slaviček¹✉, Čeněk Šváb¹, Stefan Gneiger², Jakub Hurník¹, Petr Procházka³ & Daniel Koutný¹

This research investigates the challenges and potential of Cold Metal Transfer (CMT) for Wire-Arc Additive Manufacturing (WAAM, also known as WADED – Wire-Arc Directed Energy Deposition) of AZ61 magnesium alloys. Despite the excellent properties of magnesium alloys, their processing is challenging due to high vapour pressure, low boiling point, flammability, and difficulties in maintaining a stable welding process. This study examines the effects of key CMT parameters, including boost phase current and duration, burn phase current, electrode speed, and short-circuit (SC) phase current. The results demonstrate that boost phase current, cycle time, droplet surface tension, and material consumption significantly influence deposition size and shape. By optimising these parameters, a stable and efficient welding process was achieved, improving contact angle and ensuring sufficient penetration. The optimised conditions enabled the fabrication of a 50-layer thin-walled component, highlighting the study's contribution to advancing CMT applications in WADED processes.

Keywords Magnesium alloy, Processing, Mechanical properties, CMT, WADED, WAAM, AZ61

Additive technologies are becoming increasingly important for producing complex and weight-saving shapes, offering advantages such as design flexibility, material efficiency, and reduced time to market¹. Among the various AM approaches, Wire-arc Directed Energy Deposition (WADED), also referred to as Wire Arc Additive Manufacturing (WAAM), has attracted significant attention due to its high deposition rates, low feedstock and equipment cost, and ability to fabricate large-scale structures^{1,2}. These advantages make WADED particularly suitable for applications in transportation and aerospace sectors, where lightweight and sustainable manufacturing solutions are essential.

Magnesium alloys are of special interest in this context because of their low density, high specific strength, and good thermal and damping properties, making them among the most promising structural materials for the 21st century^{3–5}. However, their processing remains challenging. Limited ductility at room temperature due to the hexagonal close-packed (HCP) crystal structure, together with a low melting and boiling point, makes stable welding of magnesium alloys difficult^{4–6}. This dilemma is further complicated by the difficulty in managing stable and continuous welding processes due to the physical properties of magnesium, such as its low density, low melting, and boiling

point^{7,8}. Conventional welding methods often result in defects such as porosity, hot cracking, and unstable arc behaviour, which limit their widespread industrial use.

WADED can be implemented using several arc welding processes including gas tungsten arc welding (GTAW), plasma arc welding (PAW), or gas metal arc welding (GMAW)⁹. GTAW welding in combination with light metals is used mainly because of its high welding arc stability and lower weld fume formation¹⁰. The disadvantage of this system in combination with WADED is the need for additional equipment for welding torch rotation and the associated limitation of welding torch orientation when printing complex structures. In addition, GTAW productivity is approximately

2–3 times lower than GMAW¹⁰. GMAW is widely used due to its high efficiency, excellent gap-bridging capability, and compensates for the loss of alloying elements during welding¹¹. However, excessive heat input during GMAW causes issues such as overheated droplets, grain coarsening, oxidation and evaporation, thermal stresses, and hot cracking¹². To overcome these drawbacks, the Cold Metal Transfer (CMT) variant of GMAW was developed, which retracts the wire during the short-circuit phase and thereby lowers the heat input and

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spatter¹³. CMT has shown promising results for the cladding and welding magnesium alloys by mitigating many issues like droplet explosion and thermal stress¹⁴.

Several recent studies have focused on CMT-WADED magnesium alloys processing^{15,16}. Wang et al. investigated the microstructure and mechanical properties of AZ31 alloy samples produced using three different CMT and CMT-P (a combination of CMT and pulse welding) parameter settings¹⁷. Their results showed that welding characteristics and parameters such as I_{boost} , $t_{I_{boost}}$, I_{sc_wait} , vd_{sc_wait} , and I_{sc_2} significantly affect the shape of the deposit. However, they only compared the outcomes of three parameter sets without analysing the influence of individual parameters in detail. Manjhi et al.¹⁸ and Zhang et al.¹⁹ studied the effect of travel speed (TS) and wire feed speed (WFS) in CMT-WADED processing of AZ31. They demonstrated that these parameters strongly influence deposit geometry, but their work did not address the role of specific CMT characteristics or other key parameters. Similarly, Bi et al.²⁰ examined the microstructure and mechanical properties of WADED AZ91 alloy, focusing on producing thick-walled components using a triangular weaving mode. While their study provided valuable insights, it did not analyse the effect of individual parameters on deposit shape. Gneiget et al.²¹ explored WADED processing of AZ61 magnesium alloy using CMT, but this alloy was briefly discussed. Their study's primary aim was to develop a new alloy (AEX11) and compare its microstructure and mechanical properties with AZ61, without a detailed analysis of process parameters. In contrast, Ying et al.²² investigated WADED of AZ61 alloy using a GTAW heat source, partially considering the effect of processing parameters. Additional studies have also examined GTAW-WADED of AZ31²³ and AZ80M²⁴ alloys. However, due to fundamental differences between GTAW and CMT processes, findings from GTAW cannot be directly applied to CMT. Therefore, a systematic investigation of individual CMT process parameters in WADED of magnesium alloys is still needed to clarify their role in controlling deposit geometry, microstructure, and mechanical properties.

Previous studies have demonstrated the potential of CMT for magnesium alloy processing and provided insights into the influence of selected parameters such as wire feed speed (WFS), travel speed (TS), different process modes (CMT, CMT-P, CMT-ADV), or deposition strategies (e.g., triangular weaving). However, a comprehensive description of the fundamental arc behaviour, specifically at the level of characteristic parameter adjustment, is still missing. In particular, the link between the current and voltage waveforms of the arc and the resulting deposit geometry has not yet been systematically established. Since deposit geometry directly affects build quality and process reliability, clarifying this correlation represents a critical step towards the effective industrial implementation of WADED for magnesium alloys.

This study provides a detailed analysis of how the main CMT characteristic parameters affect the geometry of single deposits. The objective is to identify parameter settings that yield a favourable contact angle and sufficient deposition depth across the cross-section. By clarifying the role of key parameters in welding behaviour, the work establishes guidelines for the reliable production of AZ61 alloy. Furthermore, the microstructure and mechanical properties of thin-walled parts fabricated with the optimised parameter set are evaluated. These results are compared with properties of materials produced by alternative manufacturing routes, thereby assessing the potential of CMT-WADED for the fabrication of thin-walled AZ61 components.

Materials and methods

The overall research progress is summarised in Fig. 1. The study begins with a literature review to identify research gaps and relevant CMT parameters to establish a basis for comparison. Based on the review, the range of tested CMT characteristic parameters was determined and tested in single-deposit experiments to evaluate their influence on weld geometry and cycle stability. From these results, the most suitable parameter configuration was determined. The best parameter set was applied to thin-wall fabrication, followed by microstructural and mechanical testing. Each step provided outcomes that guided the next stage of the research.

Process parameters

The single deposit test was conducted on the base material (BM) with 100×20×8 mm dimensions at room temperature before welding. The BM was fabricated from AZ91D magnesium alloy (composition in Table 1). AZ61 wire with a diameter of 1.6 mm was used as a filler material with a chemical composition provided by the distributor, as listed in Table 1. Deposits were made using a Fronius TPS 3200 CMT welding system (Fronius International GmbH, Pettenbach, Austria). The positioning movement of the welding torch was accomplished using a KUKA KR 60 HA robotic arm (KUKA AG, Augsburg, Germany).

During the tests, a constant speed of the welding torch (WS) of 10 mm·s⁻¹ and a distance from the manufactured part of 13 mm was maintained. Pure argon (99.5%) was adopted as the shielding gas with a flow rate of 18 l·min⁻¹.

A DC-CMT characteristic given by the synergic line (G3Sil), according to Wang et al.¹⁷, with modifications based on previous experiments was used as a mean value throughout the whole experiment. The characteristic is shown in Fig. 2 and consists of the 3 main phases: melting of the wire while the movement of the electrode goes upwards (boost phase), returning of the wire back to the weld pool (burn phase) and the contact and transfer of the melted droplet (SC phase).

The single-deposit experiment consisted of 21 combinations listed in Table 2 created by altering different parameters of the CMT characteristic. Five main characteristic parameters can be controlled during the CMT welding process. I_{boost} (A) – boost phase current, $t_{I_{boost}}$ (ms) – boost phase duration, I_{sc_wait} (A) – burn phase current, vd_{sc_wait} (m·min⁻¹) – the speed of electrode during the boost and burn phase and I_{sc2} (A) – SC phase current. The RCU 5000i controller was used to alter the synergic line and CMT parameters.

For an accurate description of how individual parameters affect the geometry of a single deposit, it is essential to precisely measure the current and voltage signals of the welding arc. Figure 3 presents a schematic diagram of the experimental setup for recording the characteristic waveforms. The voltage and current signals were

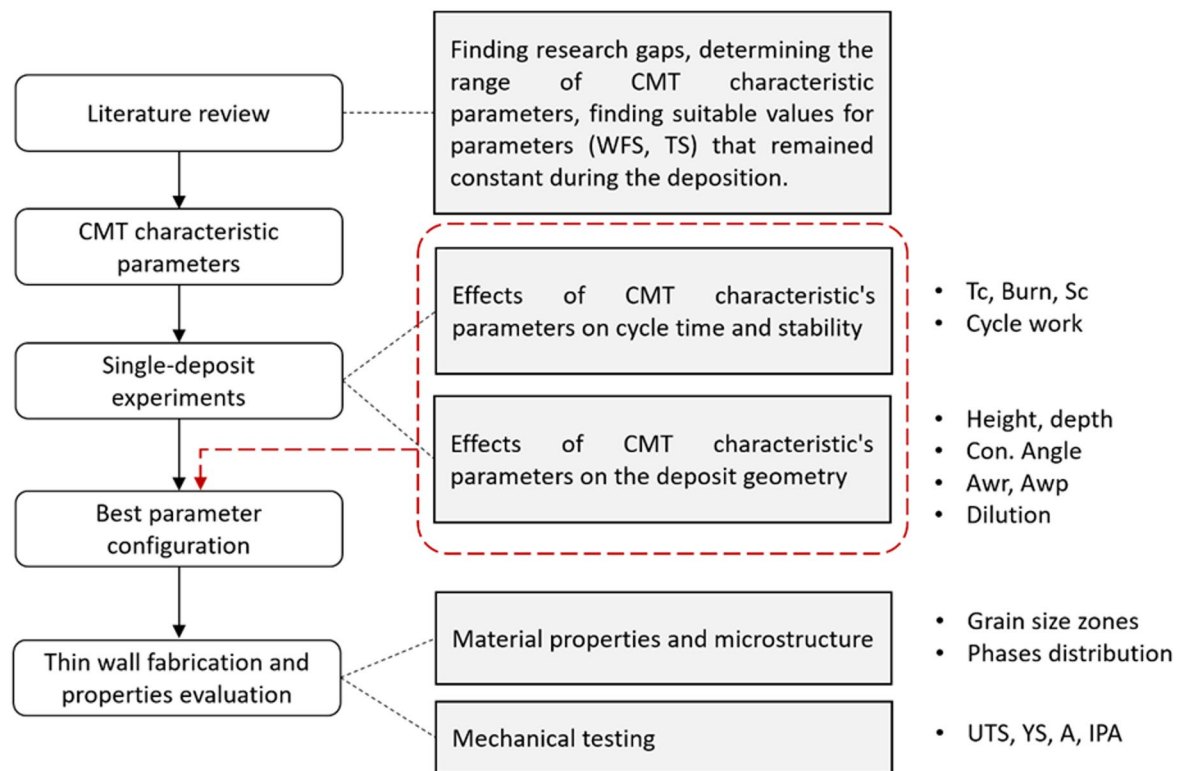


Fig. 1. Flow chart of the research progress.

Alloy	Elements	Al	Zn	Mn	Si	Fe	Cu	Ni
AZ61 (filler wire)	Min (%)	5.50	0.5	0.15				
	Max (%)	6.50	1.50	0.40	0.10	0.005	0.05	0.005
AZ91 (base material)	Min (%)	8.30	0.35					
	Max (%)	9.70	1.00	0.13	0.10	0.005	0.03	0.002

Table 1. Chemical composition of AZ61 (wire) and AZ91 (BM) magnesium alloys.

captured with a Tektronix DPO 2024B (Tektronix, Beaverton, OR, USA) oscilloscope at a sampling frequency of 312.5 kHz.

Individual phases of the CMT cycle were measured based on diagram given in Fig. 1. The average power of the cycle was calculated by the value of voltage ($U_{(x)}$) and current ($I_{(x)}$) in every step (n) by Eq. 1:

$$P_c = \frac{1}{n} \sum_{x=0}^n U_{(x)} \cdot I_{(x)} \quad (1)$$

Geometry analysis, dimensions of samples

Overall, 21 samples were fabricated and analysed, individual samples were cut, and cross-sections were polished using a series of 3 grinding discs with grit sizes: P320, P500, and P1000. Subsequently, they were polished with a P3000 paste, and the geometry and macrostructure of the deposits were observed using an Olympus SZX7 microscope (OLYMPUS, Tokyo, Japan). A diagram of the main dimensions of the deposition is shown in Fig. 4. Awp represents the penetration area (below the width line). Awr indicates the area of reinforcement (above the line).

Based on the findings obtained from the comprehensive parametric study, a set of optimized parameters suitable for WADED was determined. The schematic diagram of the CMT-WADED process is depicted in Fig. 5. The component dimensions were designed to allow the extraction of tensile specimens according to DIN 50,125. The start and end points of the trajectories were positioned outside the regions designated for specimen machining to avoid local inhomogeneities in the tensile samples.

A 100 s interlayer waiting time was applied. Using this methodology, a thin-walled component consisting of 50 layers with dimensions of 130 mm x 30 mm x 130 mm was successfully fabricated.

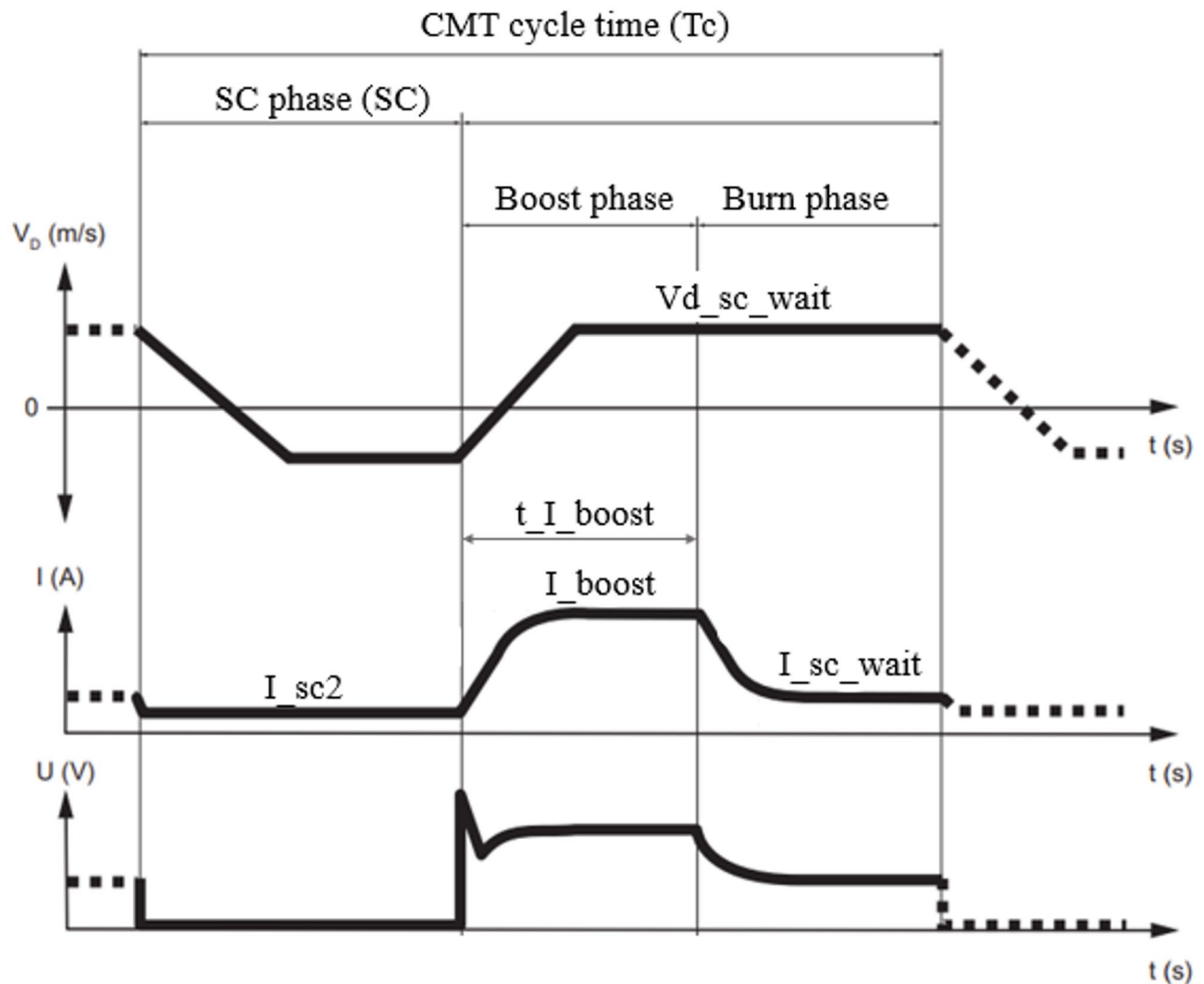


Fig. 2. CMT characteristic welding process (Modified²⁵).

Material properties and microstructure

Material properties were investigated to describe the achieved microstructure (defects, material phases, grain size) and clarify the material's mechanical behaviour.

Thermodynamic calculations in the Equilibrium state and using the Scheil model were performed using Thermo-Calc software (Version 2024a) and database TCMG6 (Thermo-Calc Software AB, Solna, Sweden). The calculation was carried out using the average chemical composition in Table 1.

Microstructural investigations were done by means of optical light microscopy (OLM) and scanning electron microscopy (SEM) on polished and polished + etched (picric acid) samples. For OLM, an optical microscope OLYMPUS BX53M with a digital camera OLYMPUS SC180 (OLYMPUS, Tokio, Japan) was used, and the microstructural images were taken by the software OLYMPUS Stream Motion 2.5. A Tescan MIRA 3 scanning electron microscope (TESCAN, Brno, Czech Republic) equipped with a field emission gun (FEG) operated at 15 kV was used for SEM analysis. The electron images were taken using a four-quadrant backscattered electron detector (TESCAN, Brno, Czech Republic) operated at a working distance of 15 mm. Energy-dispersive X-ray spectroscopy (EDS) was performed using an Octane Elect silicon drift detector (EDAX, Mahwah, NJ, USA). The fracture surfaces of the tensile specimens were examined using a digital light microscope (Keyence VHX-6000, Z250R lens, 250× magnification).

Mechanical testing

The samples for microstructural investigation and tensile tests were extracted from the thin-walled part in the way you can see in picture Fig. 6. The tensile specimens had dimensions specified by DIN 50125 (type E) and a thickness of 3 mm. Tensile tests were conducted at room temperature using a universal testing machine (Shimadzu AGX-V 100kN, Shimadzu Corporation, Kyoto, Japan). The deformation rate during testing was 0.5 mm·min⁻¹.

Deposit	I_boost	t_I_boost	I_sc_wait	vd_sc_wait	i_sc2
	(A)	(ms)	(A)	(m·min ⁻¹)	(A)
1	360	2	35	30	40
2	400	2	35	30	40
3	430	2	35	30	40
4	450	2	35	30	40
5	480	2	35	30	40
6	430	1.7	35	30	40
7	430	2.3	35	30	40
8	430	2.6	35	30	40
9	430	3	35	30	40
10	430	2	15	30	40
11	430	2	55	30	40
12	430	2	75	30	40
13	430	2	100	30	40
14	430	2	35	10	40
15	430	2	35	20	40
16	430	2	35	45	40
17	430	2	35	60	40
18	430	2	35	30	20
19	430	2	35	30	60
20	430	2	35	30	80
21	430	2	35	30	100

Table 2. The different settings of CMT characteristic used in experiment.

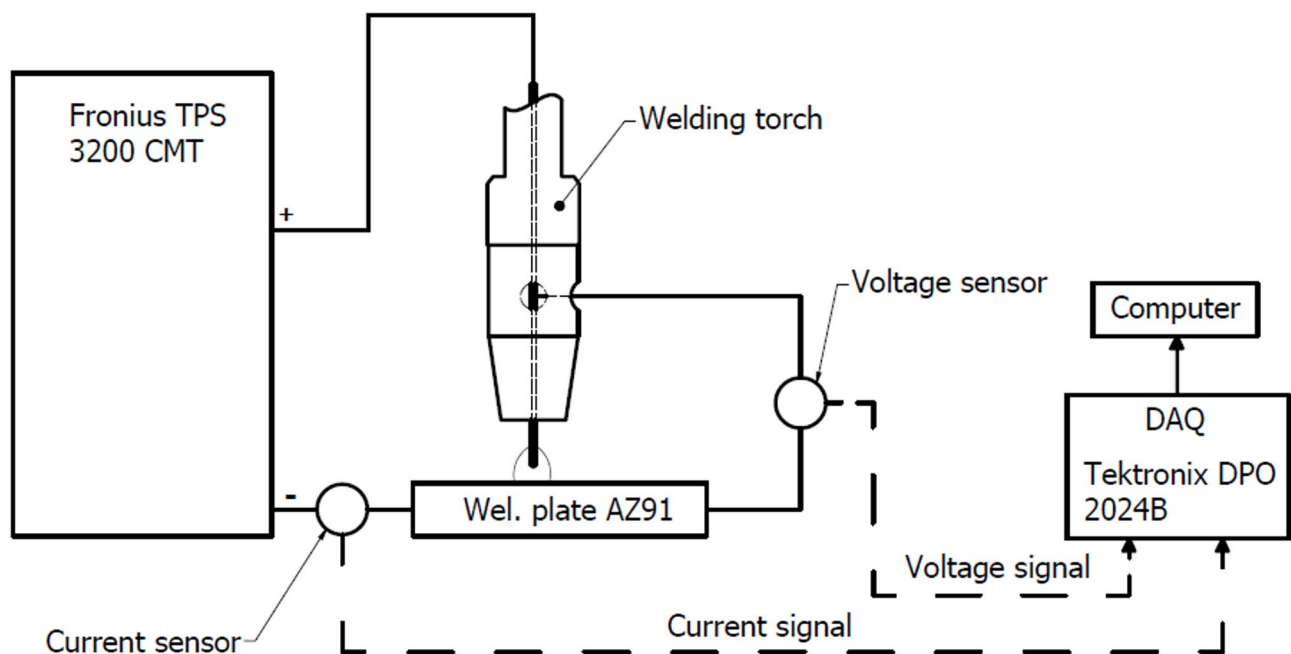


Fig. 3. Schematic of the experimental set-up of CMT welding process.

Ductility measurements were performed using a stereo 3D digital image correlation (DIC) system (Dantec Dynamics, Skovlunde, Denmark). The system incorporated two 5 MPx cameras (2448×2048 px, $3.45 \mu\text{m}$ pixel size) with 50 mm lenses. A baseline of 160 mm provided a 17° stereo angle appropriate for planar samples. The frame rate was set at 0.5 Hz. The correlation parameters were adjusted for better reliability – a facet size of 29 pixels with a spacing of 19 pixels was used.

The in-plane anisotropy percentage (%IPA) can be calculated to assess the degree of anisotropy in the tensile properties. The %IPA(X), where x represents the tensile property index, was calculated by the maximum value

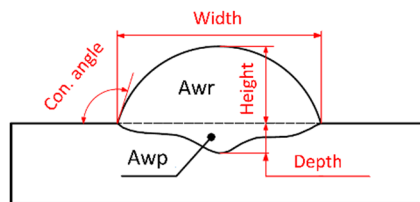


Fig. 4. Schematic of geometry of deposition.

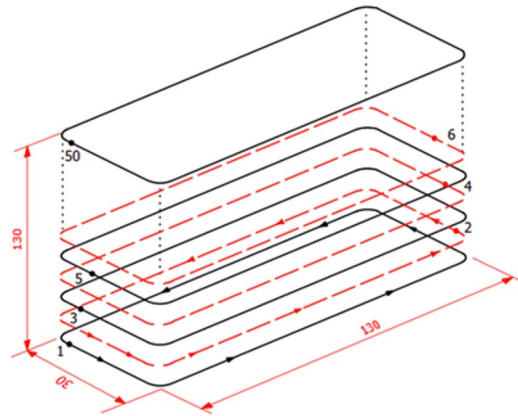


Fig. 5. The schematic diagram of the planned tool paths for production of the thin-walled component.

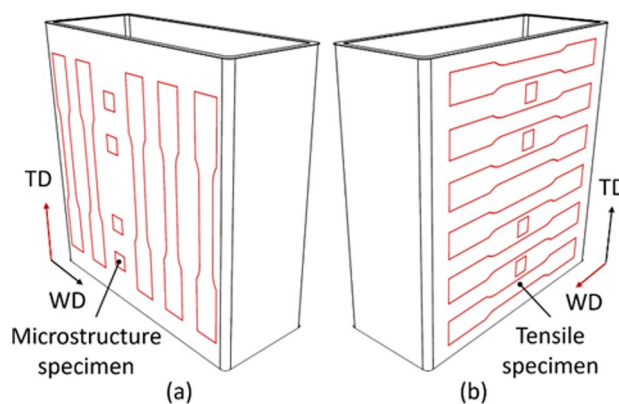


Fig. 6. Schematic illustration of specimen extraction: (a) transversal direction (TD); (b) welding direction (WD).

obtained from the building and travel directions (X_{max}) and the minimum value from the building and travel directions (X_{min}) by Eq. 2²⁶:

$$\%IPA(x) = \frac{X_{max} - X_{min}}{X_{max}} \quad (2)$$

Results and discussion

Effect of CMT characteristic

Boost phase current (I_{boost})

Increasing the current from 360 A to 480 A in the welding process led to noteworthy changes in heat input, deposition geometry, and cycle times. Initially, at 360 A, the heat input (P_c) was 1356.1 W, but it rose significantly to 2004.9 W at 400 A and further to 2396.7 W at the peak current of 480 A, see in Fig. 7 (a). Concurrently, the cycle time (T_c) was the longest (16.9 ms) at 360 A and then decreased sharply to 13.2 ms. With the further increase in current, the cycle times showed minor fluctuation.

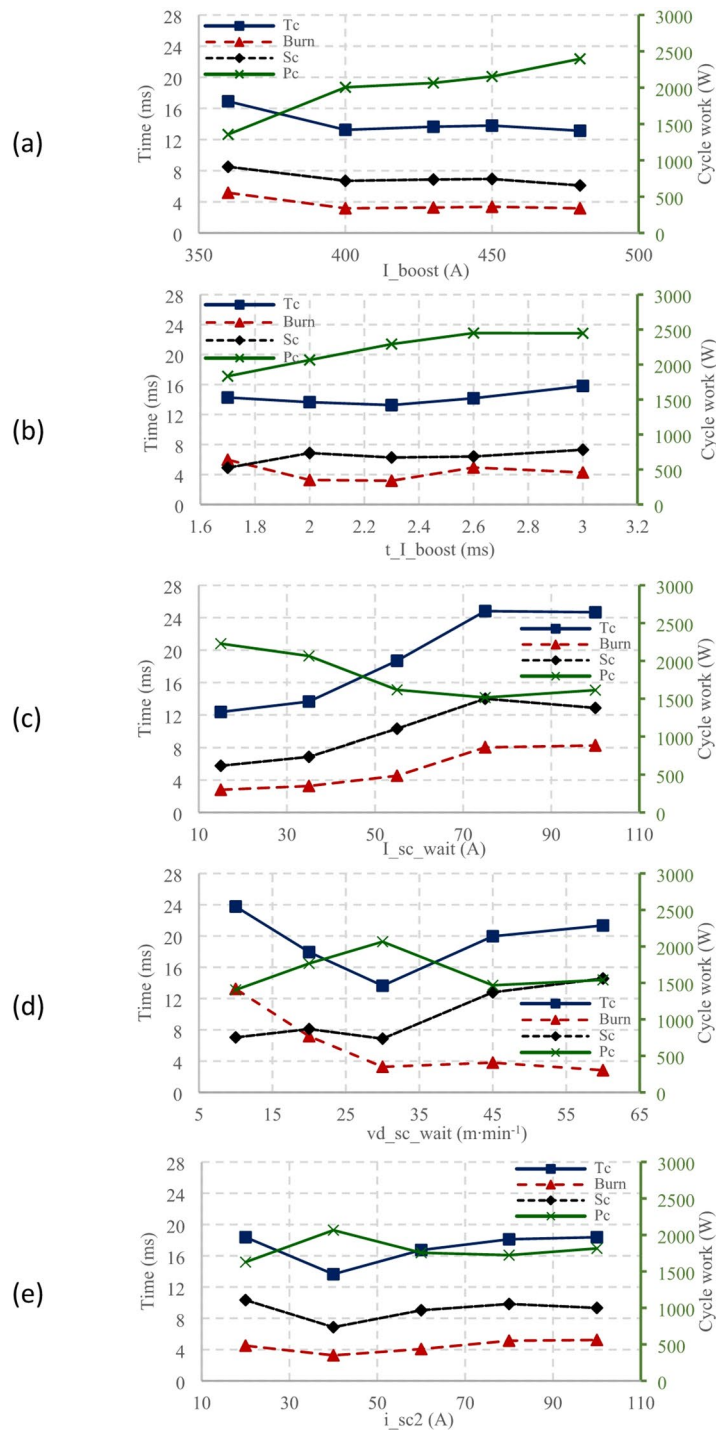


Fig. 7. Influence of characteristic parameters on energy input of CMT welding process: (a) I_{boost} , (b) $t_{I_{\text{boost}}}$, (c) $I_{\text{sv_wait}}$, (d) $vd_{\text{sc_wait}}$, (e) I_{sc2} .

The sudden drop in cycle time observed between 360 A and 400 A is atypical. As demonstrated in He's study²⁷, an increase in current typically leads to an increase in droplet size. Given that the droplet diameter exceeds the wire diameter, the duration of the burn phase would be expected to increase accordingly. However, as shown by²⁸, an increase in droplet size does not necessarily result in a longer burn time; instead, the burn phase duration remains relatively consistent despite the rising boost current. A similar trend was observed in this study in Fig. 7 (a) while increasing the boost current from 400 A to 480 A.

Regarding deposition geometry, Fig. 8 (a) and Fig. 9 (a), an increase in current led to the width expansion from 6.3 mm to 9.6 mm and a decrease in its height from 3.9 mm to 2.8 mm. This change is attributed to the variable surface tension and a smaller weld pool size, leading to a greater spread of the droplet on contact with

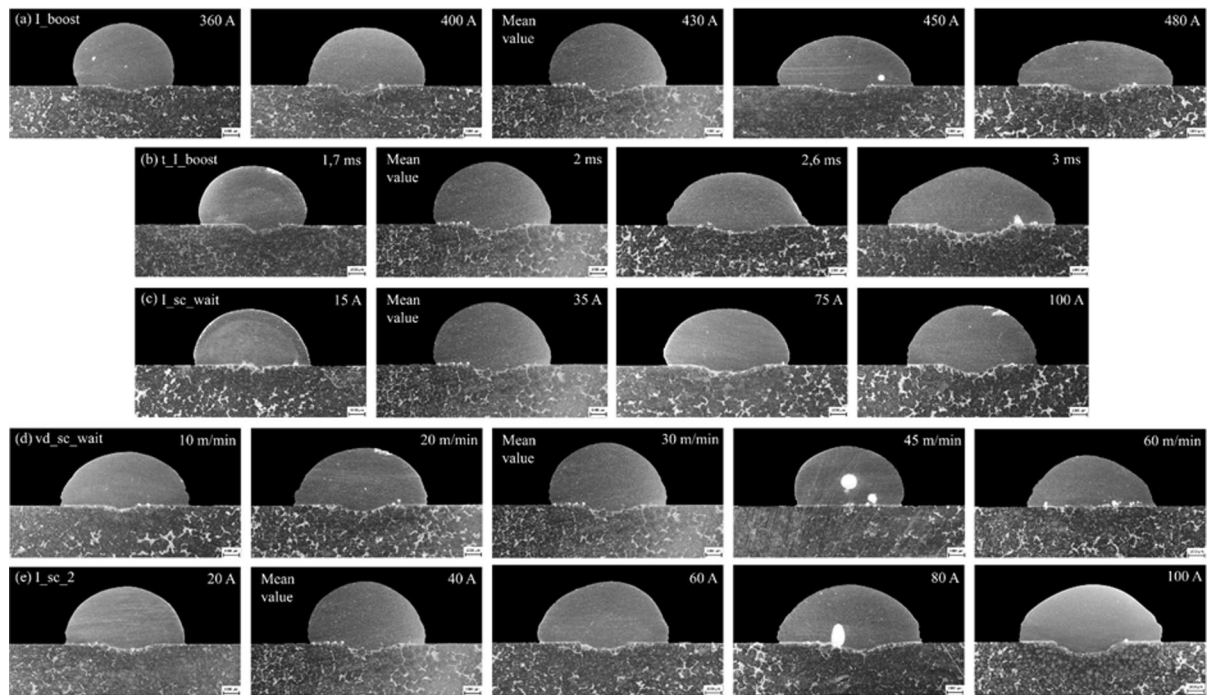


Fig. 8. Influence of parameters on deposition geometry: (a) I_{boost} , (b) $t_{I_{\text{boost}}}$, (c) $I_{\text{sc_wait}}$, (d) $vd_{\text{sc_wait}}$, (e) I_{sc2} .

the weld pool²⁸. The increase in current also resulted in an increased penetration depth from 0.34 mm to 0.51 mm. The table of precisely measured values of the deposit geometry, graphically represented in Fig. 8, is included in the dataset accompanying this publication.

The penetration area (Awp) showed some fluctuations. It increased from 0.53 mm² (at 360 A) to 1.26 mm² (at 400 A), followed by a minor reduction to 0.9 mm² (at 430 A) and a final increase to 1.6 mm² (at 480 A). The largest Awp area of 22.9 mm² was achieved at 430 A with a wire feed speed (WFS) of 6.1 m·min⁻¹. However, the penetration area decreased slightly for currents of 450 A and 480 A. This was likely because the boost current exceeded the spray transition current, and the duration of the boost phase was long enough to allow for some spray transfer and accompanying spatter to occur^{6,28}.

Boost phase duration ($t_{I_{\text{boost}}}$)

As the boost time increased, the average power also rose, reaching a peak of 2450 W at 3 ms. The total cycle time exhibited a convex trend, with a minimum duration of 13.2 ms observed at a boost time of 2.3 ms. The correlation of the trend in burn and short-circuit phases can be seen in the graph Fig. 7 (a) of the I_{boost} parameter. In contrast, in the graph Fig. 7 (b), increasing the $t_{I_{\text{boost}}}$ did not lead to a similar obvious correlation between the mentioned parameters.

In deposition geometry Fig. 8 (b) and Fig. 9 (b), increasing the boost phase duration led to an increase in the deposition's width (from 7.1 mm to 10.9 mm) and minimal decrease in its height (from 3.9 mm to 3.7 mm), which is well within the margin of error. These geometry changes resulted in a higher contact angle from 68° to 110° and an increased weld penetration depth from 0.34 mm to 0.73 mm.

Increased boost phase duration expanded the weld bead area. It led to no significant variation in bead height, but a noticeable increase in bead width. When the parameter was set to 3 ms, a substantial increase in Awp, followed by a slight increase in Awp, was observed. Simultaneously, a reduced cycle length followed the same pattern as Fig. 7 (a) at the beginning. This reduction was due to an extended burn phase at a boost phase duration of 1.7 ms, likely because the boost phase ended before the electrode's oscillatory movement²⁸. The overall cycle length increased for boost phase durations of 2.6 ms and 3 ms due to the larger droplet size.

Burn phase current ($I_{\text{sc_wait}}$)

The highest average cycle work of 2228.5 W was measured at a current of 15 A. Changes in this parameter showed a strong correlation with the current magnitude, akin to the I_{boost} and $t_{I_{\text{boost}}}$ parameters, with exceptions observed at 75 A and 100 A due to increased cycle durations reaching 24.8 ms and 24.6 ms, respectively. These duration changes led to corresponding increases in both the burn and SC phases, as illustrated in Fig. 7 (c).

Only slight increases in height, width, and contact angle with the $I_{\text{sc_wait}}$ parameter were seen, unlike the significant changes in the other parameters. However, weld penetration depth did increase from 0.27 mm at 15 A to 0.59 mm at 100 A, as per Fig. 8 (c) and Fig. 9 (c). The wire feed speed remained constant at 6 m·min⁻¹ for 35 A and 75 A.

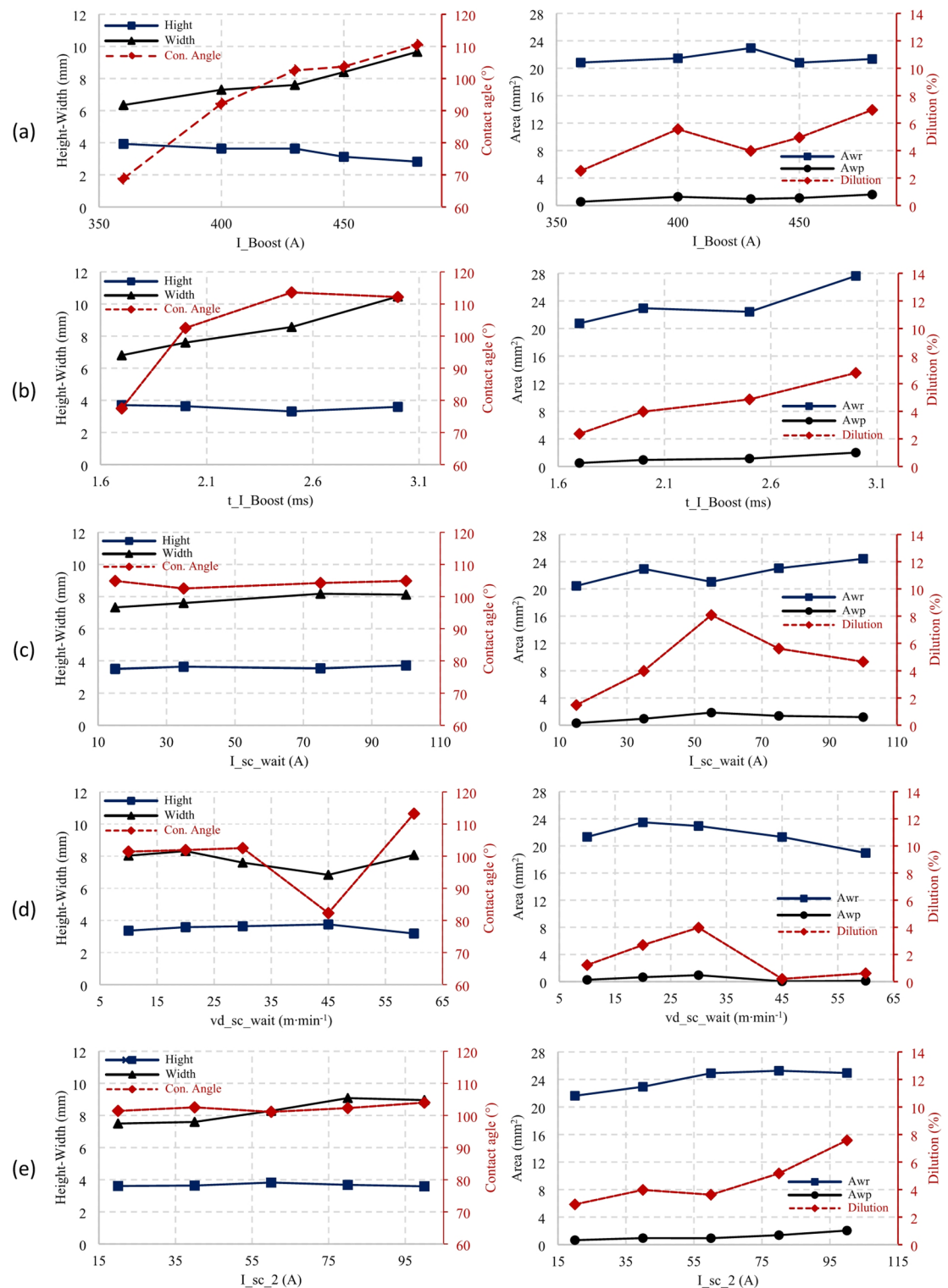


Fig. 9. Influence of characteristic parameters on deposition geometry: (a) I_{boost} , (b) $t_{I_{\text{boost}}}$, (c) $vd_{\text{sc_wait}}$, (e) $I_{\text{sc_2}}$, Awp – penetration area, Awr – reinforcement area.

The increase in the overall cycle time and the duration of each phase can be attributed to the interaction of two factors. During the boost phase, the droplet is melted, and upon transition to the burn phase, the continued current input supports further droplet growth. Simultaneously, the magnitude of the burn phase current induces a recoil effect, leading to droplet flattening and consequently prolonging the burn phase. When the droplet eventually contacts the weld pool, the electrode position is lower, and its subsequent emergence from the

solidifying weld pool takes longer, thereby extending the SC phase. A similar recoil behaviour was observed in Gomez's study²⁹.

Electrode speed during the boost and burn phase (vd_{sc_wait})

The study found the average cycle work mirrored the pattern of cycle length, peaking (2065 W) at the minimum cycle length of 13.6 ms and hitting a low (1404 W) at the maximum cycle length of 23.7 ms. In terms of geometry, the increase in speed leads to a slight increase in contact angle from 101° at $10\text{ m}\cdot\text{min}^{-1}$ to 115° at $60\text{ m}\cdot\text{min}^{-1}$, except at $45\text{ m}\cdot\text{min}^{-1}$, where the contact angle dropped to 82° , as shown in Fig. 9 (d). Penetration depth also dramatically dropped from 0.40 mm at $30\text{ m}\cdot\text{min}^{-1}$ to 0.11 mm at 45 and $60\text{ m}\cdot\text{min}^{-1}$. This corresponded to a decrease in Awr and Awp.

In addition to the aforementioned impact of electrode speed, the influence of this parameter on the stability of the characteristic profile was also observed, as shown in Fig. 10 (a, b). An unstable characteristic profile Fig. 10 (b) occurred at a $10\text{ m}\cdot\text{min}^{-1}$ electrode speed in the burn phase, which exhibited a significantly longer burn phase duration relative to the SC phase durations. Previous work observed a substantial increase in droplet size and burn phase duration in aluminium alloy welding at $5\text{ m}\cdot\text{min}^{-1}$. This increased surface tension and a deposition geometry composed of multiple continuous spot welds^{28,29}.

Analysing characteristics in Fig. 7 (d), burn phase duration decreased with speed until $30\text{ m}\cdot\text{min}^{-1}$, after which it stabilized. Conversely, the SC phase duration increased with speed due to smaller droplet formation and deeper electrode submergence into the weld pool²⁸. A cycle length increases

at $10\text{ m}\cdot\text{min}^{-1}$, $20\text{ m}\cdot\text{min}^{-1}$, $45\text{ m}\cdot\text{min}^{-1}$, and $60\text{ m}\cdot\text{min}^{-1}$ was accompanied by a decrease in cycle work. Finally, comparing maximum cycle work with penetration size showed the highest penetration at a speed of $30\text{ m}\cdot\text{min}^{-1}$, with penetration decreasing at speeds of $45\text{ m}\cdot\text{min}^{-1}$ and $60\text{ m}\cdot\text{min}^{-1}$.

SC phase current (I_{sc2})

The maximum average cycle work Fig. 7 (e), (2065 W) corresponded with the shortest cycle length (13.6 ms), and the minimum average cycle work (1627.2 W) with the most extended cycle length (18.4 ms). This correlation was reflected in the lengths of the burn and SC phases. When the current magnitude of the SC phase was changed Fig. 8 (e) and Fig. 9 (e), deposition height stayed constant (3.6–3.8 mm). Penetration depth increased from 0.38 mm at 20 A to 0.72 mm at 100 A. The contact angle stayed stable between 101° and 104° . The deposition area increased noticeably, from 0.65 mm^2 at 20 A to 2 mm^2 . Likewise, the Awr deposition area increased from 21.6 mm^2 to 25.2 mm^2 at 80 A, with a slight decrease to 24.9 mm^2 .

The I_{sc2} parameter controls the current magnitude from when the transferred droplet touches the weld pool until the electrode detaches from the pool. Despite the voltage always reaching minimum (non-zero) values in this phase, an increase in SC phase current led to increased penetration depth. This increase, however, caused subtle changes in deposition height and width.

Evaluation of parametrical study and WADED

The combination (based on previous findings) of parameters listed in Table 3 was selected to optimize deposition geometry while ensuring the stability of the process. These parameter settings were explicitly chosen to minimize

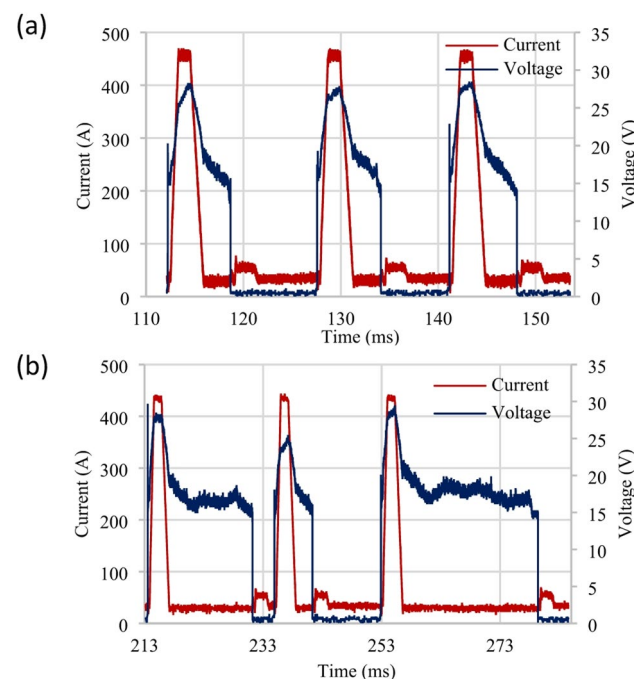


Fig. 10. Comparison of CMT characteristic: (a) Mean value, (b) vd_{sc_wait} ($10\text{ m}\cdot\text{min}^{-1}$).

I_boost	t_I_boost	I_sc_wait	vd_sc_wait	I_sc2
(A)	(ms)	(A)	(m-min-1)	(A)
430	2.5	35	30	50

Table 3. Table of CMT parameters used for WADED.

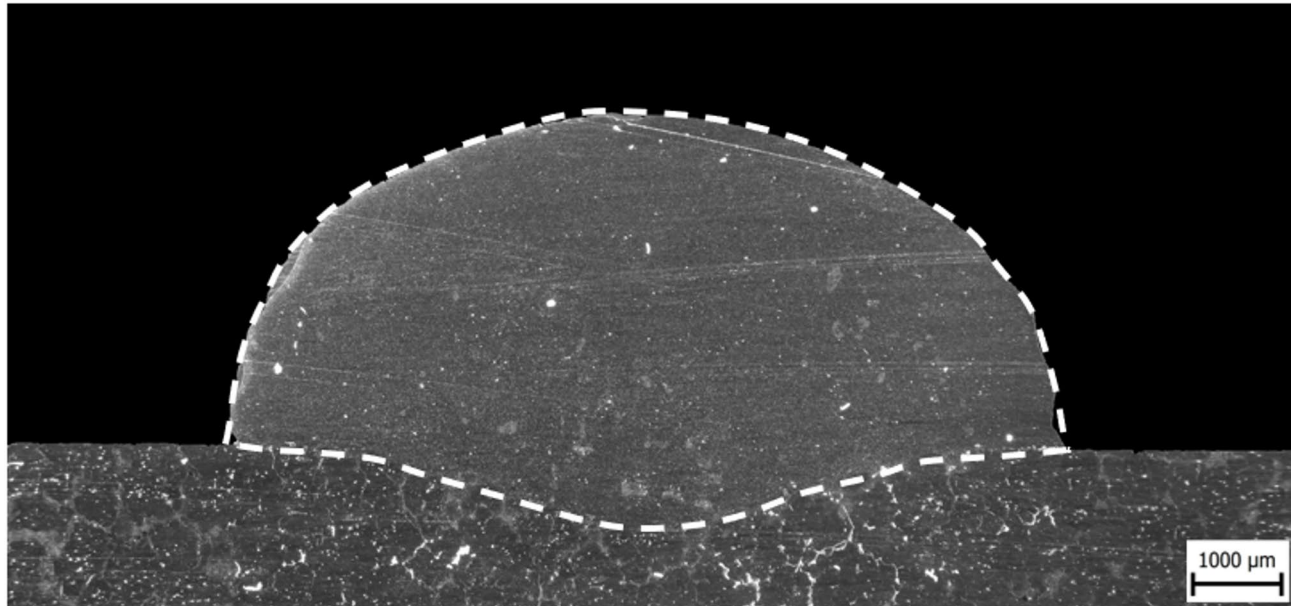


Fig. 11. Weld deposited with parameters used for WADED.



Fig. 12. AZ61 thin-walled object fabricated by WADED.

material loss, achieve the desired contact angle range of 90–120°, and prevent significant growth in deposition geometry.

The optimised parameters were implemented, resulting in a stable welding process. WADED benefited from the application of these parameters, leading to improvements in penetration depth, which was evenly distributed across the entire width of the deposition in cross-section, as shown in Fig. 11. The successful deposition exhibited a suitable contact angle of 108°, making it well-suited for thin-walled parts. The final part of Fig. 12 dimensions were 130 × 60 × 130 mm, with a wall width of 9.8 mm, indicating successful prevention of excessive overheating.

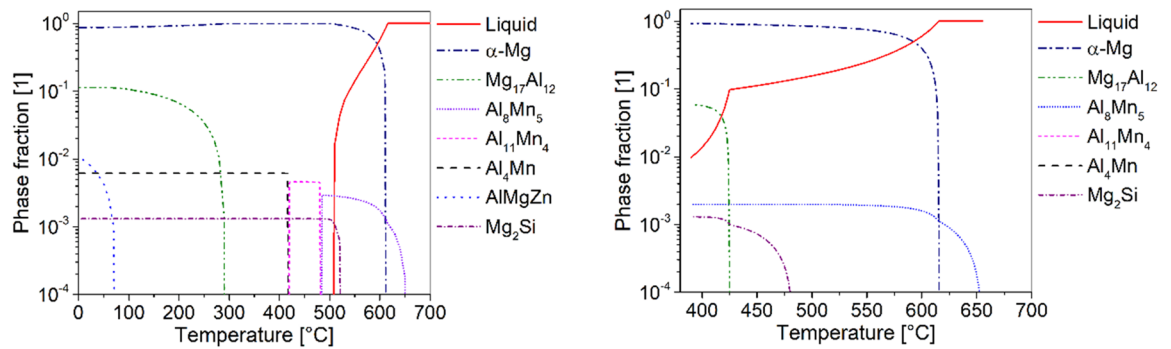


Fig. 13. Equilibrium (a) and Scheil (b) phase calculations for magnesium AZ61 using the average composition given in Table 1 (Fe, Cu, Ni omitted).

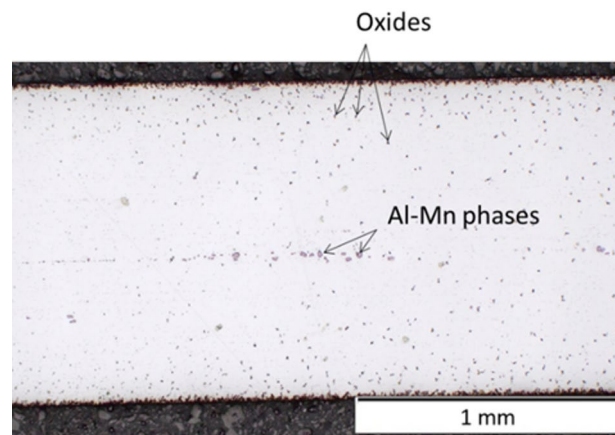


Fig. 14. Longitudinal section of the used Mg AZ61 wire.

However, it is essential to acknowledge that the production process encountered challenges. The presence of impurities in the wire and deviations in the welding trajectory introduced errors during manufacturing. Despite these obstacles, implementing the optimized parameters proved highly effective in achieving the desired deposition characteristics.

Material analysis

Phase fraction calculations for the AZ61 alloy based on the average composition in Table 1 are illustrated in Fig. 13, with (a) representing the equilibrium state and (b) using the Scheil equation. The elements Fe, Cu, and Ni were omitted from the calculations because, although they are common impurities, their content is low.

The calculations indicate that the intermetallic β -phase $\text{Mg}_{17}\text{Al}_{12}$ forms in significant quantities in addition to the Mg solid solution. This phase precipitates directly from the solid solution at temperatures below 300 °C according to the equilibrium calculation. However, at conventional (non-equilibrium) solidification in technical processes, this phase often precipitates primarily from the melt, forming a divorced eutectic β -phase at the grain boundaries⁸. This is supported by the Scheil calculation, where the $\text{Mg}_{17}\text{Al}_{12}$ phase forms at temperatures around 425 °C and, therefore, above the solidus temperature.

Furthermore, the calculations show that Al-Mn phases are formed primarily, with variable stoichiometries (Al_8Mn_5 , $\text{Al}_{11}\text{Mn}_4$ and Al_4Mn). It is known from practice that mainly Al_8Mn_5 is formed, which is stable down to room temperature³⁰. With silicon, which typically exists as an impurity in technical Mg-Al alloys, the intermetallic phase Mg_2Si can form. According to the calculations in Fig. 13, this phase forms primarily from the melt.

The equilibrium calculation also shows the formation of an AlMgZn phase, which precipitates from the solid solution at temperatures below 100 °C. However, this phase does not typically form during conventional processing conditions due to the very low diffusion rate at these temperatures.

The longitudinal section of the Mg AZ61 wire, captured using light microscopy, is depicted in Fig. 14. The wire is partially significantly oxidized, with oxides particularly prevalent on the wire surface, decreasing towards the centre. A possible cause could be improper storage of the sensitive Mg wire before processing. Furthermore, band-like arranged blocky Al-Mn phases with a maximum size of 30 μm can be observed, especially in the wire centre. It is assumed that these are primarily formed Al_8Mn_5 phases from the casting process, which align along

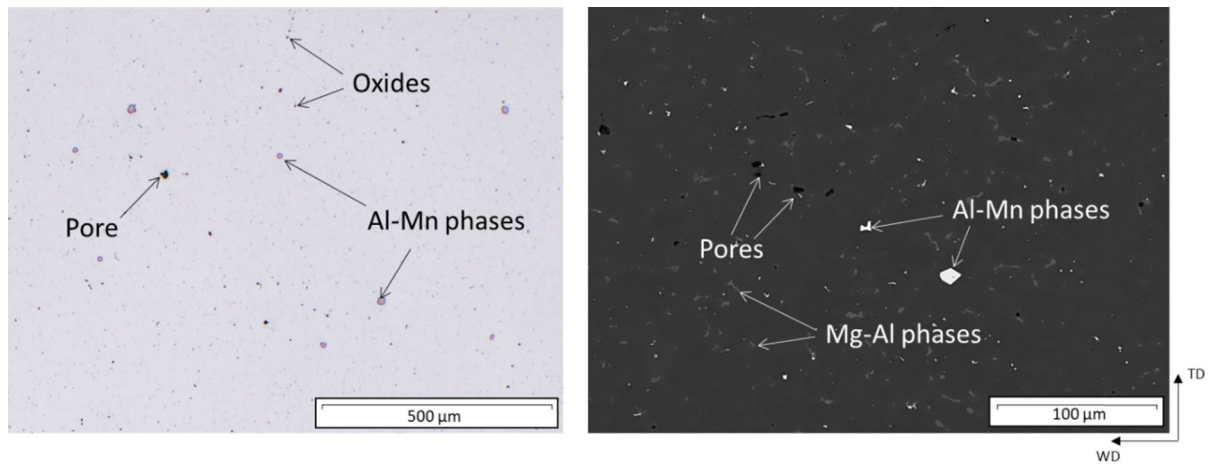


Fig. 15. Microstructure of a WADED sample.

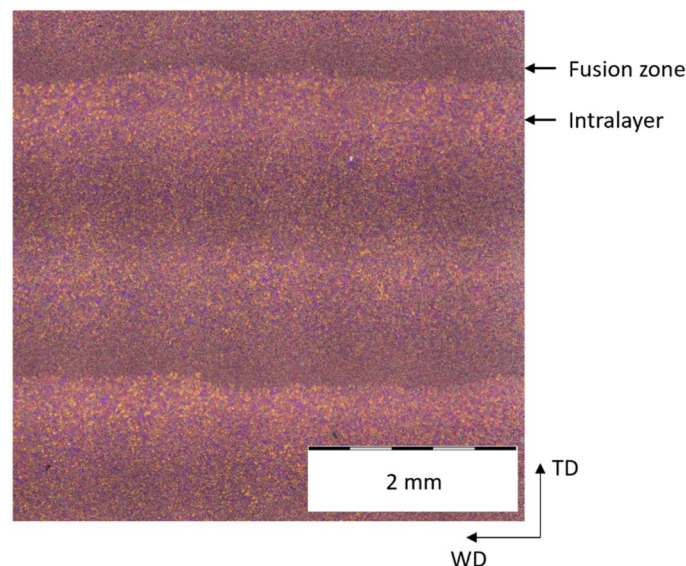


Fig. 16. Microstructure of an etched WADED sample observed under polarized light.

the deformation direction during wire manufacturing. In addition to the larger Al-Mn particles in the wire centre, the same phases can be found finely distributed throughout the wire.

Figure 15 shows the microstructure of a WADED sample. The oxides clearly visible in the wires (see Fig. 14) appear significantly smaller and more homogeneously distributed after deposition. Blocky Al-Mn phases are distributed throughout the material. Due to the high cooling rate of the WADED process, their size is somewhat reduced, reaching a maximum of 20 μm compared with that in the wire. Additionally, Mg-Al phases can be observed at the grain boundaries. It is assumed and supported by the calculations that these are β -phases ($\text{Mg}_{17}\text{Al}_{12}$), which appear as divorced eutectic. The β -phase content is low throughout the sample, so a continuous network does not form, but rather, the phase appears as individual islands. Sporadically, globular-shaped pores with sizes of up to 70 μm and localized shrinkage voids are discernible.

The microstructure of an etched WADED sample observed under polarized light is illustrated in Fig. 16. Due to the different grain sizes, the intralayer zone and fusion zone are clearly distinguishable. At the location where a new layer is applied to an already existing layer (fusion zone), the cooling rate is higher, resulting in a fine microstructure in the lower part of the new layer. Farther away from the underlying layer, the solidification rate is lower, resulting in an increased grain size. Similar microstructures were observed for Al alloys processed via WADED by Klein et al.³¹. Graf et al. have found that the zones in WADED AZ91 can also be distinguished by the different proportions

of β -phase³². However, this was not investigated in this work. Based on the micrograph, the layer height is approximately ~ 1.5 mm.

Direction	Ultimate tensile strength UTS (MPa)	Yield strength YS (MPa)	Elongation A (%)
Welding direction (WD)	239.15 ± 7.41	114.78 ± 5.42	7.96 ± 1.79
Transversal direction (TD)	173.34 ± 31.79	110.42 ± 6.14	4.16 ± 2.55

Table 4. Mechanical properties of AZ61 thin-walled object.

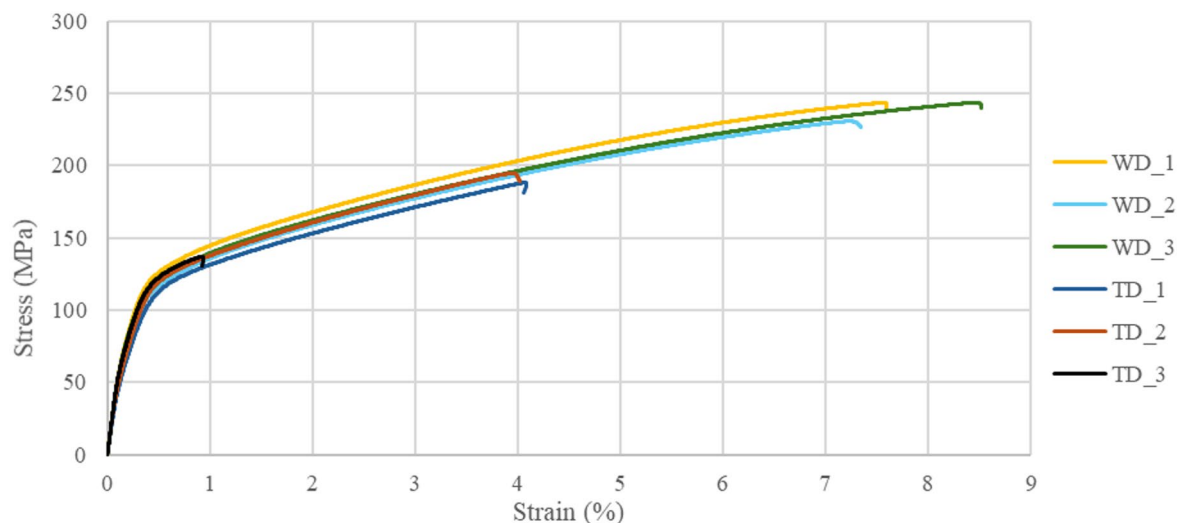


Fig. 17. Tensile results of WADED thin-walled object – WD (welding direction), TD (transversal direction).

Mechanical properties

Three samples oriented in the welding direction (WD) and three in the transversal direction (TD) were tested to obtain the mechanical properties of the AZ61 thin-walled object. The ultimate tensile strength (UTS), yield strength (YS), and elongation (A) were calculated from the stress-strain curves, and the results are listed in Table 4. Figure 17 shows the stress-strain curves of tensile specimens.

The results show that the average yield strength in the WD direction is 114.78 MPa, in the TD direction 110.42 MPa, and a %IPA(YS) anisotropy is 0.139. The average tensile strengths of the material are 239.15 MPa in the WD direction and 173.34 MPa in the TD direction, thus differing more significantly, and the %IPA(UTS) is 0.438. The greatest anisotropy is shown by the elongation %IPA(A) of 0.874, with average elongation values in the WD direction of 7.96% and the TD direction of 4.16%. The average values of yield strength, tensile strength, and elongation for the TD and WD directions, including standard deviations, are shown in the Table 4. Ultimate tensile strength in the TD direction is close to the values achieved in castings. Yield strength in both directions and ultimate tensile strength in the welding direction are higher than comparable casted magnesium alloys but lower than for AZ61 forgings.

The measured values indicate that the UTS and A of samples made from a thin-walled object exhibit significant anisotropy between the WD and TD directions. However, the overall anisotropy values can be partially distorted by significant differences between the behaviour of the samples in the TD direction.

Sample TD_3 shows significantly worse UTS and A values than the other samples in the TD direction. It can be caused, for example, by local defects in the sample. From the fractographic images of the fracture surfaces, as seen in Fig. 18 (a), it is evident that sample TD_3 contains noticeable large local pores both inside the sample and at its edge. Such porous and heterogeneous texture of fracture surface was observed only in sample TD_3. The other samples showed a homogeneous, smooth fracture surface without significant protrusions, similar to sample WD_1, see Fig. 18 (b). Such a defect at the edge of the sample TD_3 can become a stress concentrator and an initiator of crack growth in the material. This local defect would explain the deviation of the UTS and A values of sample TD_3 from the other samples in the TD direction.

However, even if deviated sample TD3 were not considered, the material would still show observable anisotropy in the TD and WD directions. The anisotropy calculated without this sample reaches approximately half the values. The anisotropy values without TD_3 are: %IPA(YS) is 0.139; %IPA(UTS) is 0.227, and %IPA(A) is 0.441. Even such anisotropy values are not satisfactory, but they are not unique. Several authors involved in WAAM production observed a similarly large anisotropy. Yang et al. 26 - Yang observed similar anisotropy values directly on the magnesium alloy AZ31. Yang et al. identified epitaxial columnar dendritic growth along the transversal direction as the leading cause of the anisotropy, whereas mechanical properties showed worse properties in the welding direction. In this work, the mechanical properties showed worse properties in the transverse direction, so another phenomenon probably causes the anisotropy.

Other common causes of anisotropy include the alternation of fusion zone and interlayer zone associated with applying material layer by layer. The material analysis section provides a more detailed description of why layers

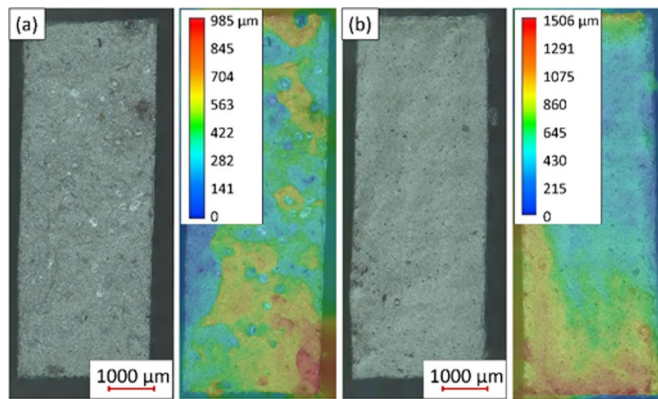


Fig. 18. Fracture surfaces after tensile testing (a) specimen TD_3 (left: optical image, right: 3D surface profile), (b) specimen WD_1 (left: optical image, right: 3D surface profile).

with different grain sizes were observed in the transverse direction. The anisotropy caused by different grain sizes across the transverse direction is described by Kumar et al.,³³ - Kumar who confirmed the connection between grain size and mechanical properties in WAAM samples according to the Hall-Petch relationship. Smaller grains act as barriers to dislocation motion, thereby increasing strength³³ - Kumar. Conversely, dislocations move more easily in regions with larger grains, leading to a deterioration in mechanical properties.

Another possible cause of anisotropy described by Kumar et al.³³ - Kumar is the effect of residual stresses. A sizeable thermal gradient occurs during manufacturing due to rapid heating and cooling cycles. These thermal cycles induce residual stresses due to uneven contraction of the deposited material.

Challenges and future work

This article is an initial study on developing suitable parameters for producing thin-walled parts from magnesium alloy AZ61. Based on the experience gained during the experiments, the process will be further optimised to achieve reliable production. The quality of the feeding wire is essential from the point of view of process stability and elimination of local defects. Impurities of any kind in the filler wire are also the cause of unwanted porosity²¹. It is necessary to store the wire correctly to prevent oxidation and clean it before production. For this reason, great emphasis will be placed on quality control of the input wire in subsequent experiments. It is also essential to maintain stable interpass temperatures and consistent conditions throughout production to minimise process instability. Real-time process monitoring further supports this by correlating arc stability with defect formation and helping to identify the sources of instability. Ensuring homogeneous process conditions and applying appropriate heat treatment after production can significantly reduce or completely eliminate anisotropy. Heat treatment reduces residual stresses and homogenizes grain size in the whole part, positively affecting mechanical properties. Guo et al. demonstrated that heat treatment enhanced mechanical properties in AZ80M fabricated using WADED and concluded that an appropriate heat treatment strategy can effectively eliminate anisotropy. Therefore, in follow-up studies, attention will be focused on process control and postprocessing, such as heat treatment or corrosion protection.

Compared to other additive technologies such as Laser Powder Bed Fusion (L-PBF) or Electron Beam Powder Bed Fusion (EB-PBF), WADED has several advantages, including production efficiency, acquisition cost, input material cost and the possibility of producing large-sized parts³⁵. Disadvantages include poorer dimensional accuracy, less complexity of the shapes fabricated and poorer surface quality. Mechanical properties of AZ61 alloy produced by L-PBF technology with appropriate heat treatment reach UTS 240 MPa, YS 124 MPa and EL 5.9%³⁶. These values are comparable to those achieved by WADED technology if proper heat treatment is applied. The mechanical properties of AZ61 alloy or similar ones produced by conventional technologies are as follows. Forgings of AZ61 alloy reach UTS 295 MPa, YS 180 MPa and EL 12%⁸. Castings are not usually made from AZ61 alloy; the most similar casting alloys are AZ63 and AZ91. The AZ63 sand-cast alloy achieves UTS of 200 MPa, YS of 97 MPa and EL of 6%, and the AZ91 sand-cast alloy achieves UTS of 165 MPa, YS of 97 MPa and EL of 2.5%⁸. The mechanical properties reported in this paper are higher than those for castings, lower than forgings, and approximately comparable to those produced by L-PBF. It is also worth noting that the optimised parameter set developed in this study was validated only for thin-walled geometries. Its applicability to bulk or thick-walled components, where heat accumulation, residual stresses, and microstructural development differ significantly, has not yet been confirmed. Future work will therefore extend parameter evaluation to volumetric parts and incorporate enhanced thermal-management strategies. Addressing these limitations will be a key step toward establishing a reliable manufacturing chain. Once these issues are resolved, WADED technology, in combination with magnesium alloys, can have a significant industrial impact. Since optimised components in the transportation sector often contain thin-walled and volumetric features, it remains essential to continue improving both the manufacturing process and the associated post-processing routes. In the long term, such developments could minimise material waste and promote sustainability through improved deposition efficiency.

Conclusion

This paper is focused on developing suitable parameters for producing thin-walled parts from magnesium alloy, identifying the impact of main process parameters on deposition geometry. Magnesium alloy wire AZ61 was deposited on base material by the CMT welding process using various characteristic parameters. The primary conclusions from these results are as follows:

- Boost phase current has significantly affected cycle time and droplet surface tension, resulting in wider and lower deposits. Change in boost duration resulted in rapid material consumption growth, resulting in a bigger deposition area.
- Electrode speed during the boost and burn phase significantly impacted the stability of the welding process and penetration depth. SC phase current directly correlates to penetration depth and has a minor effect on the width, height and contact angle.
- Through optimization of specific welding parameters, a stable welding process was achieved, leading to notable improvements in penetration depth and contact angle suitable for WADED.
- The tensile properties show apparent anisotropic behaviour due to the different grain sizes in interlayer and fusion zones. The average UTS, YS, and A of the welding direction specimens are distinctly superior to those of the transversal direction specimens. The anisotropy of the UTS, YS, and EL for the AZ61 deposit are 0.438, 0.139 and 0.874. Mitigating anisotropy in future applications will require filler wire quality improvement, precise control of thermal conditions during deposition, and implementation of suitable heat treatment strategies.
- Ultimate tensile strength in the TD direction of 173.34 MPa is close to the values achieved in castings. Yield strength in both directions, YS(TD) 110.42 MPa, YS(WD) 114.78 MPa and ultimate tensile strength in the welding direction 239.15 MPa, are higher than those for castings, lower than forgings, and approximately comparable to those produced by L-PBF.

In conclusion, adjusting CMT characteristic parameters can control the heat input and transfer process, thus resulting in the capability to optimize the welding process for WADED application.

Data availability

The data that support the findings of this study are openly available in Zenodo Repository at <https://doi.org/10.5281/zenodo.15055769>. The original version of the authors' preprint of this study is openly available in the Zenodo Repository <https://doi.org/10.5281/zenodo.15311035>.

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Author contributions

J.S. – Proposing research topic, conceptualisation, methodology, data curation, investigation, writing – original draft, review & editing; Č.Š. – conceptualisation, methodology, data curation, investigation, writing – original draft; S.G. – methodology, investigation, writing – original draft; J.H. – methodology, investigation, writing – original draft; P.P. – investigation; D.K. – acquiring research funding; technical and material support, supervision.

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Declarations

Competing interests

The authors declare no competing interests.

Additional information

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Optimisation of preheating and interpass temperature in WADED magnesium alloy AZ61: A comparative study on microstructure, residual stresses, and internal defects

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Abstract: This study examines the effect of base material preheating on Wire Arc Direct Energy Deposition (WA-DED) of AZ61 magnesium alloy. The influence of preheating on the geometry and penetration behaviour of a single deposit was first evaluated and used to define suitable conditions for stabilising layer width in thin-walled structures. Two preheating strategies were examined, demonstrating that preheating the base material to 200 °C combined with the same interpass temperature provides the highest layer width stability and improved layer uniformity. The effect of preheating on residual stress was subsequently assessed through the analysis of base material deformation, revealing a reduction in residual stress of up to 50 % compared with fabrication without preheating. Furthermore, microstructural characterisation and μ CT analysis were performed to investigate porosity formation mechanisms in the deposited material. Preheating was found to have a positive effect on wall thickness stabilisation, residual stress reduction, and porosity mitigation. Based on the combined evaluation of geometric stability, residual stress, and internal defects, a temperature of 250 °C was identified as the most suitable preheating and interpass temperature for WA-DED of AZ61 magnesium alloy.

Keywords: Magnesium alloy, preheating, interpass temperature, WADED, WAAM, AZ61

Magnesium alloys, which exhibit low density and a high strength-to-weight ratio, are increasingly used in industries where the product's final weight is crucial. For example, reducing weight while replacing aluminium and steel automotive parts with magnesium can reach up to 50%¹. The use of such parts in the automotive and aerospace industries allows for a reduction in fuel consumption, resulting in more economical and environmentally friendly vehicle operation¹⁻⁴. Furthermore, the excellent heat dissipation performance and electromagnetic and radio frequency shielding properties of Mg alloys make them good candidates for commercial portable 3C products and military electronic devices^{1,5}. Together with its high recyclability, magnesium alloys, as stated in recent review articles^{2,3}, are considered the most promising green engineering materials in the 21st century. However, the production of magnesium parts is limited due to the hexagonal crystal structure and the limited number of easily activated slip modes. Therefore, magnesium alloys exhibit plastic deformation anisotropy, limiting the use of traditional manufacturing processes based on plastic deformation. Thus, 90% of magnesium-based parts are produced by casting^{6,7}. This represents the most economical manufacturing route for transforming magnesium alloys into shaped components. Still, it is also connected with numerous casting defects and the production of scrap metal^{6,8,9}.

The above-mentioned limitations of magnesium are the driving force behind the improvement of conventional techniques and the development of new manufacturing methods. One of the primary research directions suggested in the literature¹⁰ is the further improvement of advanced processing technologies for Mg alloys. Various Additive manufacturing (AM) techniques have been the subject of magnesium research in recent years due to their many advantages over conventional technologies, such as free-forming and complex shape manufacturing with excellent dimensional stability and high material efficiency^{4,10}.

One of the most explored AM techniques for metals is Laser powder bed fusion (LPBF). Although research on magnesium alloys shows promising results^{6,8,11,12}, LPBF is associated with high machine costs, size limitations, complicated process optimisation, and concerns regarding powder handling and storage, as well as safety issues. Moreover, magnesium alloys with low melting point temperatures are highly susceptible to evaporation due to localised heat and process complications resulting from melt pool instability^{13,14}.

Another AM technique highly preferable for the deposition of Mg alloys is Wire Arc Direct Energy Deposition (WADED), also known as Wire Arc Additive Manufacturing (WAAM). This process utilises a wire feeder system, an electronic arc heating source and an industrial robot. During the process, filler metal wire, shielded by inert gas, is heated through an electric arc and deposited as a bead on the base material. A controlled atmosphere chamber is not required, leading to more efficient use of shielding gas. WADED is widely used in the rapid prototyping of large parts due to its high deposition rate, high material utilisation rate, and low cost^{2,6,15,16}. Unlike the LPBF technique, WADED is not constrained by platform or build chamber size. However, arc-based additive manufacturing presents challenges, as several process parameters must be monitored, and optimisation is necessary¹⁵. That is crucial to addressing common WADED challenges, including insufficient or excessive penetration, residual stress, deformation caused by high heat input, reduced dimensional accuracy, formation of cracks and porosity^{16,17}.

Regions with increased porosity in the welds are formed mainly due to hydrogen, either as an impurity in the weld material and/or base metal or from the air humidity. The magnesium surface is covered by magnesium oxide when exposed to the atmosphere. If moisture is present, the oxide is converted to magnesium hydroxide¹⁸, which can be easily introduced to the melt pool during welding. The hydrogen solubility in molten magnesium is relatively high and decreases substantially during cooling, resulting in the formation of pores¹⁹. Such inhomogeneity can act as a stress concentrator, reducing the mechanical performance of the final product. Also, when internal stresses, caused by the high coefficient of thermal expansion of magnesium, are generated during the welding process, cold cracks can initiate from the pores in solidified weldments.

The formation of cold and hot cracks when welding magnesium alloys is a topic that has been widely discussed in connection with WADED technology. Various studies^{9,10,20,21} have shown that crack size and susceptibility to crack formation during welding increase with the magnitude of electric current and heat input. Additionally, a higher aluminium content in magnesium alloys can increase their susceptibility to cracking.

Various mechanisms can form hot welding cracks in magnesium alloys. During solidification, when part of the weld metal is already solidified, but regions with higher concentrations of alloying elements remain in the liquid state, internal stresses can result in cracking at the interface. Similarly, cracks are formed in the heat-affected zone at the interface between the solid solution and the liquified low-melting-point phases. Furthermore, Huang et al.²⁰ experimentally demonstrated that multiple thermal cycles, simulating multiple welding passes, increase internal stress and hot cracking susceptibility in AZ magnesium alloys. Such facts are important for the optimisation of the WADED process, because previously deposited layers are affected in the same manner as the heat-affected zone during the building process. The quality of interfaces between layers not only affects the final quality of multilayer components but also indicates issues in the building process.

Although the principles of welding crack formation can be applied, as they primarily depend on microstructure and the presence of internal stress, differences in residual stress distribution between WADED components and multi-pass welds have been experimentally demonstrated through numerical modelling and experimental verification²². Besides crack formation, internal stress also results in deformation of the final WADED parts^{17,23,24}, especially in thin walls. The magnitude of residual stress varies with the progress of the deposition^{17,22–24}, depending on the number of layers, dimensions of the base material, and welding parameters^{17,23,25}. For multilayer components, deposition pattern is also a significant influential factor and mitigation of residual stress can be achieved by use of deposition patterns with short track length^{22,24}. Thermomechanical model tested for thin walls of Inconel, Incoloy and stainless steel

showed, that preheat temperature, more specifically difference between solidus and preheat temperature, is far more influential for residual stress magnitude than base material thickness or welding parameters²³. On the contrary, experimental work on an aluminium alloy²⁵ showed the predominant effect of substrate thickness on residual stress distribution over interlayer temperature and the number of layers.

In addition to overall defects, thin walls produced by WADED are prone to non-stability in layer width, also known as necking. Such a defect appears in the first few layers before the balance between heat input and heat dissipation of the molten pool is established²⁶. Different layer widths have been observed in aluminium²⁷ and magnesium²⁶ alloy thin walls. Necking in AZ80 magnesium alloy was related to the temperature change of the base material, which heated up rapidly up to 150 °C during the fabrication of the first five layers. After that, the width of the layers remains stable²⁶. A similar phenomenon was observed for the temperature fields in a titanium alloy multilayer component²⁴. A steady state of temperature fields was achieved at the third layer, resulting in a stable stress distribution, but an increasing magnitude.

In theory, to maintain the welding parameters for optimum penetration in the WADED process, we can prevent necking, component deformation and cracking due to residual stress by minimising the temperature gradient, either by lowering the heat input or by preheating the substrate. It has been experimentally proven that increasing the interpass temperature from 50 to 100°C can reduce internal stresses by ~20% in aluminium alloy²⁵. Preheating to 350°C reduces residual stress in mild steel by 39% and 54%, depending on the direction relative to the deposition direction²⁸. Another approach to mitigating the thermal gradient is to minimise heat input during deposition by using progressive techniques or process modifications. In contrast to using a conventional power source, the Cold Metal Transfer (CMT) process, in which the wire is retracted after the short circuit, effectively mitigates the defects mentioned above due to its controlled heat input¹². This method reduces heat input by approximately 33% compared to conventional welding power sources and minimises spatter²⁹. Experimental work on aluminium alloy³⁰ also showed that the use of the CMT process leads to a smaller, colder, and faster cooling melt pool, which results in lower pore content.

Despite extensive research on WADED of aluminium and steel alloys, and isolated studies on magnesium alloys, the combined influence of substrate preheating and interpass temperature on microstructure evolution, residual stress development, and internal defect formation in AZ61 magnesium alloy processed by CMT-WADED has not yet been systematically investigated. In this study, the influence of preheating and interpass temperature on the microstructure, porosity, and residual stress of an AZ61 multilayer sample prepared by CMT-WADED was evaluated. As the first step of the deposition optimisation process, the influence of parameters and preheating on single-deposit samples with various preheating temperatures was tested. As the next step, width changes in thin walls were analysed, and the use of two preheating strategies as compensation was tested. Based on the obtained results, a process for multilayer thin-walled samples was designed, and a series of samples was fabricated with various preheating temperatures and analysed.

Materials and methods

Experimental setup

All welding experiments were conducted using a Cold Metal Transfer (CMT) power source, Fronius TPS 3200 CMT, paired with a KUKA KR 60 HA 6-axis robotic arm, controlled by Rhinoceros 6 software. During the tests, a constant speed of the welding torch (WS) of $10 \text{ mm} \cdot \text{s}^{-1}$ and a distance of 13 mm from the manufactured part were maintained. Pure argon (99.5%) was used as the shielding gas with a flow rate of $18 \text{ l} \cdot \text{min}^{-1}$. The AZ61 wire, with a diameter of 1.6 mm and the chemical composition given in Table 1. was used as the filler material. In the previous work³¹, the filler wire was manually cleaned with SiC paper. In the present study, a custom-built automatic rotary device was used to remove surface impurities. After mechanical cleaning, the wire was annealed at $380 \text{ }^{\circ}\text{C}$ for 11 minutes to convert residual magnesium hydroxide into magnesium oxide, which has less adverse effect on the magnesium welding process¹⁹. The base material, an AZ31 alloy plate, was mounted on a custom preheating table with overall dimensions of $450 \times 550 \text{ mm}$, an output of 3.2 kW, and a maximum preheating temperature of $400 \text{ }^{\circ}\text{C}$, with a precision of $\pm 1 \text{ }^{\circ}\text{C}$.

Table 1. Chemical composition of AZ61 magnesium alloy wire

Elements	Al	Zn	Mn	Si	Fe	Cu	Ni
Max (%)	6.50	1.50	0.40	0.10	0.005	0.05	0.005

Two sets of process parameters, listed in Table 2, were used: reference parameters (REF), established in previous work³¹ for AZ91 base material, and a second parameter set designed to achieve a deeper weld penetration (DWP) on the AZ31 base material used in this study. Welding parameters, such as I_{boost} , $t_{I_{\text{boost}}}$, and $I_{\text{sc_wait}}$, directly influence the geometry of the interface between the deposited material and the base material or a previously deposited layer. This relationship has been systematically investigated in previous study³¹. For adequate interlayer bonding, the underlying layer must be sufficiently remelted along the entire interlayer interface to avoid lack-of-fusion defects³². At the same time, excessive penetration and remelting should also be avoided, as excessive remelting increases local heat input, leading to higher thermal gradients and potentially promoting distortion and the accumulation of residual stress. Accordingly, welding parameters must be optimised in conjunction with the preheating temperature, which also influences the interlayer interface geometry and remelting behaviour.

Table 2. CMT parameters used for WADED

Parameters description	Abbreviation	I_{boost} (A)	$t_{I_{\text{boost}}}$ (ms)	$I_{\text{sc_wait}}$ (A)	$vd_{\text{sc_wait}}$ ($\text{m} \cdot \text{min}^{-1}$)	I_{sc2} (A)
Reference parameters (developed in prev. study)	REF	430 A	2,5 ms	30 A	$30 \text{ m} \cdot \text{min}^{-1}$	50 A
Deeper weld penetration parameters	DWP	400 A	3,5 ms	30 A	$30 \text{ m} \cdot \text{min}^{-1}$	50 A

Filler wire, cross-section specimens from single-deposit samples and thin walls were used for metallographic observation. Samples were prepared in the usual way, ground with SiC papers and polished with diamond pastes. Some samples were etched using an etchant based on acetic and nitric acid to reveal microstructural features. Optical microscopes, Olympus SZX7 and Zeiss Axio Observer Z1m, were used for observation and documentation. A scanning electron microscope with a field emission gun, Zeiss AG Ultra Plus, was used for more detailed observation and chemical analysis.

Experimental plan and workflow

In this work, the influence of preheating and interpass temperature on the microstructure, porosity, and residual stress of multilayer samples was investigated. Preheating temperatures in the range of 50 to 350 °C were selected based on values commonly applied during welding of AZ-type magnesium alloys¹⁸.

At first, the penetration depth and geometry of single-bead deposits produced at various preheating temperatures were investigated to establish suitable limits for subsequent multilayer fabrication. In the next step, changes in wall width with an increasing number of deposited layers were analysed in order to prevent necking in the lower part of the samples, and two preheating strategies were proposed and experimentally tested. Based on these results, a deposition process for thin-walled multilayer samples was designed, and a series of samples was fabricated under different preheating conditions and subsequently analysed.

Figure 1 illustrates the overall methodology of the present work in the form of a flowchart. Each experimental step is described in greater detail in the following sections.

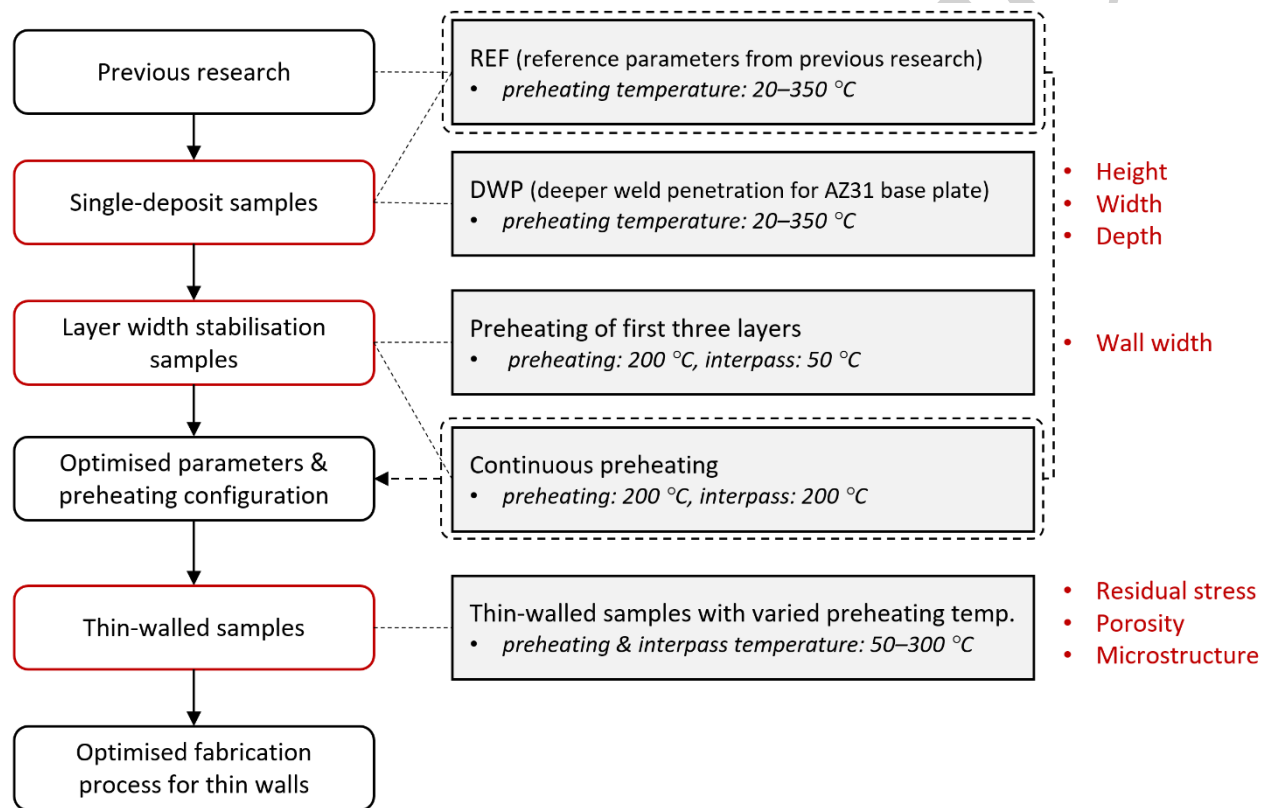


Fig. 1. Experimental methodology of the present work

Single-deposit samples

Two sets of eight single-deposit samples were fabricated with different preheating temperatures, namely 50°C, 100°C, 150°C, 200°C, 250°C, 300°C and 350°C, including a reference sample fabricated without preheating. Each single-deposit weld bead was 50 mm long. The first set was produced using the REF parameters; for the second set, DWP parameters were used.

After production, an optical 3D scanner (ATOS Triple Scan III) was used to measure all samples. Based on the scanned data, the height and width of the weld bead were evaluated using GOM Inspect 2018 software. Values were measured in six sections through the central part of the sample. For further examination of weld penetration depth, three cross-section samples from the central part of each single-deposit weld bead were cut and prepared for metallographic observation.

Layer width stabilization samples

Thin-walled samples consisting of 15 layers with a length of 50 mm and a layer height of 2.6 mm were fabricated without preheating in order to determine the width change with increasing number of layers. The width of the individual layers was measured using an optical 3D scanner (ATOS Triple Scan III) and evaluated at 15 levels with an offset of 2.6 mm.

Two strategies were proposed for preheating optimisation and tested at a preheating temperature of 200 °C. For the first strategy, the base material was preheated to 200 °C and the thin wall was fabricated with an interpass temperature of 200 °C (Strategy_200°C). For the second strategy, the base material was preheated to 200 °C, and three layers were fabricated with an interpass temperature of 200 °C. The rest of the sample was manufactured with an interpass temperature of 50 °C without preheating (Strategy_200/50°C). The number of layers fabricated with a higher interpass temperature was chosen based on the width changes observed in thin-walled parts fabricated without preheating.

Thin-walled samples

Based on the results from previous steps, six thin-walled samples consisting of 10 layers, with a length of 60 mm, and a layer height of 2.6 mm, were fabricated using REF parameters and preheating temperatures from 50 °C to 300 °C. Interpass temperatures were equal to the preheating temperature for the fabrication of the whole sample.

The thin-walled sample production configuration is inspired by general principles used in the Bridge Curvature Method, applied in LPBF, for the rapid assessment of the influence of process parameters on residual stress³³. The base material was mounted to the heated table on only one side of the sample, as depicted in Fig. 2, and scanned using a handheld optical scanner (Scantech SIMSCAN 22). After the fabrication, the thin-walled sample was allowed to cool down to room temperature and scanned again. Both scans were aligned and measured using GOM Inspect 2018 software to assess the deflection in the direction normal to the preheating table surface, which occurs due to residual stress resulting from a large temperature gradient within the material. Measured plastic deformation was then used for comparative analysis of the residual stress in the samples produced at different preheat and interpass temperatures.

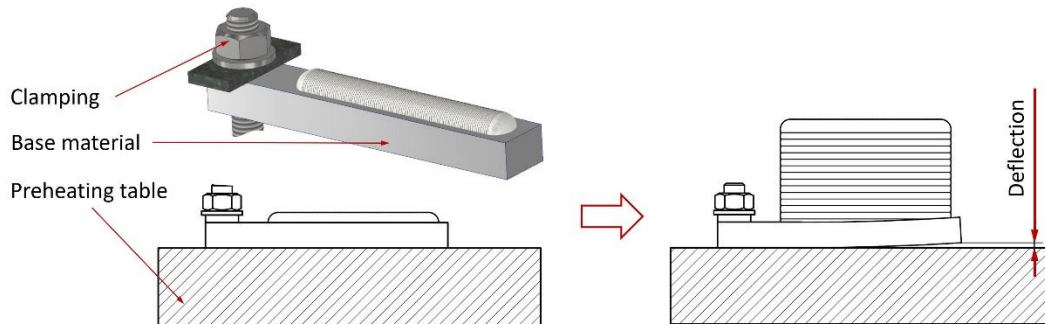


Fig. 2. Thin-walled specimen used for analyzing residual stress-induced deflection

All produced samples were inspected using a μ CT system (Procon CT-Alpha) to analyse the relative material density. Specimens were scanned at a 45 μ m isometric voxel resolution using a cone-beam μ CT device. The voltage was set to 230 kV while the applied current was 200 μ A. A copper filter was used to minimise beam hardening effects. The pore distribution, shape, and size were evaluated using the VGStudioMax 2023.1, using an ISO50 threshold and a minimum of eight voxels for a segmented pore.

For further examination, cross-section samples approximately 20 mm from the end of each sample were cut and prepared for metallographic observation. Samples were used for microstructural analysis and grain size evaluation according to ASTM E112-13, utilising the line intercept method and ImageJ image analysis software.

Results and discussion

Geometry of the single-track weld bead

The measurement results regarding the influence of base material preheating temperature on the geometry of a single-track weld bead indicate that increasing the preheating temperature leads to an increase in bead width and penetration depth, while reducing the weld bead height. The results and selected cross-sections of the weld bead preheated to a temperature of 250°C are shown in Fig. 3.

The weld bead height decreased from 4.08 mm to 2.75 mm (32.6%) for REF parameters and from 4.25 mm to 4.16 mm (2.1%) for the DWP parameters as can be seen in Fig. 3a. The results show that the DWP parameters with lower I_{boost} current and its longer duration result in a slower downward trend and significantly reduce the influence of preheating on decreasing weld bead height.

The effect of preheating on the weld width is very significant for both welding parameter sets. The weld width of the REF samples increased from 7.23 mm to 11.3 mm (56.4%) and from 9.18 mm to 12.19 mm (32.8%) for DWP samples. Both lines in Fig. 3b. show a similar growth trend and similar standard deviations, indicating comparable weld width stability.

The samples produced with the REF parameter settings did not achieve a sufficient penetration depth until they were preheated to temperatures over 100°C (Fig. 3c). Up to this temperature, weld penetration occurred only locally along the length of the weld bead, and cohesion with the base material was minimal. Stable penetration depth was observed from a preheating temperature of 150°C to 350°C, where the penetration depth increased from 0.3 mm to 1.15 mm. Additionally, samples preheated to a temperature below 200 °C exhibited a higher porosity concentration at the interface between the weld bead and the base material, which was likely caused by surface contaminants or corrosion residues⁷.

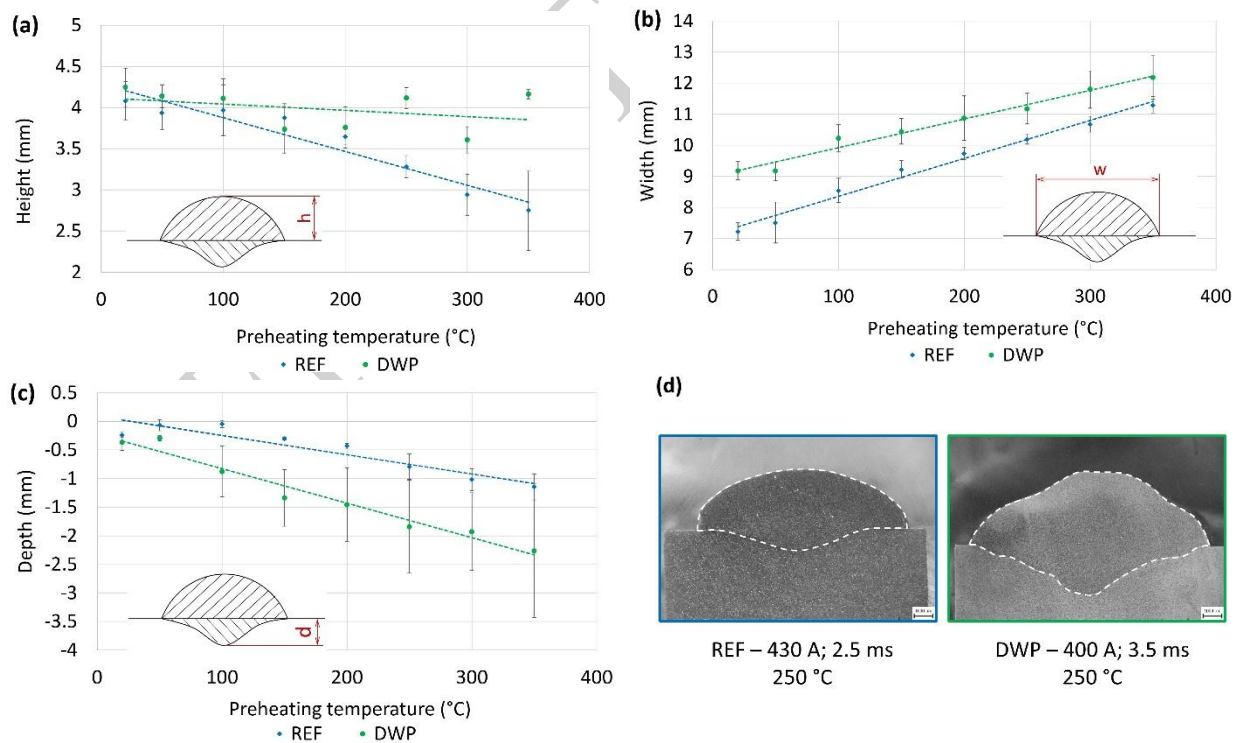


Fig. 3. (a) High (b) width (c) penetration depth of single-track weld bead (d) cross-section of selected samples

For the DWP samples, the parameter change resulted in a significant shift in the penetration depth. In the sample without preheating (20°C), the penetration depth was 0.37 mm, and gradually increased to 2.26 mm for a preheat of 350°C. The weld penetration depth growth trends in Fig. 3c are similar for REF and DWP parameter sets. However, a noticeably higher standard deviation of penetration depth is observed for the DWP parameters, indicating reduced penetration stability along the weld length. At a preheating temperature of 250 °C, both REF and DWP parameter sets result in continuous penetration along the entire interface between the deposited material and the base material, as shown in Fig. 3d.

Layer width stabilisation

This part of the optimisation aimed to eliminate wall necking in the first layers of thin wall caused by enhanced heat sinking through the base plate within the first few layers²⁷. Such a phenomenon can be significant because magnesium alloys have good thermal conductivity; therefore, a longer period is required to establish a balance between heat input and heat dissipation of the molten pool²⁶. However, as the necking is caused by a temperature gradient between the substrate and the molten pool, it can be compensated or even eliminated by preheating the substrate. In the first phase, two thin-walled samples with REF and DWP parameters were produced and scanned (see Fig. 4).

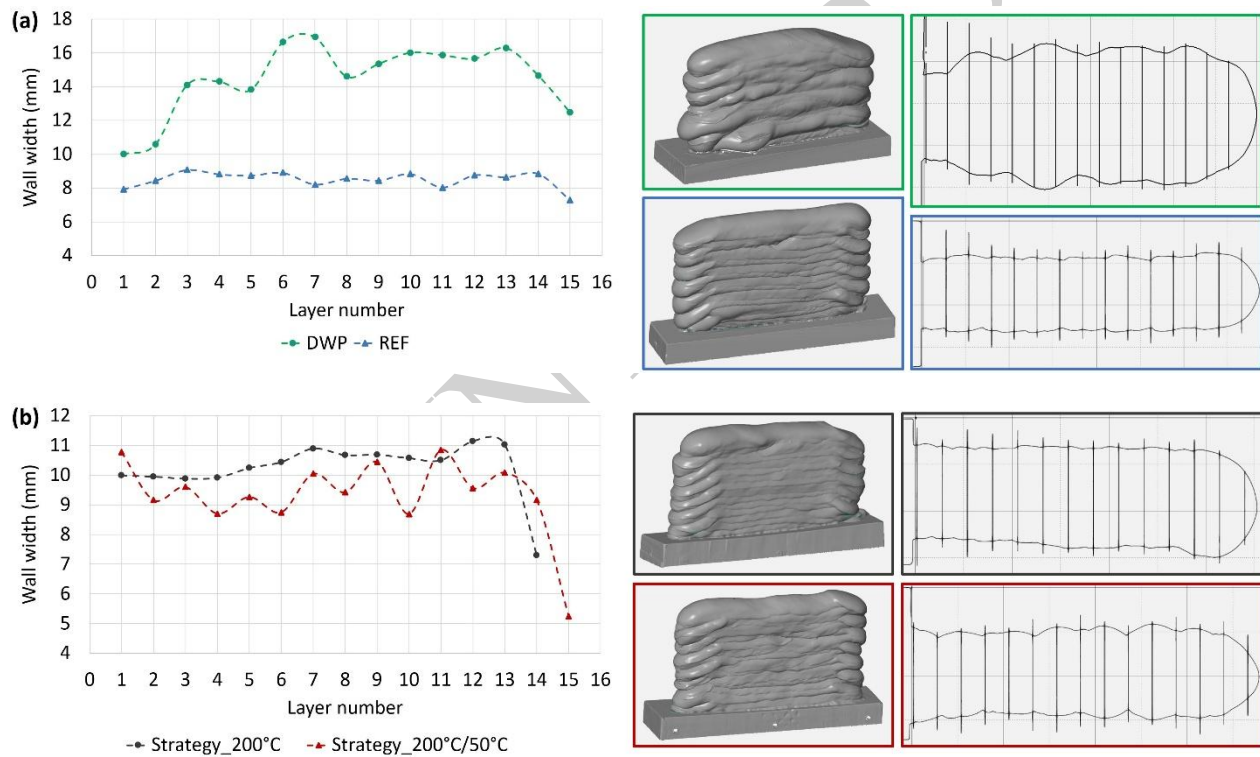


Fig. 4. Preheating of thin walls to eliminate necking, a) REF and DWP parameters without preheating, b) REF parameters with preheating at 200°C and 200/50°C

The width of the first layer of a thin wall produced with DWP parameters corresponds to 10 mm, the average wall width is 15.5 mm, so the percentage increase is 54.4%. This change in width is significant and cannot be compensated by the preheating alone. In addition, the wall produced with DWP parameters exhibits significantly unstable behaviour across the height of the sample. For this reason, the DWP parameters were assessed to be unsuitable for producing multi-layer thin walls. They were not used to produce further samples.

The width of the first layer of a thin wall produced without preheating with REF parameters corresponds to 7.9 mm. The average wall width is 8.6 mm, resulting in an 8.7% increase. The observed necking can be compensated by preheating. According to the single-deposit experiment, the first layer should reach the required width already at a preheating temperature of 150 °C. However, due to insufficient penetration during preheating at 150 °C while using REF parameters, a minimum suitable preheating temperature of 200 °C was selected for optimisation. This temperature should ensure the minimum required penetration along the entire interface between the deposited material and the base material and, at the same time, sufficient width of the first layer.

Subsequently, two samples were produced using the Strategy_200°C and Strategy_200/50 °C approaches. Stable layer width with no significant difference between the first and remaining layers was observed for the sample produced by Strategy_200°C (see Fig. 4b). The width of the first layer was 10 mm, the average layer width was 10.3 mm, and the percentage increase was only 3.2%. At the same time, a reduction in the height of the individual layers was observed, resulting in a decrease in the overall height of the sample, which was 34.6 mm. For comparison, the height of the sample produced with REF parameters without preheating was 39.5 mm. When using Strategy_200/50°C, a slight decrease in layer width was observed after the third layer, from 10.8 mm on the first layer to 8.7 mm, with an average layer width of 9.5 mm, corresponding to a 17.5% reduction in width. Overall, this strategy led to poorer layer width stability across the sample height. On the other hand, there was no significant reduction in layer height, and the final sample measured 38.3 mm in height. It was 3.6 mm higher than the sample produced by Strategy_200°C and 1.2 mm lower than the sample with REF parameters without preheating.

While a preheating and interpass temperature of 200°C stabilises the layer width and prevents necking at the bottom, it also reduces the layer height. This should be considered when planning trajectories for producing taller, thin-walled objects, mainly to maintain a constant distance between the contact tip and the fabricated thin wall.

Thin-walled samples

Residual stress was measured indirectly as the amount of plastic deformation of the base material after deposition. The base material was fixed to the preheating table at one end, and its deflection in the direction normal to the preheating table surface, was measured. Deflection profiles along the edge of individual samples caused by plastic deformation are shown in Fig. 5b.

As can be seen from the preheating temperature–deflection dependency in Fig. 5a, up to a temperature of approximately 150 °C, preheating does not affect the amount of residual stress remaining in the thin-walled sample. In this temperature range, the maximum measured deflections were approximately 1.27 mm. In the range from 150 °C to 300 °C, a significant reduction in deformation and thus in the amount of residual stress in the body was observed. At a preheating temperature of 200 °C, the deflection decreased to 1.13 mm, and at 250 °C, it further decreased to 0.77 mm. At the maximum tested preheating temperature of 300 °C, a deflection of 0.63 mm was measured, indicating a reduction in deformation of up to 49.6%. From the shape of the curve interpolated by the measured points, it can be assumed that the results asymptotically approach a value outside the investigated preheating temperature range. Higher preheating temperatures would likely not result in further significant reductions in deformation and thus residual stress.

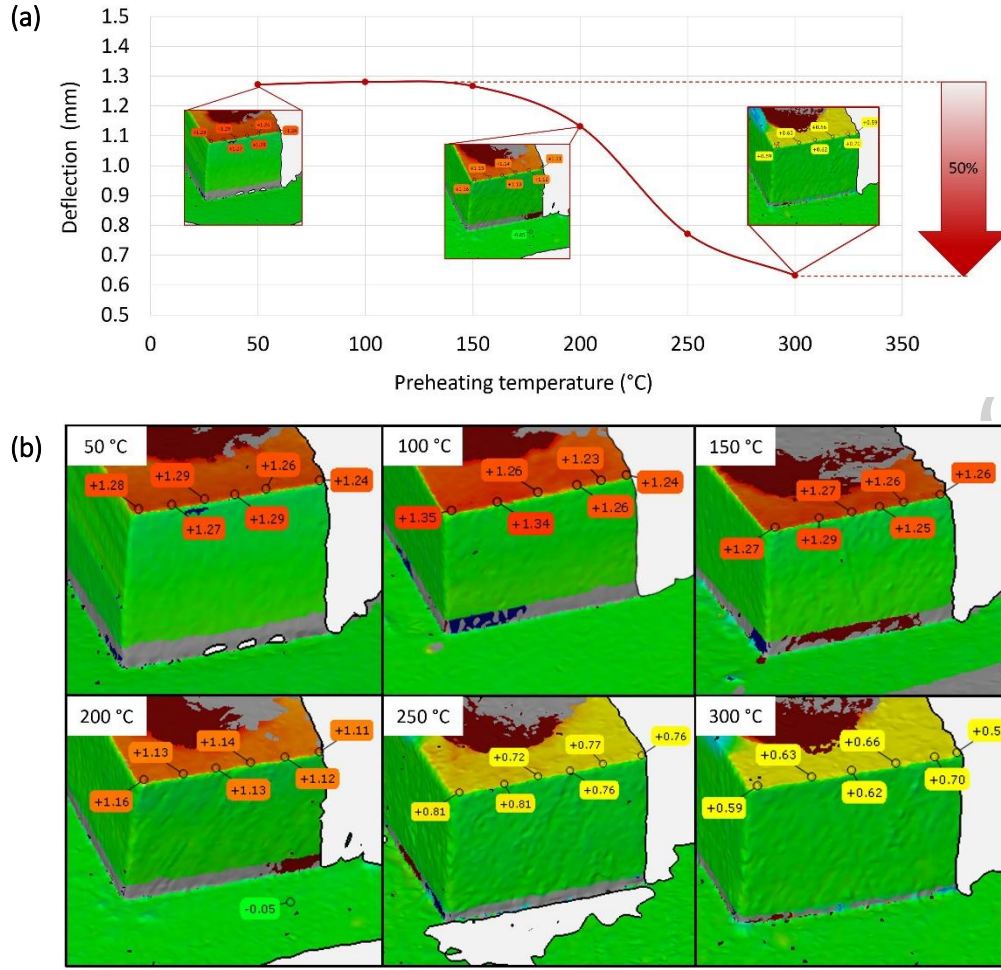


Fig. 5. Effect of preheating on the residual stress: (a) preheating temperature-deformation dependency, (b) plastic deformation in samples produced with different preheating temperatures

The μ CT images of individual samples with different preheat temperatures are shown in Fig. 6, along with the overall porosity ratios. The highest porosity level, 2.1%, was observed in the sample preheated to 50 °C and the lowest porosity of 0.01% was measured in the sample preheated to 250 °C. In general, overall porosity decreased with increasing preheating and interpass temperature.

As can be seen from the images, all samples contain small pores, primarily in the interlayer regions, as well as rather sizeable columnar pores. A detailed pore size analysis revealed that the majority of pores exhibit very small volumes. Most detected pores have a volume of up to 0.0007 mm³, and more than 75% of all pores in all measured samples are smaller than 0.01 mm³. The pore size distribution and the total number of pores vary with the preheating temperature; however, we can clearly see that the interlayer porosity diminishes with increasing preheat temperature, showing very low porosity ratios for samples fabricated at 250 °C and 300 °C.

In contrast, samples preheated to lower temperatures exhibit large columnar pores extending across multiple deposited layers, with individual pore volumes reaching up to 48 mm³. To the best of our knowledge, such internal defects have not been previously reported in the literature^{2,6,8,12,15,19,30}. Further material analysis is presented in the following sections. Interestingly, the formation of columnar pores does not occur when higher preheating and interpass temperatures are used.

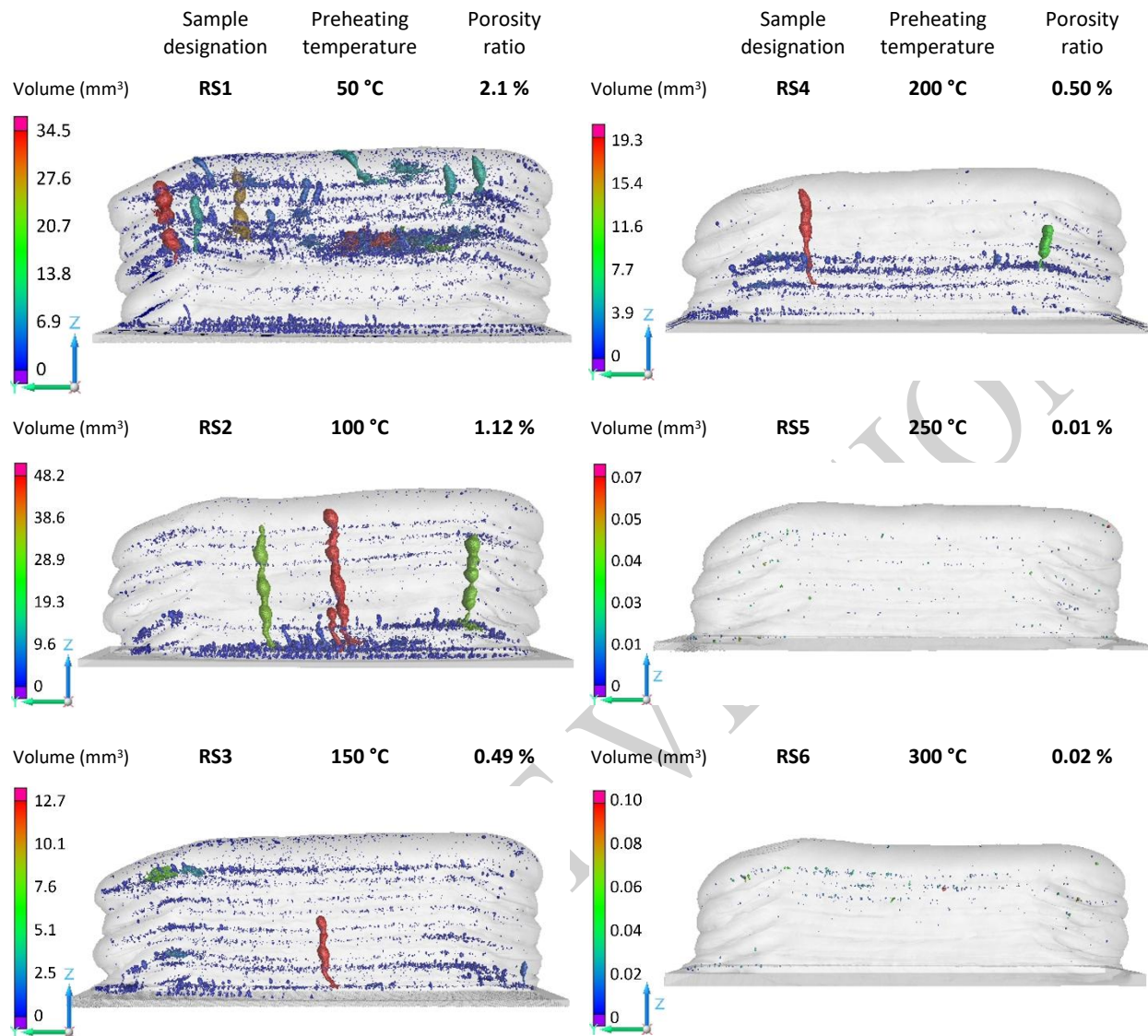


Fig. 6. Porosity measurement using μ CT scanning

All fabricated samples exhibited a layered macrostructure with a distinct interface containing coarse grains (CG) followed by very fine grains (FG). The macro- and micro-morphology of samples fabricated with preheat temperatures of 250 °C and 100 °C are shown in Fig. 7. These samples were chosen specifically because the preheat temperature of 250 °C is the lowest temperature that results in the elimination of columnar pores and low interlayer porosity. On the other hand, the sample fabricated at a preheat temperature of 100 °C contained interlayer porosity, several columnar pores, and sizable globular pores. The sample fabricated with a higher preheat temperature was fairly symmetrical; therefore, only half of the image is presented. On the contrary, the sample fabricated at a lower temperature was asymmetrical; the presence of columnar and large globular pores resulted in variations in layer height and width.

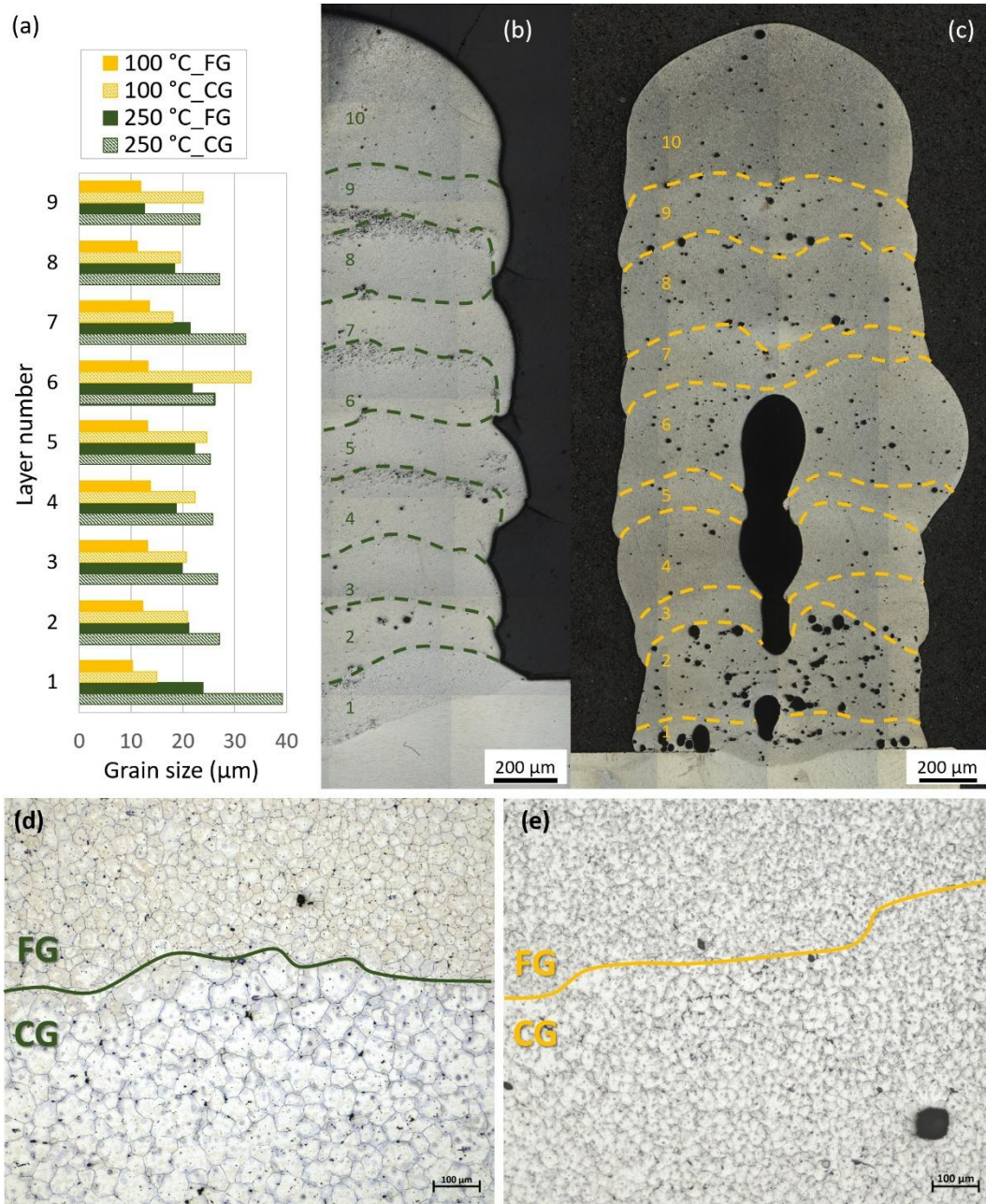


Fig. 7. Thin wall samples fabricated with preheating: (a) grain size in fine grain (FG) and coarse (CG) areas in individual layers; overall macro images of samples fabricated with preheat temperature of (b) 250 °C and (c) 100 °C; microstructures on the 1-2 layer interface of sample fabricated with preheat temperature of 250 °C and (e) 5-6 layer interface of sample fabricated with preheat temperature of 100 °C

Microstructural observations of all samples fabricated with different preheat temperatures revealed similar microstructures, including solid solution, Al-Mn-based particles, and the $Mg_{17}Al_{12}$ phase, with significantly higher amounts precipitated in the top part of the samples. Depending on the preheat temperature, an increase in the quantity of the $Mg_{17}Al_{12}$ phase started between layers 5 and 7. Similarly, differences in grain sizes were observed in individual layers, as seen from the grain size–layer number dependence in Fig. 7a. Each layer starts with a fine grain region (FG) and ends with a coarse grain region (CG). In a sample fabricated at a preheat temperature of 100 °C, the FG size varies between 10.24 and 13.71 μm, with stable values in the middle part of the sample and slightly smaller grains in two layers in the top and two layers in

the bottom of the sample. The CG size, on the other hand, was very small in the first layer (15.01 μm) and gradually increased up to the sixth layer (33.19 μm). It should be noted that the size difference between the fifth and sixth layers was 26%, and the difference between the sixth and seventh layers was 46%. The finer grains in the first layer can be attributed to the effective cooling achieved through heat transfer to the preheated base plate, which was preheated to 100 $^{\circ}\text{C}$. In comparison, the maximum temperature measured on the sample was ~ 350 $^{\circ}\text{C}$.

In the sample fabricated at a preheat temperature of 250 $^{\circ}\text{C}$, the tenth layer contained the finest grains. In contrast, the first layer contained grains of the largest size, in both the CG and FG regions. The middle part of the sample was formed by grains with sizes oscillating around 20.5 μm and 27 μm for the FG and CG regions, respectively. High values in the first layer were caused by less effective heat transfer to the base plate, which was preheated to a relatively high temperature. Additionally, a relatively high interpass temperature resulted in a higher maximum temperature measured on the sample, namely ~ 470 $^{\circ}\text{C}$. Fine grains in the top resulted from more effective cooling, which was least affected by the rest of the sample.

Grain sizes in all fabricated samples measured on the 3–4 (bottom) and the 8–9 (top) layer interfaces are shown in temperature dependence in Fig. 8. These specific layers were chosen as the grain size should not be affected by the position on the sample, i.e. by the peak values in the first layers or 5–6–7 layer interfaces. As can be seen from the results, grain size coarsening follows the same trend in the top and bottom areas. Most results comparing both areas show similar results with differences within the standard deviation, except for the peak value in the bottom CG area of the sample fabricated at 300 $^{\circ}\text{C}$ preheat and interpass temperature.

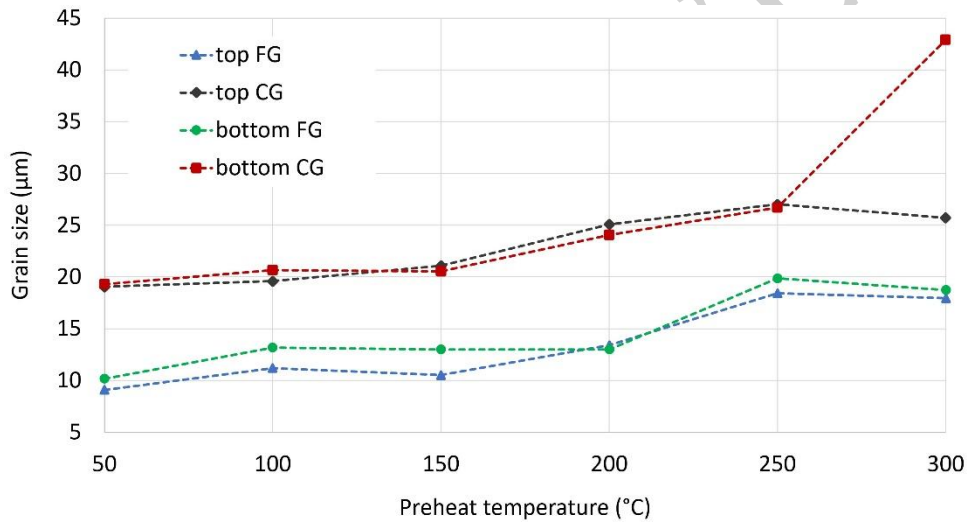


Fig. 8. Average grain size in the fine grain and the coarse grain regions in samples fabricated with different preheat temperatures

Grain size in the first layers and in the top and bottom areas was clearly affected by the preheat temperature. When preheat and interpass temperatures were low enough to allow sufficient cooling via heat transfer to the base plate, the overall temperature of the sample was lower. The maximum temperature measured on the sample without preheating was ~ 260 $^{\circ}\text{C}$, and every 50 $^{\circ}\text{C}$ increment in preheat and interpass temperature caused an increase in maximum sample temperature by 40 – 50 $^{\circ}\text{C}$. As was reported by Chen et al.³⁴, AZ61 alloy exhibits an insignificant increase in grain size at annealing temperatures up to 350 $^{\circ}\text{C}$, even after a 25-hour dwell time. Higher annealing temperatures of 400 $^{\circ}\text{C}$ and 450 $^{\circ}\text{C}$ result in a significant increase in grain size. Therefore, we can expect that when the maximum sample temperature exceeds 400 $^{\circ}\text{C}$, which corresponds to preheat and interpass temperatures of 200 $^{\circ}\text{C}$, the grain size should increase slightly throughout the whole sample, even after 30 minutes, which is the maximum fabrication time for the presented thin-walled samples. A more significant increase in grain size is expected in samples with a maximum temperature over 450 $^{\circ}\text{C}$, i.e., a 300 $^{\circ}\text{C}$ preheat and interpass temperature. In addition, with a

higher preheat temperature, cooling is slower, and the sample is held over the critical temperature for a longer period.

The appearance and number of pores corresponded to observations using μ CT. Interlayer porosity, the number of small pores on the interface between individual layers, can be clearly visible in macrographs and micrographs in Fig. 7. The interlayer porosity is most likely caused by impurities present in the filler material. Magnesium is a highly reactive metal that readily reacts with atmospheric moisture, forming hydroxides on the wire surface. These hydroxides decompose when introduced into the melt pool, releasing hydrogen gas¹⁹. Due to the rapid solidification rate of magnesium, the gas does not have sufficient time to escape the molten pool, and surface tension traps it within the material, resulting in the formation of pores. When a higher preheat and interpass temperature is used, the overall temperature of the sample increases, allowing for slower cooling and sufficient time for gas to escape before the newly deposited layer solidifies.

A similar mechanism would explain the shape of columnar pores (see Fig. 9). Firstly, a large pore is formed during the solidification of the deposited layer. During the deposition of subsequent layers, localised remelting can occur around the trapped gas, allowing it to migrate upward and result in vertical porosity channels tilted in the deposition direction in every layer. These columnar pores were not observed in samples preheated to temperatures of 250 °C and higher, most probably due to the high overall temperature of the sample and slower solidification. On the other hand, no influence of lower preheat and interpass temperature can be observed, and the presence of columnar pores in the samples seems to be very random. This fact, along with the nonuniform distribution of interlayer porosity, suggests that even though porosity can be minimised by technological means, it is not caused by process parameters or errors in deposition.

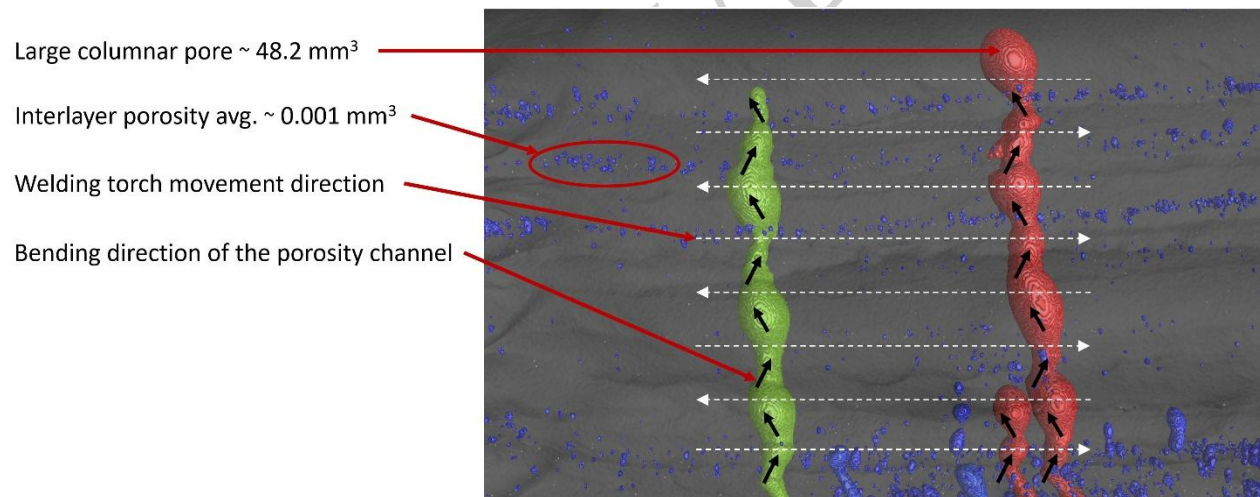


Fig. 9. Columnar porosity formation in a WADED thin-walled sample fabricated with preheat temperature of 100 °C

A similar hypothesis, called the penetration effect, was used to explain the removal of pores from the upper part of the previous layer due to remelting by Derekar et al.³⁰, who tested the influence of interpass temperatures 50 and 100°C (with the same preheat temperature for initial layers) in Al-Mg-Mn alloy thin walls fabricated by CMT in comparison with the pulsed-MIG method. Basically, the penetration effect hypothesises that the previous layer is affected only to the extent equal to the penetration depth and shape of a single deposit. In other words, a shallow shape and a small penetration depth of the CMT deposit do not influence the interface between the two previous layers and their interlayer porosity. Samples produced with a higher interpass temperature by CMT showed a slightly higher level of porosity (0.031 vs. 0.041%) but a less uniform distribution of pores, compared to the lower temperature of 50 °C. Based on our data obtained at higher preheat temperatures, we can hypothesise that the result of Derekar et al. suggests that the temperature of 100°C also affects pores in the lower part of the previous layer to some extent, but is not high enough to allow escape of gas. This is also, to a certain extent, hypothesised by the authors and

suggested as a topic for further research. Even though the parameters of their process differ from those of our experiment, comparison is possible because they used an alloy with similar solidus and liquidus temperatures¹⁸, good thermal conductivity, and based their penetration effect hypothesis primarily on the shape and depth of penetration. Derekar et al achieved penetration ~ 1.55 mm, approximately 60% of the layer height, and our welds exhibited the same shape and penetration, only $\sim 10\%$ at a preheat temperature of 100°C and yet migration of trapped gas through the layers, resulting in the formation of columnar pores, was clearly possible.

Furthermore, a decrease of over 50% in porosity was achieved compared to a preheating temperature of 50°C . As the significant decrease in overall porosity and complete mitigation of columnar pores in our experiment were achieved at a penetration depth of 23% (preheat temperature of 250°C), we can assume that penetration depth is not the key factor in porosity reduction by remelting of successive layers. Derekar et al.³⁰ also compared results from pulsed-MIG and CMT. They hypothesised that the relatively low heat input from CMT cannot raise the temperature high enough to allow pores to escape by hydrogen diffusion. However, as the hydrogen diffusivity in our experimental material, magnesium, is very low compared to aluminium³⁵, including also their solid solutions (Mg) and (Al), this hypothesis is also not consistent with our results.

More detailed observations and chemical analysis were performed to explain the origin of columnar pores. As can be seen from the secondary electron micrographs in Fig. 10, the surface of the columnar pore contains a high number of particles with sizes ranging from 1.3 to $3.6\ \mu\text{m}$. Chemical analysis (Fig. 10d) confirmed the presence of aluminium and manganese in particles with no traces of iron. Even though the energy of Mn K β and Fe K α are very close, the absence of signal at $7.058\ \text{keV}$, the energy of Fe K β , proves that the last peak in the spectrum belongs to manganese.

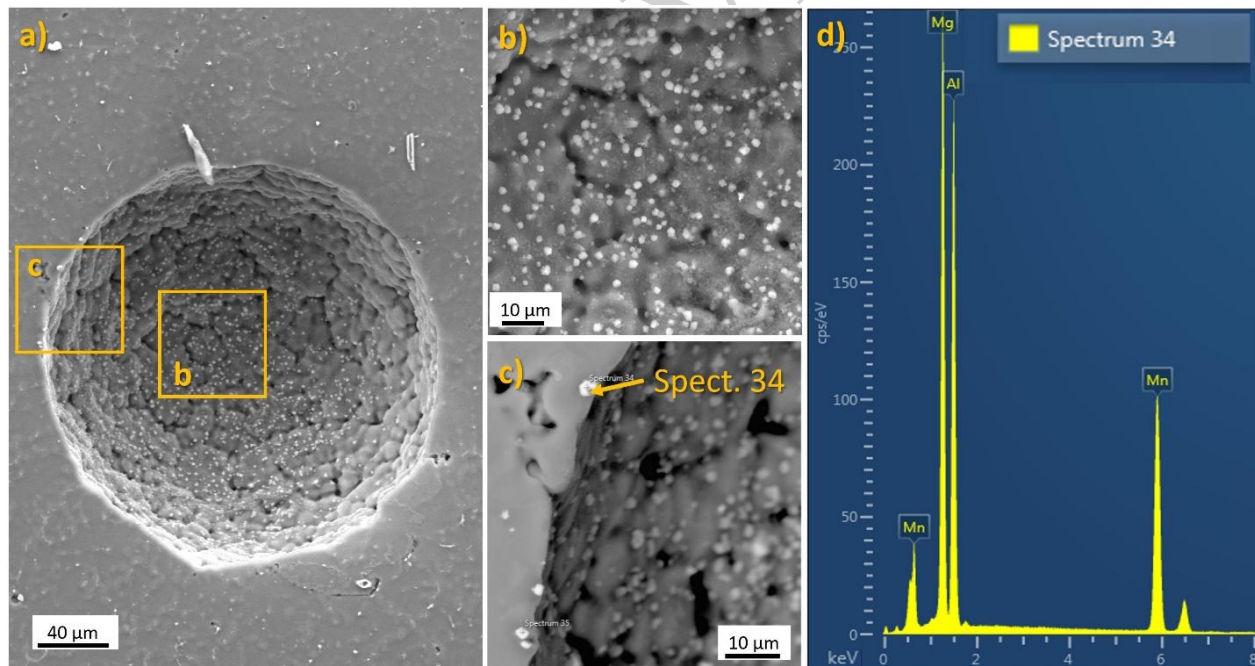


Fig. 10. (a) SEM image of columnar pore in sample fabricated with preheat and interpass temperature of 100°C , (b), (c) detailed images of particles, (d) energy dispersive X-ray spectrum of particle marked in image (c)

Further surface analysis along the columnar pore was performed to explain the presence of high amounts of secondary particles on the pore surface. Besides Al-Mn particles, higher amount of oxygen and locally traces of calcium were observed, as can be seen from image of representative area in Fig. 11. Elemental maps of Mg and Zn are not presented as the signal was detected all over the mapped area on the sample with no visible deviations, except for Al-Mn particle in central part, where less signal for Mg K α was detected.

Higher amount of Al-Mn particles could be explained by oxide layer on the pore surface. A similar phenomenon is documented in AZ alloys castings^{36–38}, where clusters of Al-Mn-based particles are attached to entrained oxide films. Particles in molten metal are either entrapped by entrained oxide or nucleate on entrained oxide and grow during cooling. During the WADED process, when a layer is deposited and part of the previous layer with a sizeable pore is remelted, the oxide film from the pore surface can act as entrained oxide and result in entrapment and/or nucleation of Al-Mn particles in the melt pool.

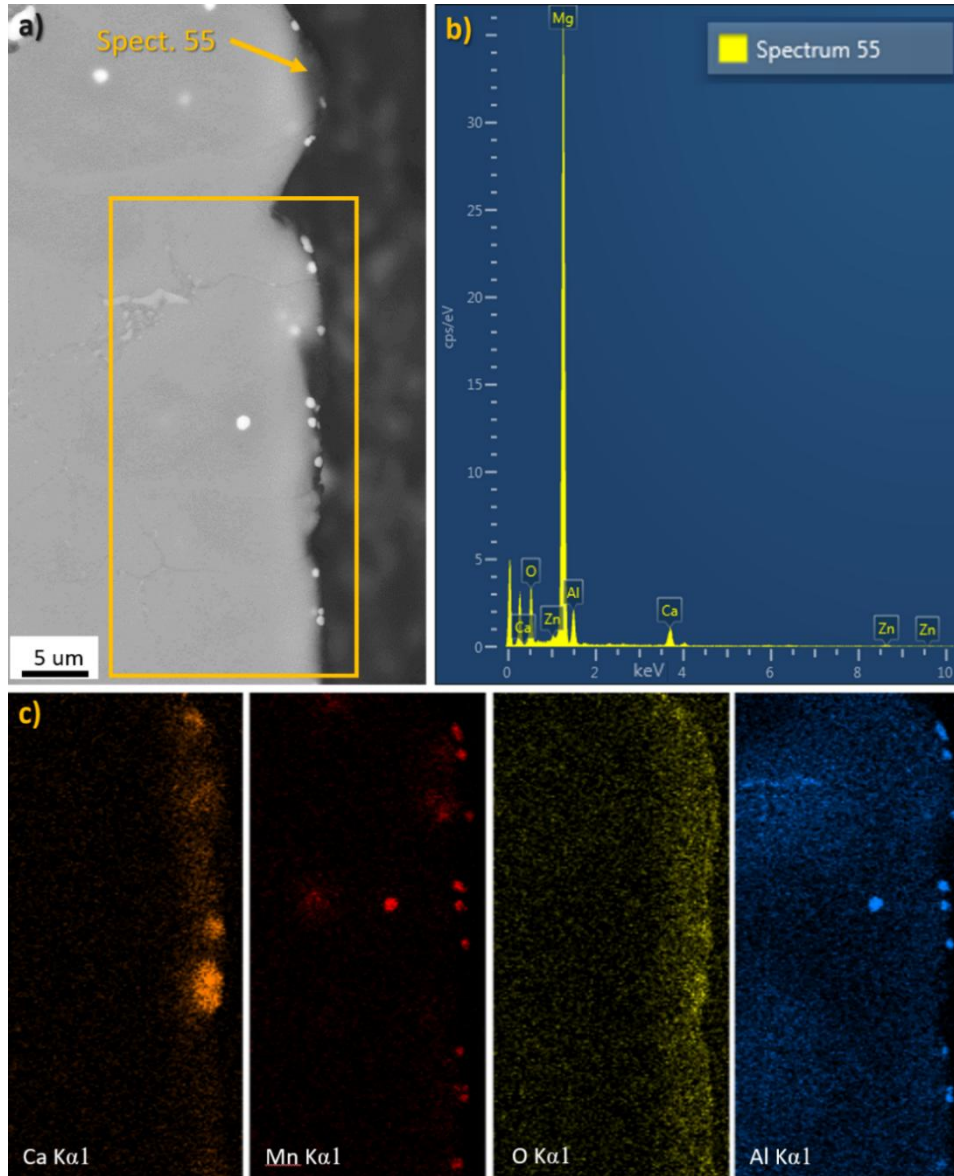


Fig. 11. (a) SEM image of the side of columnar pore with (b) energy dispersive X-ray spectrum of spot marked in image and (c) elemental mapping of marked area

Traces of Ca were observed only on the surface of the columnar pores, suggesting a deposition process as its origin, with three possible sources of contamination: shielding gas, base material and filler material. Shielding gas was supplied by a verified supplier, and no columnar pores or other contaminations were observed in other experiments while using the same batch of shielding gas. Since the columnar pores start in different layers in the thin-walled samples, we can also exclude the influence of the base material.

To confirm the origin of Ca on the surface of the pore, filler wires were analysed (see Fig. 12). As the wire after pre-treatment did not show any traces of Ca, thick oxide layer or other contaminants, analysis of the wire without pre-treatment is also presented. Microstructural components in filler wire were consistent with the microstructural analysis of thin wall deposits, but analysis of wire surface revealed a higher amount of oxygen and locally traces of Calcium (see Fig. 12c). Calcium, as part of the surface oxide layer, most probably comes from lubricants used in wire drawing processes, in which calcium stearate is very widely used³⁹ and entered the deposition process as a result of an insufficiently efficient pre-treatment of filler wire.

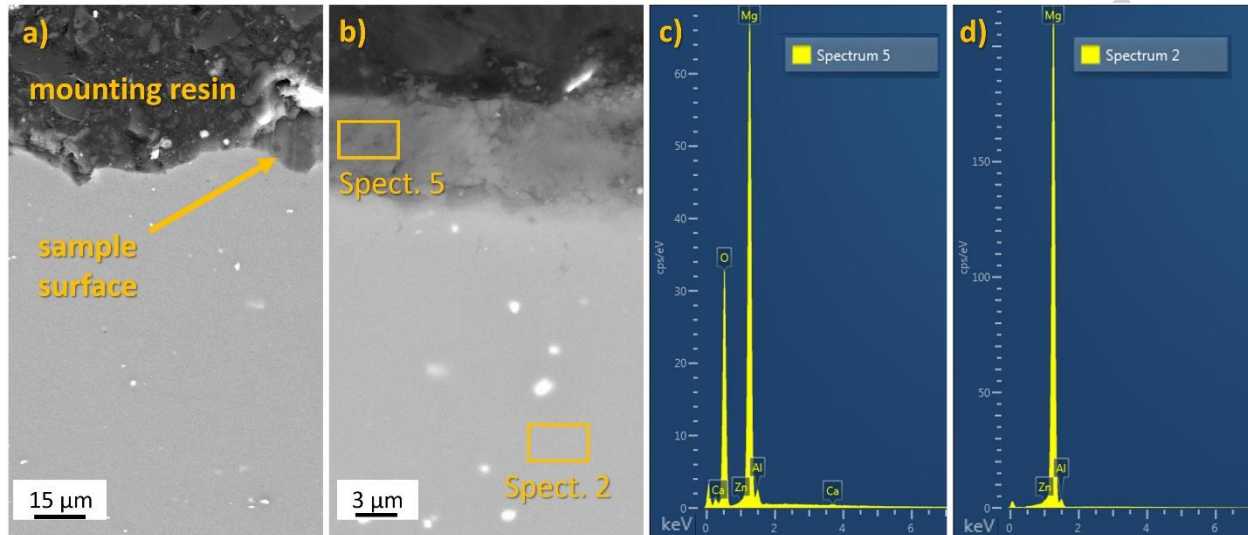


Fig. 12. SEM images of the wire (a) after pre-treatment and (b) before pre-treatment, (c) (d) energy dispersive X-ray spectra of areas marked in the image

The high reactivity of magnesium causes an oxide layer to form on the wire surface. As expected, pre-treatment of the filler wire before deposition is mandatory. In our previous experiments, wires were prepared manually with SiC papers, which guaranteed a clean surface of the wire, and no columnar pores were observed in the produced samples. This study employed an approach that is more suitable for industrial applications. In the first step, the wire was mechanically cleaned by a custom-built automatic rotary device and then annealed to convert any remaining hydroxides into oxides. Although the analysis of the wire after pre-treatment did not reveal any oxides or contaminants, only a few randomly chosen pieces of wire were analysed. Further analysis of filler wire and optimisation of the process are necessary to ensure that no impurities, such as Ca, can enter the deposition process.

Although the automated pre-treatment process was not as effective as the manual process and resulted in severe material defects in thin-wall samples fabricated with lower preheat and interpass temperatures, the experiment demonstrated that preheating can successfully reduce interlayer porosity. Results also suggest that preheating temperatures of 250°C and above can prevent the formation of columnar pores; however, further examination is needed to prove this hypothesis.

Despite the higher production cost associated with preheating, it significantly reduces the residual stress level and has a positive impact on thin-walled geometry. A more effective automated pre-treatment process, incorporating sufficient preheating and interpass temperature, is a promising method for producing thin-walled samples from AZ magnesium alloys by WADED.

Conclusion

This study confirmed that preheating and interpass temperature control are key to improving the quality of AZ61 magnesium alloy parts produced using the WADED method. Based on the findings, the following conclusions can be drawn:

- Preheating to 200 °C effectively stabilised layer width, eliminated necking in the initial layers, and ensured consistent part geometry throughout the build.
- Increasing the preheating temperature reduced deflection of the base material, corresponding to a residual stress decrease. The magnitude of deflection within thin-walled parts can be reduced by up to 50% using base material preheating.
- The amount and morphology of pores were strongly affected by preheating. While low preheating temperatures led to a high level of interlayer porosity and the presence of columnar pores, these defects were almost completely suppressed at a preheating temperature of 250 °C.
- The identified optimum range of 200–250 °C represents a balance between layer stability, stress reduction, and defect minimisation, making it suitable for practical application, depending on the specific requirements of the product properties.

Although this study was limited to relatively simple thin-walled specimens, the observations form a valuable basis for extending the approach to more complex shapes. The results can contribute to streamlining manufacturing workflows and supporting more sustainable production of magnesium alloys. A logical continuation of this work will be to investigate the mechanical properties and verify the identified temperature window on geometries of higher structural complexity.

Author contributions

J.S. – Proposing research topic, conceptualization, methodology, data curation, investigation, writing – original draft, review & editing; S.Z. – methodology, investigation, writing – original draft; M.H. – methodology, investigation, writing – original draft; S.S. – technical and material support, methodology, investigation; D.K. – acquiring research funding; technical and material support, supervision.

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Declaration of competing interests

The authors declare no conflict of interest.

Data availability statement

The data that support the findings of this study are openly available in Zenodo Repository at <https://doi.org/10.5281/zenodo.16881884>.

The original version of the author’s preprint of this study is openly available in the Zenodo Repository at <https://doi.org/10.5281/zenodo.17005427>.

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Short Communication

Influence of deposition strategy on the microstructure of volumetric WADED-fabricated AZ61 magnesium alloy components

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ABSTRACT

This study investigates the impact of various deposition strategies on the microstructure of volumetric components fabricated from AZ61 magnesium alloy using Wire Arc Direct Energy Deposition (WADED), also referred to as Wire Arc Additive Manufacturing (WAAM). Because coarse grains can significantly reduce mechanical performance, thermal simulations were performed to identify low-cooling-rate regions associated with different deposition trajectories. Three deposition strategies were analysed: ZigZag, Spiral, and S-pattern. The simulated thermal profiles were validated by optical grain size analysis and hardness measurements. In the selected critical regions, where the largest grains are expected, the S-pattern achieved the narrowest range in grain size and hardness distribution. This trajectory also exhibited the finest microstructure in most critical region, with the smallest average grain size. Additionally, it maintained good geometrical accuracy, indicating that it offers a promising route for producing high-quality magnesium-alloy components via WADED.

1. Introduction

Magnesium alloys have attracted growing attention due to the increasing demand for lightweight structural materials [1,2]. Magnesium is characterised by its low density, high specific strength, high specific stiffness, and excellent castability. Due to this unique combination of properties, magnesium alloys are widely utilised in the automotive, aerospace, and space industries [1,3–8]. However, their hexagonal close-packed (HCP) crystal structure provides only a limited number of slip systems, resulting in poor formability at room temperature [1,2,7,9].

Among magnesium alloys, Mg–Al–Zn systems (AZ series) are the most widely used [10]. The AZ61 magnesium alloy, containing 6 wt. % aluminium and 1 wt. % zinc, is a representative alloy whose mechanical properties are strongly influenced by grain size and the presence of precipitates. Alloys with finer grains exhibit higher tensile strength, hardness, and elongation, while tensile strength can be approximated from hardness [11]. A commonly used simplified relationship for this approximation is $\sigma = H/0.3$ [12]. Grain refinement can be achieved through higher cooling rates, whereas the volume fraction of precipitates depends on alloying content, cooling rate, and subsequent heat treatment [3].

Due to the limited formability of magnesium, casting is commonly used for producing semi-finished products. However, casting often leads to defects such as porosity, cracking, and coarse grains [1,2,13]. Among

various casting methods, high-pressure die casting (HPDC) is the most suitable for magnesium alloys, as it provides extremely high cooling rates that produce finer grain structures and improved mechanical properties [14].

Additive manufacturing (AM) offers an alternative approach to casting, providing high cooling rates and enabling refined microstructures along with design flexibility [2,15]. Metal additive manufacturing (MAM) encompasses several techniques, including powder bed fusion (PBF), direct energy deposition (DED), sheet lamination, and friction stir deposition [7,16]. One promising DED-based process is Wire Arc Direct Energy Deposition (WADED) [1]. However, the additive manufacturing of magnesium alloys remains challenging due to their low melting and boiling points combined with high reactivity, which often leads to defect formation [8,9,17,18]. These defects can be mitigated by optimising process parameters and employing suitable welding modes. The Cold Metal Transfer (CMT) variant of WADED, a modified form of MIG welding, reduces heat input by approximately 33 % and therefore represents the most appropriate option [6,16].

In recent years, WAAM/WADED has been increasingly applied to a variety of magnesium alloys, including AZ series [9,10], or Mg–RE systems [10,19], where improved corrosion performance, refined microstructures, and reduced defect formation have been achieved through process optimisation and controlled thermal input. Several studies have demonstrated that the selection of alloy type, interpass temperature, and deposition parameters can substantially affect grain

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Table 1

Process parameters.

I_boost	t_I_boost	I_sc_wait	vd_sc_wait	I_sc2	Travel speed	Preheat temperature
[A]	[ms]	[A]	[m•min ⁻¹]	[A]	[mm•s ⁻¹]	[°C]
430	2,5	35	30	50	15	250

refinement and the performance of WAAM-produced magnesium components [19,20]. These developments illustrate the growing potential of WAAM for manufacturing structurally demanding magnesium-alloy parts.

WADED-manufactured parts typically consist of thin-walled and volumetric sections. Thin-walled parts are formed from single-track deposits, whereas volumetric parts are constructed from multiple adjacent beads. To achieve sufficient fusion between neighbouring deposits, the overlap distance must be adequately defined, for which several models have been proposed [21,22]. The Flat-Top Overlapping Model (FOM) assumes that overlapping regions fill the valleys between adjacent deposits, resulting in a flat surface [21]. In contrast, the Tangent Overlapping Model (TOM) does not assume a flat surface but rather calculates adjacent beads with tangent edges, providing superior geometrical accuracy [22].

Various deposition strategies can be employed for volumetric WADED parts, including Raster, ZigZag (Meander), Contour, and Contour + ZigZag trajectories [23]. Because the start and end points of weld deposits often exhibit poor geometrical accuracy [24], continuous trajectories, such as Spiral or ZigZag, are preferred to minimise surface irregularities. Different strategies influence both geometrical accuracy and thermal distribution [23].

Deposition patterns also affect heat accumulation, which can reduce cooling rates and lead to casting-like defects. The recently introduced S-pattern trajectory was designed to improve thermal uniformity during deposition [25]. Comparisons among ZigZag, Spiral, and S-pattern strategies demonstrated that the S-pattern provides the most uniform thermal distribution, followed by Spiral and ZigZag [26]. Grain size analysis of Al-4046 deposits revealed large, elongated grains in the coarse-grain regions for the ZigZag and Spiral paths. In contrast, the S-pattern resulted in more equiaxed morphologies and fewer coarse grains [23].

While the studies mentioned above demonstrate differences among deposition strategies, they may also have several unresolved issues regarding the variability of microstructure in different locations. In particular, previous analyses have not fully addressed how grain morphology changes across various regions of the deposit. Although Köhler et al. (2021) [23] identified the S-pattern trajectory as the most effective in terms of thermal distribution, the microstructural analysis was performed only at a single sample location. To accurately determine the grain size range, critical zones of each trajectory must be identified. This concept is further supported by FEM-based maximum temperature maps reported in the same work [23]. To accurately assess these zones, cooling rate analysis is necessary, which can be performed using FEM simulations to identify regions with the lowest cooling rates.

This study focuses on evaluating deposition trajectories for the WADED fabrication of volumetric components made from AZ61 magnesium alloy. To enable WADED to serve as an alternative to conventional casting, it is essential to suppress pore formation and coarse-grain development. While porosity can be minimised by adjusting process parameters, achieving a homogeneous grain structure requires careful selection of the deposition strategy. In this work, FEM simulations are employed to identify critical zones with low cooling rates for each trajectory, followed by experimental grain size and hardness analyses to assess the impact of deposition strategy on microstructural refinement.

Table 2

Single-track deposit characteristics for fabrication of volumetric parts.

Thermal Power P	Deposit Width w	Deposit Height h	Overlapping distance D1	Overlapping distance D2
[W]	[mm]	[mm]	[mm]	[mm]
1404,51	10,46	3,01	6,97	5,23

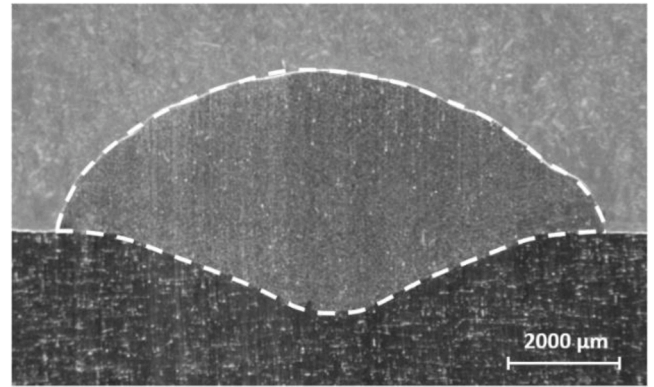


Fig. 1. Single-track deposit geometry used for the fabrication of volumetric parts.

2. Materials and methods

2.1. Materials and equipment

To investigate the influence of deposition strategy on microstructure, specimens were prepared via Wire Arc Direct Energy Deposition (WADED). A Cold Metal Transfer (CMT) power source (Fronius TPS 3200 CMT) was integrated with a six-axis robotic arm (KUKA KR 60 HA), providing high positional repeatability and motion precision. The feedstock material was AZ61 magnesium alloy wire with a diameter of 1.6 mm, while an AZ91 alloy plate served as the substrate to ensure metallurgical compatibility. Pure argon was used as the shielding gas at a flow rate of 20 l•min⁻¹ to prevent oxidation. The substrate was fixed to a temperature-controlled base plate to maintain consistent thermal conditions during the deposition process.

2.2. Process parameters

The process parameters used for fabricating volumetric samples were selected based on previous studies [27,28] and are listed in Table 1. These settings produced weld deposits with a wide and low geometry, resulting in a deposit width of 10.46 mm and height of 3.01 mm in this study. A single-track was deposited using these parameters, and the input thermal power was calculated to be 1404.51 W based on the measured current and voltage, assuming a 100 % thermal efficiency. The thermal efficiency of the CMT process is still under investigation, but several values have already been reported in the literature. Depending on the process conditions, the thermal efficiency may range from 60 % to 90 % [29]. In this study, an efficiency of 100 % was assumed as a simplification, since the exact value is unknown and the input thermal power is primarily used to identify trends in the FEM simulation rather

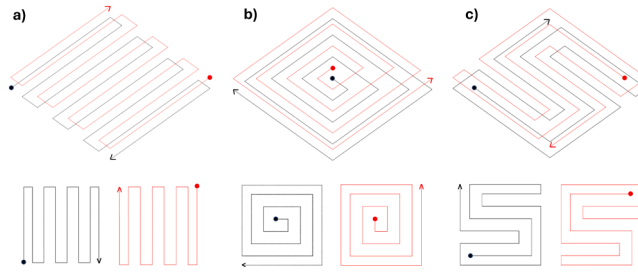


Fig. 2. Deposition strategies for manufacturing of testing cubes a) zigzag, b) spiral, c) S-pattern.

Table 3
Parameters for thermal simulation.

Parameter	Unit	Selected value
Cluster Volume	mm^3	21
Material Deposition Rate	$\frac{\text{mm}^3}{\text{s}}$	315
Preheating Temperature	$^{\circ}\text{C}$	250
Thermal Power	W	1400
Ambient Temperature	$^{\circ}\text{C}$	22 $^{\circ}\text{C}$
Convection Coefficient	$\frac{\text{W}}{\text{m}^2 \cdot \text{K}}$	10

than to obtain precise absolute values. The overlapping distances, D1 (≈ 0.738 w) and D2 ($= 0.5$ w), were determined using the Flat-Top Overlapping Model (FOM) [30] due to its simplicity and ease of application. Here, D1 represents the overlap between inner deposits, while D2 denotes the overlap at the edges of adjacent parallel patterns. All the deposit characteristics are listed in Table 2. The resulting deposit geometry is shown in Fig. 1.

2.3. Deposition strategies

Three deposition path strategies were selected for detailed analysis: the Zigzag pattern, which follows a linear back-and-forth scanning path (Fig. 2a); the Spiral pattern, consisting of a continuous outward spiral from the center (Fig. 2b); and the S-pattern, a hybrid raster method designed to periodically reverse and shift the path direction (Fig. 2c). Each deposition strategy covered a square build area of 45.34×45.34 mm, with dimensions chosen based on the overlapping distance between deposits. The final height of each sample was 50 mm, built up in 20 successive layers. The inter-layer time delay was adjusted according to the top surface temperature. Specifically, each layer was allowed to cool to an interpass temperature of 250 $^{\circ}\text{C}$, as measured using a thermal camera (Workswell WIC Industrial). To maintain geometrical precision, the trajectory was rotated by 180 $^{\circ}$ for each successive layer, with the spiral pattern additionally changing the direction of rotation. All three volumetric samples were subsequently fabricated according to deposition strategies on a single cylindrical build plate with a diameter of 160

mm and a height of 25 mm, made of AZ91 alloy. After the manufacturing process, all samples were measured using an optical 3D scanner (ATOS Triple Scan III). The scanned data were processed in the GOM Inspect 2018 software to evaluate the geometry of fabricated volumetric samples. Geometry was evaluated on the upper face of the cube. The measurement was performed by translating the surface plane of the substrate to the centroid of the upper surface of the cube and subsequently mapping the deviations from this plane.

2.4. Thermal simulation

Thermal modelling was performed using Ansys 2025 R1 with the DED Process add-on to simulate temperature evolution during deposition. The simulation accounted for transient heat transfer, including temperature-dependent thermal conductivity and specific heat capacity.

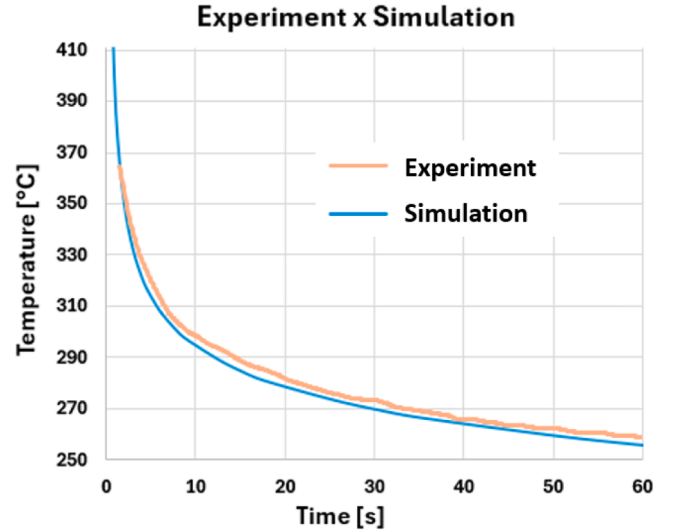


Fig. 4. Comparison of simulated and experimentally measured cooling curves for a single-track deposit.

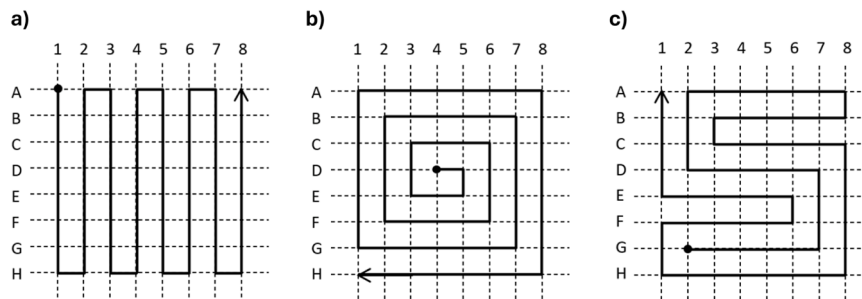


Fig. 3. Map of the cells in a) zigzag trajectory, b) spiral trajectory, c) S-pattern trajectory.

Based on the overlapping distance of inner deposits and the deposit height, the simulated deposit size was set to 7×3 mm. The simulation in Ansys 2025 R1 allows only rectangular deposits and dimensions of 7×3 mm have the same cross-sectional area and height as the experimentally obtained deposit. Thermal power input was derived from measurements taken during the fabrication of a single-track deposit. Other adjustable simulation parameters are listed in Table 3.

The thermal behaviour of the deposits depends on the distance from the substrate. Previous studies have shown that after approximately the eighth layer, the thermal behaviour becomes largely homogenised [31]. For this reason, the top layer (the 20th layer) was selected for the simulation, as it is expected to represent layers with homogenised thermal behaviour and provides convenient access for subsequent sampling. To identify regions with a higher probability of thermal accumulation and grain coarsening, the deposition area was divided into an 8×8 grid (Fig. 3). Cooling rates were evaluated for each grid cell, allowing for the precise detection of thermally critical regions most susceptible to microstructural degradation. The resulting cooling rate data were then used to select sampling locations and to compare the effectiveness of different deposition strategies in controlling heat accumulation.

2.5. Simulation validation

The selected simulation parameters were validated by comparing simulated cooling rates with experimentally measured cooling behaviour during the deposition of a single-track deposit. Thermal imaging was used to record the cooling curve. From this data, the time required for the track to cool from the peak temperature to the preheat temperature level was determined. This experimentally measured cooling duration was then compared with the simulated cooling profile obtained under identical process conditions, including power input, preheat temperature, and environmental settings (Fig. 4). The close agreement between simulated and measured cooling rates confirmed the accuracy of the thermal model and its suitability for predicting heat distribution in more complex volumetric builds.

2.6. Metallography and microstructural analysis

Multiple zones were analysed from each sample, including one located in the critical region of each deposition trajectory and several others in selected zones to validate the simulation results. Samples were sectioned with a material allowance of 1.7 mm for subsequent grinding and processing. The specimens were ground using SiC papers with grit sizes of 320, 600, 1000, 2500, and 4000, and then polished with colloidal silica on a grinding and polishing machine (SAPHIR 250 A2-ECO). The polished surfaces were etched with acetic Picral solution for 50 s. Grain size analysis was always performed at the mid-height of the weld deposit.

Grain structures were examined using optical microscopy (KEYENCE VHX-6000). Due to the low visibility of grain boundaries, manual outlining of grains was performed prior to automatic grain size evaluation using ImageJ software. The average grain size was calculated as a weighted mean based on the measured grain areas.

2.7. Microhardness

Microhardness was evaluated in the same regions where grain size measurements were performed. The measurements were conducted using a Qness 60 A+ microhardness tester with a load of 0.1 kgf (HV0.1). For each region, ten individual indentations were made to ensure statistical reliability. The average hardness value for each zone was calculated from these measurements. However, it is not possible to effectively compare the measured values for different deposition strategies. As already mentioned in Section 2.3, the samples were produced consecutively on a single build plate; therefore, the samples produced

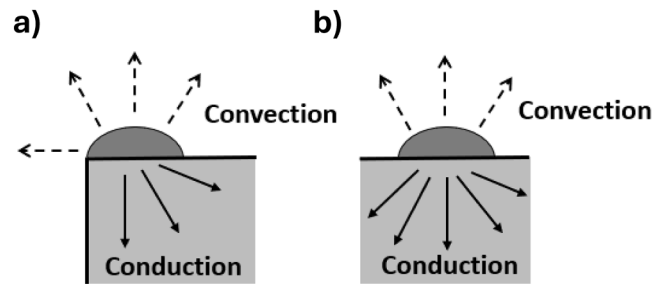


Fig. 5. Difference between cooling rates at the a) edge of the part, b) centre of the part.

earlier were thermally affected by the production of the subsequent ones. For this reason, only the microhardness values within each cube were compared in the discussion section, not between them.

3. Results and discussion

3.1. Simulation results

Three important findings emerged from the thermal simulations of the deposition trajectories (1F, 2F, 3F). First, the cooling rate progressively decreased with the accumulation of deposited material (1F). Second, slower cooling rates were observed near the outer edges of the component (2F). Lastly, the corners of the deposition trajectories exhibited lower cooling rates compared to the straight sections (3F).

The heating of the base material can explain the progressive decrease in cooling rate (1F). As the volume of deposited material increased, the substrate temperature rose, leading to reduced cooling efficiency. Similar effects have been reported in thin-walled parts, where pronounced temperature gradients develop during deposition [32].

The slower cooling rate observed at the outer edges of the cube (2F) can be attributed to less effective heat dissipation caused by the absence of adjacent material (Fig. 5). A comparable phenomenon occurs in multi-layer deposition: the farther the deposition is from the substrate, the slower the cooling rate becomes due to the reduced heat conduction through surrounding material [33].

Lower cooling rates in the corners (3F) result from the overlapping thermal influence of multiple path directions, leading to localised heat accumulation. In addition to thermal effects, material build-up can also occur in the corners of the deposition trajectories [34]. Excessive deposition may further intensify heat accumulation; however, this aspect was not included in the present simulation. This phenomenon could not be implemented in the FEM model because it would require dynamic geometry development along the trajectory and prediction of the shape of the bead. Importantly, the simulated critical zones were located in the corner areas, which means that omitting corner build-up does not change the location of these areas. The critical areas will remain the same, only the effect of grain coarsening was amplified.

Together, these three observations highlight the regions most susceptible to microstructural coarsening and emphasise the importance of trajectory design in controlling the thermal behaviour of WADED builds.

Critical zones with the lowest cooling rates were identified for each deposition strategy. All critical zones were consistent with the general trends described earlier. For clarity, only the critical regions and representative zones that support the previously discussed findings are shown in the figures below (Fig. 6).

For the ZigZag trajectory, the critical zone was located near the end of the deposition path, at the outer edge and in the corner of the component (H8, Fig. 6a). Fig. 6a illustrates the regions with low cooling rates (H1, H7, and H8) and the zones where the cooling rate decreased progressively with material accumulation (H3 and H5). Fig. 6b presents the corresponding cooling rates for these zones. Zone H1 exhibits a low cooling rate due to its position in the component corner. Zones H7 and

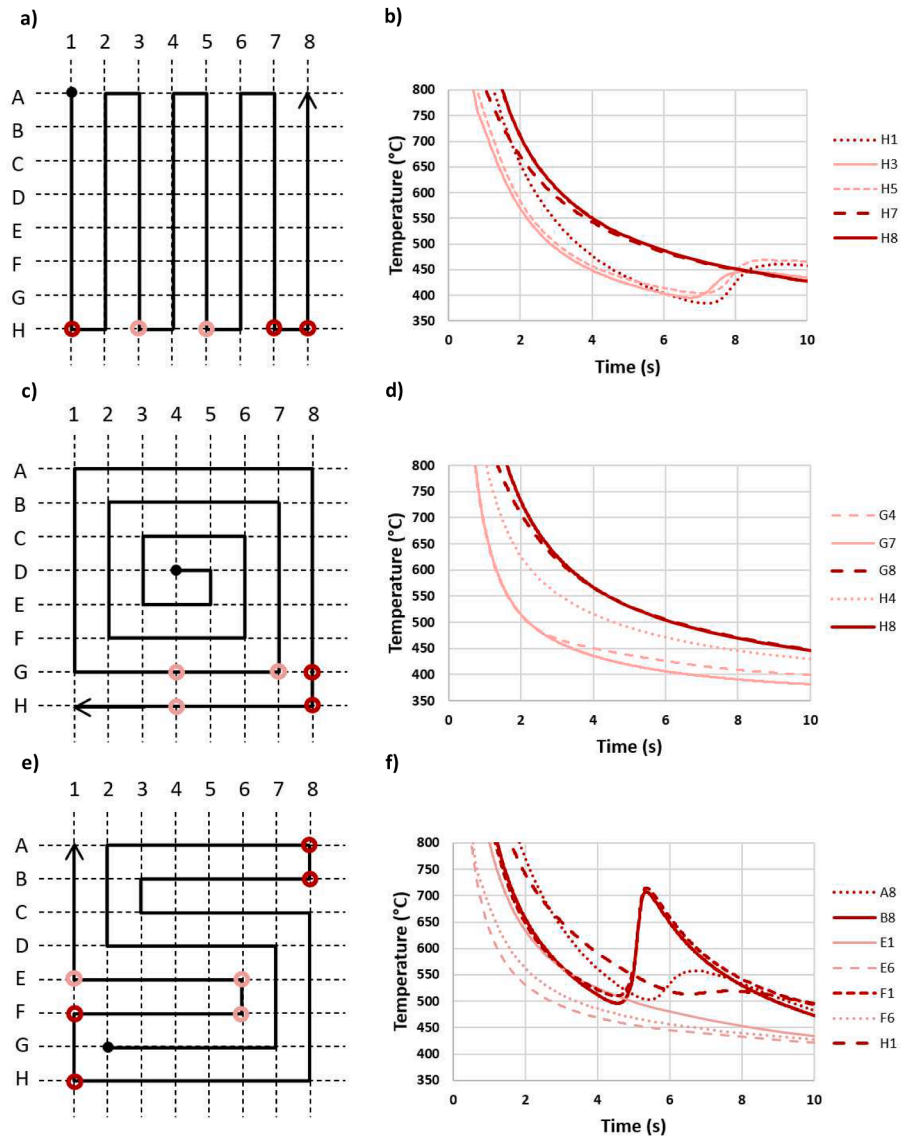


Fig. 6. Positions of critical zones in a) zigzag, c) spiral, e) S-pattern. cooling rates of critical zones in b) zigzag, d) spiral, f) S-pattern.

H8 are located near the end of the trajectory, with H8 also situated in a corner, making it the most thermally critical region. The cooling curves of zones H1, H3, and H5 exhibit minor reheating at approximately the 8-s mark. This effect is assumed to be insufficient to influence the resulting

microstructure.

Similar to the ZigZag trajectory, the Spiral trajectory exhibited its most critical zone near the end of the deposition path, at the corner of the trajectory and the component itself (H8, Fig. 6c). A slightly less

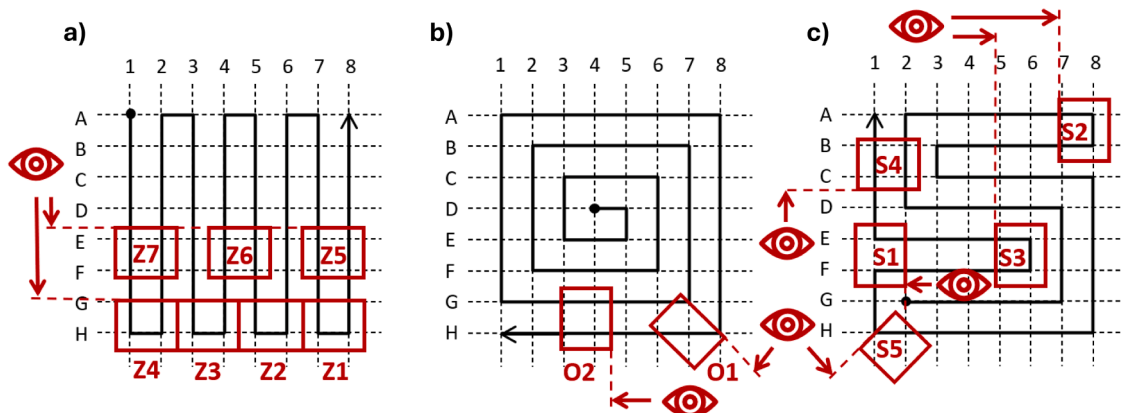


Fig. 7. Selected metallography samples orientations of the a) zigzag trajectory, b) spiral trajectory, c) S-pattern trajectory.

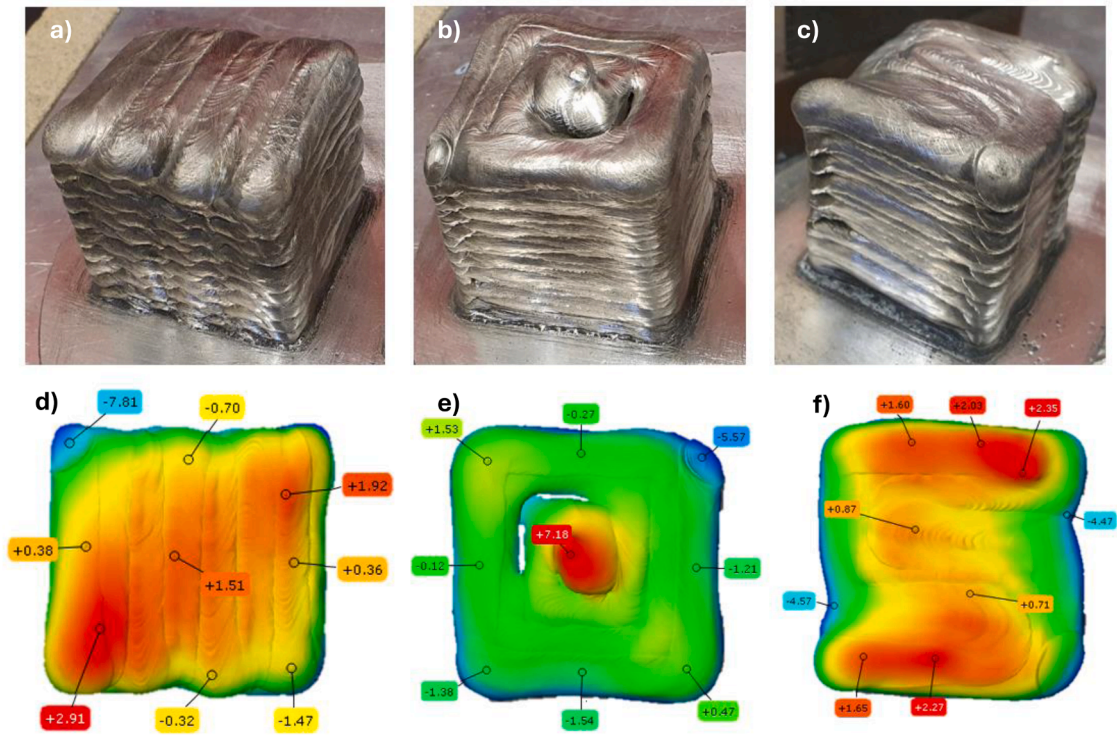


Fig. 8. Volumetric samples fabricated with a) ZigZag trajectory, b) spiral trajectory, c) S-pattern trajectory, geometrical deviations of the d) zigzag trajectory, e) spiral trajectory, f) S-pattern trajectory.

critical zone, G8 (Fig. 6c), is located adjacent to H8. For comparison, the cooling rate of zone G7 is also shown in Fig. 6d. This zone is positioned at the corner of the trajectory but not at the outer edge of the cube, and it lies farther from the end of the path. To further compare the thermal behaviour of corner and straight path sections, zones G4 and H4 were analysed. As shown in Fig. 6d, the straight sections exhibited higher cooling rates than their corner equivalents, confirming that corners are more prone to thermal accumulation.

For the S-pattern trajectory, the lowest cooling rates were observed at the corners of the deposition path and along the outer edges of the cube (Fig. 6e,f). Accordingly, Fig. 6e,f highlights primarily these regions. Zones E6 and F6 were also included, as this area contains two consecutive path corners, which are expected to exhibit reduced cooling rates even though they are not located at the cube's edge. The lowest cooling rates were found in cells A8, B8, F1, and H1. Zones F1 and B8 displayed similar cooling behaviour, including comparable reheating patterns approximately at the 5-s mark (Fig. 6f). The rapid reheating immediately after deposition was the reason these cells were identified as critical. However, determining the precise location of the most critical zone is challenging for this trajectory type. A quantitative assessment of reheating effects would be required to clarify this behaviour, but such an analysis lies beyond the scope of this study.

3.2. Sample selection

Samples for grain size analysis were selected based on the results of thermal simulations conducted for each deposition strategy. For every trajectory type, one or more sampling locations were chosen to capture the regions exhibiting the most critical microstructure. Additional locations were analysed to validate the simulated identification of these critical zones.

For the ZigZag trajectory, sample Z1 represented the critical location (Fig. 7a). To verify finding 1F, samples Z2, Z3, and Z4 were analysed (Fig. 7a). To examine finding 2F, additional samples Z5, Z6, and Z7 were selected.

For the Spiral trajectory, sample O1 corresponded to the critical zone (Fig. 7b). To verify finding 3F, sample O2 was analysed. Because the thermal influence of the trajectory corner needs to be distinguished from that of the component edge, additional inner-path samples (G7 and G4) were included.

For the S-pattern trajectory, samples S1, S2, and S5 contained the predicted critical zones (Fig. 7c). Sample S3 was located in a region of thermal accumulation but not at the cube edge. Although simulations suggested that this zone was non-critical, it was nevertheless selected for verification.

Sample S4 was selected to represent the condition where all three general findings from the simulations are simultaneously valid. Under these circumstances, samples S4, O2, and Z5 are located in regions with comparable theoretical cooling conditions. All three are positioned along the edge of the component, approximately 15 mm before the end of the deposition path, and on a straight section of the trajectory. The only variable among them is the thermal distribution characteristic of each deposition strategy.

3.3. Fabrication of volumetric samples

Three cubes were manufactured using the deposition strategies described in Fig. 2. Several geometrical inaccuracies were observed, both during the fabrication process and in the final parts.

The ZigZag trajectory exhibited the highest geometrical precision on the top surface (Fig. 8a,d). Minor deviations were observed at the segments corresponding to the start of the trajectories, which were slightly raised (+2.91 mm, +1.92 mm), and at the ends of the trajectories, which were lowered (−7.81 mm, −1.47 mm). Suppose the end of the trajectory of the final deposited layer, which substantially reduces the local surface height, is excluded from the evaluation. In that case, the range of height deviations is from −1.47 mm to +2.91 mm.

The Spiral trajectory showed the poorest geometrical accuracy among the three cubes. Since the start of each layer coincided at the exact location, these regions were significantly raised (+7.18 mm,

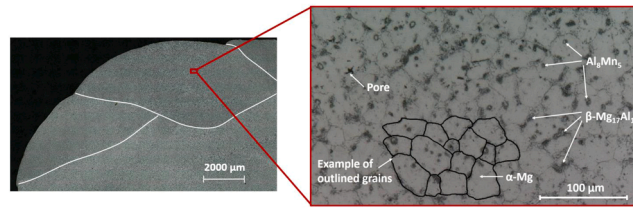


Fig. 9. Example of microstructure analysis.

Fig. 8b,e). The ends of the trajectories were slightly lower (−5.57 mm, −1.38 mm), although the deviation was minimal. Excluding the end of the final deposit, the range of height deviations is from −1.58 mm to +7.18 mm.

The S-pattern trajectory demonstrated acceptable geometrical precision. The start points were raised slightly less than in the ZigZag trajectory (+2.35 mm, +1.65 mm, Fig. 8c,f). The main limitation in terms of precision was observed along the sides of the part, where the trajectory turns from the edge toward the middle of the component and the height is significantly lowered (−4.47 mm, −4.57 mm). The range of height deviations is from −4.57 mm to +2.35 mm.

The Spiral trajectory exhibited the most extensive deviation range and contained an unacceptable defect at the centre of the surface, which could ultimately lead to process failure during manufacturing. The S-pattern trajectory showed a narrower deviation range with no significant defects observed. The ZigZag trajectory demonstrated the smallest deviation range and was free of detectable defects.

3.4. Grain size comparison

The microstructure analysis revealed Al–Mn and Mg–Al phases, which is consistent with the previous findings by Slavíček et al. [27]. Grain size was analysed at the centre of each weld deposit in the selected sampling locations (Fig. 9). Minimal number of analysed grains for each location is 85. ImageJ software was used to measure the area of individual grains. Grain diameters were then calculated based on the equivalent circular area. The average grain size for each region was determined using a weighted mean and weighted standard deviation, with the weight corresponding to the measured grain area.

Grain size was analysed along each deposition trajectory following the procedure described above. For the ZigZag trajectory, the critical zone (G8, Fig. 10a,d) exhibited a grain size of 51.88 µm. No larger

average grain sizes were observed along this trajectory, confirming that the simulation correctly predicted the critical location (Table 4). Slightly increasing grain sizes were observed in zones G2, G4, and G6 (Fig. 10a). The average grain size at zone E4 was smaller than in E1 and E8 (Table 4). These results confirm findings 1F and 2F: grain size tended to

Table 4

Average grain sizes in selected locations.

ZigZag		Spiral		S-pattern	
Position	Avg. size [µm]	Position	Avg. size [µm]	Position	Avg. size [µm]
E1	37,51 ± 10,08	G4	34,76 ± 9,47	A7	32,31 ± 8,64
E2	35,48 ± 10,78	G7	37,97 ± 12,33	B7	38,52 ± 8,84
E4	32,03 ± 8,01	H4	39,75 ± 13,50	C1	32,54 ± 9,25
E5	26,46 ± 6,63	H8	60,27 ± 22,69	E2	30,60 ± 8,87
E7	33,48 ± 8,51			E5	30,68 ± 6,99
E8	44,36 ± 13,92			F2	39,01 ± 11,31
G1	29,55 ± 7,42			F5	26,32 ± 7,14
G2	36,96 ± 11,01			H1	44,71 ± 11,63
G3	28,73 ± 7,73				
G4	37,66 ± 10,00				
G5	30,79 ± 9,82				
G6	38,09 ± 11,61				
G7	32,89 ± 9,52				
G8	51,88 ± 12,58				

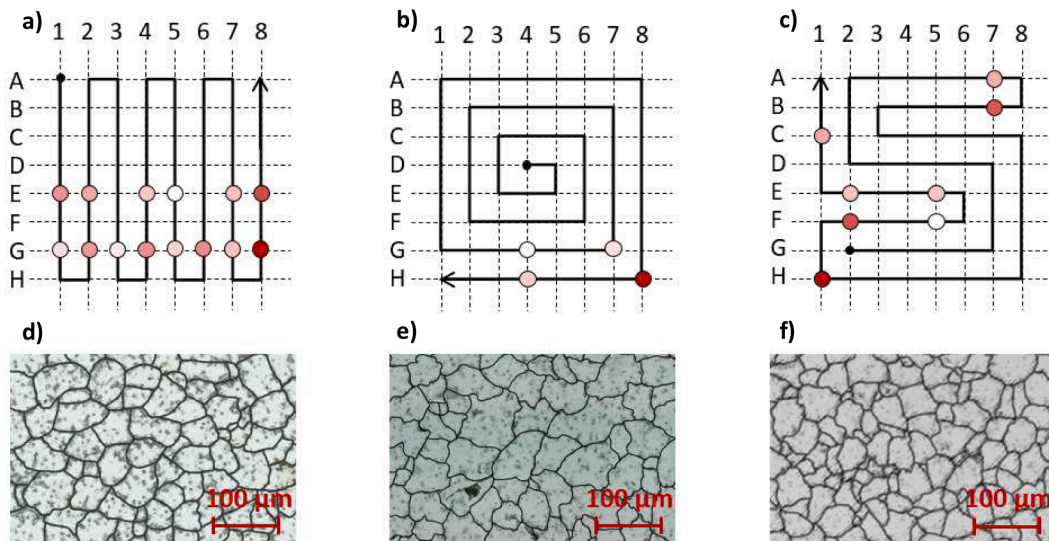


Fig. 10. Grain size measurement locations of the a) zigzag trajectory, b) spiral trajectory, c) S-pattern trajectory; microstructure evaluation in critical zones d) ZigZag trajectory (G8), e) spiral trajectory (A8), f) S-pattern trajectory (H1).

Table 5

Evaluation of the grain sizes of various deposition strategies.

	ZigZag	Spiral	S-pattern
Max. grain size	51.88 μm	60.27 μm	44.71 μm
Grain size range	26.46 – 51.88 μm	34.76 – 60.27 μm	26.32 – 44.71 μm
S4, O2, Z5	44.36 μm	39.75 μm	32.54 μm

increase along the trajectory, and regions at the edges of the component exhibited larger grains compared to the middle, regardless of whether they were at the start or end of the trajectory.

For the Spiral trajectory, the maximum grain size of 60.27 μm was found at the critical location predicted by the simulation (H8, Fig. 10b, e). Straight sections of the trajectory exhibited smaller grain sizes compared to the corners, regardless of their position relative to the cube edges (Table 4). This observation confirms finding 3F. As the straight segments were analysed later along the deposition path, these results were not influenced by the first general finding (1F).

In the S-pattern trajectory, the maximum grain size of 44.71 μm was observed in zone H1 (Fig. 10c, f). Zones B7 and F2 also exhibited slightly larger grains, consistent with the nearly identical cooling rates predicted by the simulation for these areas. Regions E5 and F5 showed smaller grains, in agreement with the simulation predictions (Table 4).

Regions of samples S4 (S-pattern), O2 (Spiral), and Z5 (ZigZag) were selected as representative zones located near the edge of the component, approximately 15 mm before the end of the trajectory, along straight segments. These locations satisfy all three general findings derived from the simulations, providing comparable cooling conditions. The only varying parameter among these regions was the thermal distribution associated with each deposition strategy.

This comparison indicates that the S-pattern trajectory offers the most uniform thermal distribution, followed by the Spiral trajectory, while the ZigZag trajectory exhibits the least effective thermal control (Table 5). These observations are consistent with the analysis reported by Sun et al. (2021) [26].

The S-pattern trajectory exhibited the narrowest grain size range among all trajectories (Table 5). It should be noted that the selected zones were chosen to capture the most critical regions rather than to represent the full range of grain sizes; therefore, the reported range for the entire cube is purely illustrative.

The Spiral trajectory produced coarser grains than the ZigZag trajectory, despite its predicted thermal advantage. This discrepancy may result from local overheating within the Spiral strategy, although the overall thermal distribution was still better than that of the ZigZag trajectory. An alternative explanation could be an evaluation procedure error: the critical sample from the Spiral trajectory was extracted at a 45° angle relative to the deposition direction, which may have distorted the apparent grain size due to non-orthogonal viewing. The grains are generally oriented in the direction of deposition, corresponding to the predominant cooling direction [35]. However, in the corner region of

the trajectory, the cooling direction becomes unclear. Therefore, a 45° viewing angle was selected as theoretically orthogonal to the expected direction of cooling. Non-orthogonal observation relative to the cooling direction may lead to an apparent increase in the observed grain size. It should also be noted that, although the Spiral sample was extracted from the exact critical zone, the ZigZag sample was taken adjacent to the critical zone due to spatial constraints on collecting multiple samples.

3.5. Microhardness comparison

Microhardness measurement locations (Fig. 11) were selected to either validate the general findings from the simulations or to examine the critical zones of each deposition strategy. Ten indentations were made for each zone, and the average values and standard deviations are reported below (Table 6).

For the ZigZag trajectory, hardness decreased progressively along the trajectory (G2, G4, G6, G8), confirming finding 1F. Zone E1 exhibited lower hardness than zone E4, in agreement with finding 2F. In the Spiral trajectory, lower hardness was observed at the corners (H8) compared to the straight segments (H4), confirming finding 3F. Critical zones in both ZigZag and Spiral trajectories corresponded to the lowest hardness values.

In contrast, the S-pattern trajectory did not show a similar trend, as hardness values were relatively uniform across all analysed zones. This observation is consistent with the narrower grain size range observed for this trajectory. Any apparent deviations in the S-pattern critical zones may be attributed to measurement error when considering the limited variation in grain size.

A notable observation is the overall lower hardness of the S-pattern cube compared to the ZigZag and Spiral cubes. The three cubes were manufactured consecutively in the order ZigZag–Spiral–S-pattern, while all remained on the heating plate at 250 °C. Consequently, the ZigZag and Spiral cubes likely experienced considerable precipitation hardening. S. Coletto reported that AZ91 subjected to isothermal ageing at 250 °C exhibited a significant increase in hardness within just one hour (≈approximately 10 %) [36]. Since AZ61 has a composition similar to AZ91, a comparable behaviour can be expected. Therefore, comparisons of hardness were made only within each cube, rather than between

Table 6

Microhardness in selected locations.

ZigZag		Spiral		S-pattern	
Position	HV	Position	HV	Position	HV
E1	66,04 ± 2,47	H4	66,81 ± 2,86	B7	63,58 ± 3,41
E4	68,06 ± 2,95	H8	64,54 ± 3,28	C1	63,52 ± 2,85
G2	69,39 ± 5,35			F2	61,40 ± 2,71
G4	66,95 ± 3,22			H1	62,56 ± 2,81
G6	67,19 ± 3,52				
G8	65,22 ± 2,99				

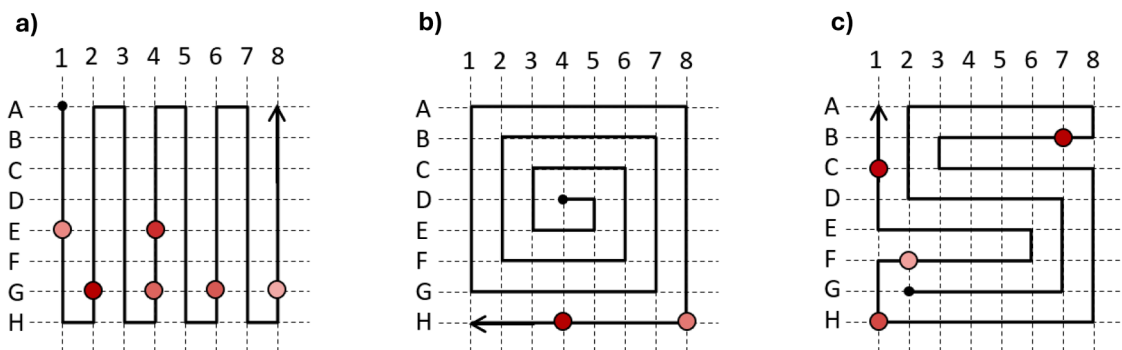


Fig. 11. Microhardness measurement locations of the a) ZigZag trajectory, b) spiral trajectory, c) S-pattern trajectory.

cubes.

3.6. Interrelation of the discussed results

In the previous sections, the effects of deposition trajectories on thermal distribution, grain size, and microhardness were discussed separately. In this paragraph, the results are placed into a broader context to highlight their interrelationships. Faster cooling of the melt pool promotes a higher nucleation rate [37], resulting in finer microstructures. In the context of our study, this effect can be illustrated using the Zig-Zag trajectory. Points H3 and H5 experienced significantly faster cooling, which is reflected in the microstructural analysis, showing a finer grain structure at these locations. Connection between grain size and mechanical properties is generally described by the Hall-Petch relationship [38]. This relationship implies that smaller grains act as barriers to dislocation motion, thereby increasing strength. Better mechanical properties usually correspond to higher microhardness [12]. Microhardness in relation to grain size was described in this study, for example, on the Spiral trajectory. Here it is clear that the finer grain structure at point H4 corresponds to higher microhardness. The same conclusion can be drawn from the Zig-Zag trajectory and points G8 and G2. With these results the direct link between cooling rate, grain-size, and microhardness is confirmed.

5. Conclusion

This study demonstrated the significant influence of deposition path strategies on the microstructure of volumetric WADED components made from AZ61 magnesium alloy. Thermal simulations successfully predicted regions of thermal accumulation and low cooling rates, which were confirmed by optical grain size and microhardness measurements. The S-pattern trajectory yielded the finest and most uniform microstructure, characterised by the lowest maximum grain size (44.71 μm), the narrowest grain size range (26.32–44.71 μm), and effective thermal control, while maintaining acceptable geometrical accuracy. In contrast, the ZigZag trajectory offered superior geometrical precision but less consistent microstructural quality, and the Spiral trajectory generated the worst geometrical accuracy and the coarsest grains (60.27 μm). These findings demonstrate that careful trajectory design is a powerful tool for tailoring microstructure in WADED components. They provide a foundation for process-aware strategies in the additive manufacturing of lightweight structural parts.

Data statement

The data that support the findings of this study are openly available in Zenodo Repository at <https://doi.org/10.5281/zenodo.17506822>. The original version of the author's preprint of this study is openly available in the Zenodo Repository at <https://doi.org/10.5281/zenodo.17506854>.

CRediT authorship contribution statement

Filip Seidler: Writing – review & editing, Writing – original draft, Visualization, Methodology, Investigation, Data curation. **Jakub Slavíček:** Writing – review & editing, Methodology, Investigation, Data curation, Conceptualization. **Daniel Koutný:** Supervision, Funding acquisition.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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Data availability

I have shared the link to my data.

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7 CONCLUSIONS

This dissertation investigated the development of the WADED process as a manufacturing method for structured magnesium alloy components. The scientific questions addressed the main problems identified from the literature review. Through the presented scientific papers, the dissertation aimed to develop a reliable and efficient manufacturing process.

Despite growing interest in wire-based additive manufacturing, the application of WADED to magnesium alloys remains insufficiently explored. While extensive research has been conducted on WADED of aluminium and steel alloys, only a limited number of isolated studies have addressed magnesium alloys. Existing works typically focus on selected parameters such as wire feed speed and travel speed, compare different process modes, develop new materials, or evaluate corrosion behaviour, but often without addressing process stability, defect formation, thermal management, or residual stress development. Furthermore, the fabrication of volumetric components, which is essential for large-scale structured applications, has remained largely unexplored. This dissertation addresses these gaps by combining process development, numerical simulation, and experimental validation to establish a coherent workflow for WADED processing of magnesium alloys.

The results presented in this dissertation demonstrate that reliable wire arc directed energy deposition of magnesium alloys requires a suitable combination of fabrication parameters and process-related factors. The initial study showed that deposition path planning significantly affects the geometrical accuracy of thin-walled and overhanging structures, and proposed path-planning recommendations to limit the accumulation of layer-height deviations. Although the investigated sequencing strategies did not lead to a systematic reduction of residual stresses for the analysed wall lengths, they proved effective in extending the limits of unsupported overhangs. Subsequent studies identified material pretreatment, optimisation of characteristic CMT parameters, and thermal management as key factors governing reliable WADED processing of AZ61 magnesium alloy. These factors were shown to strongly affect process stability, weld bead geometry, metal transfer behaviour, and defect formation. The role of thermal management was further demonstrated through systematic variations in preheating and interpass temperatures, revealing an influence on penetration depth, arc stability, wall geometry, residual stress development, porosity formation, and microstructural evolution. Finally, the investigation was extended to volumetric components. Thermal numerical simulations were used to predict regions susceptible to grain coarsening, which were subsequently examined through targeted microstructural analyses. The results demonstrated that local cooling rates, grain-size distribution, and microhardness variation in volumetric components can be effectively influenced by the applied deposition strategy. Deposition strategies that promote a more uniform heat distribution result in a finer microstructure and improved mechanical response.

The presented thesis contains original results extending the knowledge in the field of wire arc directed energy deposition of magnesium alloys. The results were confronted with previously published studies, and the main contribution of the thesis can be summarised into the following points:

- Deposition path planning was identified as a key design parameter influencing geometrical accuracy. Appropriate distribution of weld start points and end points and alternation of deposition directions significantly reduce the accumulation of layer-height deviations and improve dimensional accuracy.
- Although the investigated sequencing strategies did not lead to a systematic reduction in residual stresses for the tested 120 mm-long thin walls, they proved effective in extending the limits of overhangs fabrication. Using these strategies, overhangs with inclination angles up to 90° were successfully fabricated without support structures or part repositioning. Fabrication without repositioning is then advantageous, especially for large-scale manufacturing.
- Magnesium alloy filler wire surface purity, characteristic CMT parameters, and thermal management were identified as the key factors controlling WADED process stability. With optimised parameters, mechanical properties superior to those of cast magnesium alloys and comparable to those achieved by LPBF were obtained.
- Increasing preheating and interpass temperatures reduces residual stresses but simultaneously promotes grain coarsening, making temperature optimisation essential. For AZ61 magnesium alloy, an optimised preheating and interpass temperature of 250 °C resulted in a 40% reduction in residual stress, while maintaining sufficient grain size.
- Preheating and interpass temperatures of 250 °C also proved favourable for single-track geometry and porosity mitigation. Thin-walled samples produced under these conditions achieved a relative material density of 99.98%.
- Volumetric path-planning strategies promoting uniform heat distribution reduce grain coarsening and dimensional deviations, as confirmed by combined numerical simulations and experimental validation. An example is the S-pattern strategy, which provides the lowest maximum grain size and the narrowest grain-size and microhardness range, indicating homogeneous mechanical performance.

The practical applicability of the developed WADED approaches was demonstrated through the fabrication of a structured magnesium alloy component combining thin-wall and volumetric structures. The manufactured component highlights the potential to reduce the buy-to-fly (BTF) ratio significantly. At the same time, the application of deposition strategies and process approaches reported in the presented papers is expected to ensure favourable mechanical and material properties. Overall, this dissertation contributes to WADED processing of magnesium alloys and provides a foundation for future research and industrial implementation of this technology in lightweight structural applications.

Regarding the tested hypotheses, the obtained results are summarised in the following conclusions:

Q1: How does deposition path planning influence the geometrical accuracy and residual stress of components fabricated by WADED?

C1a: The geometric accuracy of thin-walled samples demonstrates that the different WADED path-planning strategies significantly influence the resulting component geometry. Deposition strategies that redistributed or alternated weld start and end locations across the component's height effectively mitigated excessive material accumulation and promoted a more uniform layer height. In contrast, thin walls with a higher number of welds starting or ending repeatedly located in layers above each other exhibited a systematic accumulation of local height deviations, leading to pronounced geometric irregularities. A similar effect was observed in volumetric components, where a spiral deposition trajectory with weld starts positioned in successive layers resulted in excessive material accumulation in the central part of the sample and significant deterioration of geometric accuracy. In deposition strategies in which the weld start and end were strategically varied, such large geometric deviations were not observed. **Thus, the H1a hypothesis was not falsified.**

C1b: Although deposition strategies employing modified welding sequences were expected to promote a more uniform heat distribution, the measured deflections after partial cutting did not exhibit any consistent trend indicating a reduction in residual stress levels. Possible reasons for the absence of the expected behaviour include the inappropriate sample design or the use of a CMT-based welding process. The original hypothesis was based on results obtained using traditional welding methods, such as SMAW and GMAW, which are typically characterised by higher heat input than CMT. Therefore, the use of CMT may reduce heat input to such an extent that welding sequence strategies do not lead to further reductions in residual stresses when fabricating this type of sample. To thoroughly verify the effect of sequence strategies on residual stresses, it would be necessary to test a larger number of samples of various lengths. A reduction in residual stress could be observed in longer samples. However, based on the presented results, the investigated WADED path planning strategies for 120 mm-long thin walls had no observable effect on reducing residual stresses in thin-walled components. **Thus, the H1b hypothesis was falsified.**

Q2: What are key process-related factors and how they affect process stability and defect formation during WADED of AZ61 magnesium alloy?

C2: The conducted research confirmed that the key process-related parameters for WADED of AZ61 magnesium alloy include the appropriate setting of CMT characteristic parameters and effective thermal management. In addition, another critical process-related factor was identified, namely the surface purity of the filler wire. Excessive amounts of oxides or hydroxides on the wire surface were found to adversely affect welding arc stability. Consequently, three different wire-cleaning approaches were investigated, and the most stable process behaviour combined with the lowest defect occurrence was achieved when the wire surface was cleaned immediately before entering the welding source. This approach effectively limited oxide contamination, improved arc stability, and reduced defect formation. Regarding CMT process parameters, the welding boost phase current and its duration were found to significantly influence bead geometry, arc stability, and droplet transfer behaviour. Improper parameter combinations led to unstable metal transfer and increased susceptibility to porosity, whereas optimised settings enabled a stable and repeatable deposition process. Thermal management, particularly base plate preheating and interpass temperature control, was identified as another critical factor. Increasing the preheating temperature led to reduced porosity formation and improved melt pool stability. By combining effective wire surface cleaning, optimised thermal conditions, and appropriately selected CMT parameters, reliable WADED fabrication of AZ61 magnesium alloy components was achieved, with relative material density reaching 99.98%. **Thus, the second hypothesis was not falsified.**

Q3: How do thermal management strategies affect the development of microstructure and residual stresses during WADED of AZ61 magnesium alloy components?

C3: During the fabrication of volumetric components, heat accumulation and regions with reduced cooling rates were found to promote grain coarsening, potentially degrading mechanical performance. Grain-size analysis of volumetric samples confirmed that the deposition strategy notably influences microstructural evolution. Strategies such as the S-pattern, which provide a more uniform heat distribution, resulted in the lowest maximum grain size and the narrowest grain-size range (26.32–44.71 μm). The results confirmed the expected trend that increasing the base-material preheating and interpass temperatures reduces residual stresses while promoting grain growth in thin-walled samples. Preheating and interpass temperatures of 250 °C were identified as optimal, providing a balanced compromise between residual stress mitigation and microstructural coarsening. At this temperature, the base material deflection, used as an indicator of residual stress level, was reduced by approximately 40%. At the same time, the grain size in coarse-grained regions increased from 19.2 to 26.9 μm . This grain size remains acceptable, as it corresponds to the lower bound of grain-size range observed in volumetric WADED-fabricated components. **Thus, the third hypothesis was not falsified.**

8 LIST OF PUBLICATIONS

8.1 Publications related to the thesis topic

SLAVÍČEK, J.; FRANKE, J.; JAROŠ, J.; KOUTNÝ, D.; Strategies for Wire Arc Additive Manufacturing of thin walls and overhangs, *Journal of Mechanical Science and Technology*, 2023, vol. 37, no. 1, p. 5529-5534, ISSN: 1738-494X.

SLAVÍČEK, J.; ŠVÁB, Č.; GNEIGER, S.; HURNÍK, J.; PROCHÁZKA, P.; KOUTNÝ, D.; Wire-arc directed energy deposition of AZ61 magnesium alloy fabricated by cold metal transfer: Processing, microstructure, and mechanical properties, *Scientific Reports*, 2026

SLAVÍČEK, J.; ZEMAN, S.; HORYNOVÁ, M.; SENCK, S.; KOUTNÝ, D.; Optimisation of preheating and interpass temperature in WADED magnesium alloy AZ61: A comparative study on microstructure, residual stresses, and internal defects, (*PREPRINT- under review*)

SEIDLER, F.; SLAVÍČEK, J.; KOUTNÝ, D.; Influence of deposition strategy on the microstructure of volumetric WADED-fabricated AZ61 magnesium alloy components, *Additive manufacturing letters*, 2025, vol. 16, p. 1-10, ISSN: 2772-3690

8.2 Other publications

CHOVANČÍKOVÁ, V.; GULAN, L.; KOREC, F.; NOVÁK, P.; SLAVÍČEK, J.; ĎURIŠKA, M.; Comparison of methods of topological optimization of a selected module of a mobile working machine and assessment of the possibilities of production by additive technology, *Advances in Science and Technology Research Journal*, vol. 19, no. 10, p. 199–211. doi:10.12913/22998624/206088.

8.3 Conference contributions

SLAVÍČEK, J.; HURNÍK, J.; KOUTNÝ, D. *Strategy for wire arc additive manufacturing of thin walls*. Reliability and Design of Additively Manufactured Materials, 4.–6.10. 2022, Belgrade, Serbia.

SLAVÍČEK, J.; ŠVÁB, Č.; ZEMAN, S.; SEIDLER, F.; HORYNOVÁ, M.; GNEIGER, S.; SENCK, S.; KOUTNÝ, D. *Process development for thin-walled AZ61 components produced by WADED*, 1.st MEBioSys conference, 9.–10.9.2025, Brno, Czech Republic

8.4 Other results

Functional specimens (RIV-G)

SLAVÍČEK, J.; ZEMAN, S.: Design of preheated base plate for WAAM processing of difficult-to-weld materials. IMID, FME, BUT, Technická 2896/2 616 69 Brno, functional sample (2024).

FLORIÁN, V.; KREJČÍŘÍK, P.; SLAVÍČEK, J.; KŘIVOHLAVÝ, P.; ŠKAROUPKA, D.; KOUTNÝ, D.: Structure for positioning of service peripherals. IMID, FME, BUT, Technická 2896/2 616 69 Brno, functional sample (2024).

VESPALEC, A.; HURNÍK, J.; SLAVÍČEK, J.; KRČMA, M.; KOUTNÝ, D.; SIMON, P.; BARTÁK, L.; KREJČÍ, M.: 3D device made by RPT technology designed to guide mat or fabric for open profile V29. GDP KORAL, s.r.o. Za mlýnem 5 666 01 Tišnov, functional sample (2023).

VESPALEC, A.; HURNÍK, J.; SLAVÍČEK, J.; KRČMA, M.; KOUTNÝ, D.; SIMON, P.; BARTÁK, L.; KREJČÍ, M.: 3D device made by RPT technology designed to guide mat or fabric for closed profile SQ50x50x5. GDP KORAL, s.r.o. Za mlýnem 5 666 01 Tišnov, functional sample (2023).

Software (RIV-R)

HURNÍK, J.; KRČMA, M.; VESPALEC, A.; SLAVÍČEK, J.; KOUTNÝ, D.; SIMON, P.: Parametric generator of guide fixtures for the production of closed composite profiles, software (2023).

Utility model (U1)

SLAVÍČEK, J.; PALOUŠEK, D.; JOHN, J.; KOIŠ, J.; KRESA, J.; VIA ALTA a. s.: Print head with nozzle rotation for 3D printing of concrete mixtures. IMID, FME, BUT, VIA ALTA a. s., Patent number: 34847, Utility model (2021).

9 LITERATURE

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Statement on the use of AI tools

During the preparation of this dissertation, specific AI-based tools were used in a supportive role. DeepL was used for translation purposes. Grammarly and ChatGPT were used to improve the language quality, clarity, and stylistic consistency of the text. SciSpace and Ai2 Allen were used to assist with scientific information searching and literature exploration. These tools were not used to generate original scientific content, experimental results, data analysis, or interpretations. The author remains fully responsible for the content, conclusions, and scientific integrity of this work.

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LIST OF ABBREVIATIONS AND SYMBOLS

<i>AM</i>	Additive Manufacturing
<i>DED</i>	Directed Energy Deposition
<i>WADED</i>	Wire Arc Directed Energy Deposition
<i>WAAM</i>	Wire Arc Additive Manufacturing
<i>EBDED</i>	Electron Beam Directed Energy Deposition
<i>BJ</i>	Binder Jetting
<i>GMAW</i>	Gas Metal Arc Welding
<i>GTAW</i>	Gas Tungsten Arc Welding
<i>TIG</i>	Tungsten Inert Gas welding
<i>WIG</i>	Wolfram Inert Gas welding
<i>PAW</i>	Plasma Arc Welding
<i>MIG</i>	Metal Inert Gas welding
<i>MAG</i>	Metal Active Gas welding
<i>CMT</i>	Cold Metal Transfer
<i>CMT-Pulse</i>	Cold Metal Transfer with pulsed current
<i>CMT-Advanced</i>	Cold Metal Transfer with alternating polarity
<i>WFS</i>	Wire Feed Speed
<i>WFR</i>	Wire Feed Rate
<i>TS</i>	Travel Speed
<i>WS</i>	Welding Speed
<i>I_{boost}</i>	Boost phase current
<i>t_{I_{boost}}</i>	Boost phase duration
<i>I_{sc_{wait}}</i>	Burn phase current
<i>I_{sc2}</i>	Short-circuit phase current
<i>vd_{sc_{wait}}</i>	Electrode speed during boost and burn phase

P_c	Average power of a single CMT cycle
T_w	Heat input
η	Heat transfer efficiency
<i>SC phase</i>	Short-circuit phase
A_{wp}	Area of weld penetration
A_{wr}	Area of weld reinforcement
<i>FZ</i>	Fusion Zone
<i>HAZ</i>	Heat-Affected Zone
<i>BTF</i>	Buy-to-Fly ratio
<i>HCP</i>	Hexagonal Close-Packed crystal structure
$\alpha\text{-Mg}$	Alpha magnesium solid solution
$\beta\text{-Mg}_{17}\text{Al}_{12}$	Beta intermetallic phase
<i>RE</i>	Rare Earth elements
<i>UTS</i>	Ultimate Tensile Strength
<i>YS</i>	Yield Strength
<i>EL</i>	Elongation
<i>AZ series</i>	Magnesium–Aluminium–Zinc alloy system
<i>ZK series</i>	Magnesium–Zinc–Zirconium alloy system
<i>AM series</i>	Magnesium–Aluminium–Manganese alloys
<i>AE series</i>	Magnesium–Aluminium–Rare Earth alloys

