

INVESTIGATION OF PROCESS PARAMETERS ON THE QUALITY OF ALUMINIUM COMPONENTS MANUFACTURED THROUGH WIRE ARC ADDITIVE MANUFACTURING (WAAM)

A Thesis submitted to Gujarat Technological University

for the Award of

**Doctor of Philosophy
in
Mechanical Engineering**

by

**Patel Dhaval Bharatbhai
(189999919042)**

under supervision of

Dr. Pujara Akshay Ashvinkumar



**GUJARAT TECHNOLOGICAL UNIVERSITY
AHMEDABAD**

[August - 2026]

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ABSTRACT

Heat accumulation is the primary obstacle to producing dimensionally accurate, microstructurally sound aluminium components by wire arc additive manufacturing (WAAM). As successive layers are deposited, the conductive heat extraction path lengthens and the substrate temperature rises progressively, causing the molten pool to widen, grain structure to coarsen, and mechanical properties to become increasingly anisotropic. Despite growing interest in active thermal management strategies, no published study had applied the same rig, the same ER5356 feedstock, and the same full characterisation suite to a panel of cooling strategies under identical controlled conditions.

This thesis addresses that gap through a systematic experimental investigation of seven interpass cooling strategies for GMAW-based WAAM of ER5356 aluminium alloy thin-wall structures: Natural Cooling (NC), Passive Cooling with substrate-attached heat sinks (PC), Active Cooling at two Layer Edge to Water Distance values (A20: LEWD = 20 mm; A10: LEWD = 10 mm), the novel Tandem Immersion Active Cooling strategy (TIAC), its bidirectional deposition variant (TIAC-BD), and Cryogenic Cooling using liquid nitrogen (LN₂). Each condition was characterised across seven domains. Thermal analysis used K-type thermocouples to record interpass temperature histories and compute the Heat Accumulation Index. Geometric quality was assessed by measuring layer height, total and effective wall width, surface waviness, and deposition efficiency on transverse cross-sections. Mechanical testing included uniaxial tensile testing in four orientations, Charpy V-notch impact testing in two orientations, and Rockwell superficial hardness profiling along the build height. Bulk chemical composition was measured by optical emission spectrometry to verify alloy conformance and quantify Mg content variation across conditions. Microstructural analysis combined optical microscopy, SEM, and EDS to characterise grain morphology, columnar-to-equiaxed transition, and precipitate distribution. Phase analysis by X-ray diffraction quantified lattice parameter shifts and crystallographic texture. Fractographic examination by SEM identified the dominant failure mechanisms on tensile and Charpy fracture surfaces.

The principal novel contribution is the TIAC architecture, which extends the existing Near Immersion Active Cooling concept by combining substrate-level water immersion with a synchronised inter-layer impingement jet. This dual heat-extraction pathway reduces the Heat Accumulation Index by 62.3% relative to NC (from 145.9 to 55.0 °C) while operating under a heat input 2.9 times greater than the Phase 1 immersion-only conditions. TIAC consistently outperforms all single-mode cooling strategies across all characterisation metrics. TIAC-BD, which applies bidirectional deposition under the same TIAC cooling, achieves the lowest surface waviness of all conditions (0.78 mm,

a 68.7% reduction from NC) and the most spatially homogeneous wall, though bulk mechanical properties are equivalent to unidirectional TIAC. LN₂ cooling produces the finest grain structure (42.5 µm) but generates the highest substrate-level heat accumulation (HAI = 170.3 °C) because the cryogen acts only on the deposit surface. This, combined with lower impact toughness and higher property scatter relative to TIAC, limits its practical applicability despite its aggressive cooling potential. A data-driven multi-criteria performance matrix confirms TIAC as the optimal strategy, delivering the best balance of deposit quality, build productivity, and practical implementability for WAAM of ER5356 aluminium alloy.

Keywords: WAAM, ER5356, aluminium alloy, heat accumulation, active cooling, Tandem Immersion Active Cooling, TIAC

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List of Abbreviations

A10	Active Cooling with LEWD of 10 mm
A20	Active Cooling with LEWD of 20 mm
AM	Additive Manufacturing
ASTM	American Society for Testing and Materials
AWS	American Welding Society
AR_i	Impact Anisotropy Ratio (E_{CV}/E_{CH})
AI_t	Tensile Anisotropy Index (UTS_{max}/UTS_{min} across TV, TT, TB, TIN)
AR_t	Tensile Anisotropy Ratio (UTS_{TB}/UTS_{TV})
BSE	Back-Scattered Electron
CAD	Computer-Aided Design
CET	Columnar-to-Equiaxed Transition
CH	Charpy Horizontal (specimen orientation)
CMT	Cold Metal Transfer
CNC	Computer Numerical Control
CSV	Comma-Separated Values
CTWD	Contact Tip to Work Distance
CV	Charpy Vertical (specimen orientation)
DAQ	Data Acquisition
DE	Deposition Efficiency
DED	Directed Energy Deposition
DMLS	Direct Metal Laser Sintering
EBM	Electron Beam Melting
EBSD	Electron Backscatter Diffraction
EDS	Energy Dispersive X-ray Spectroscopy
EWV	Effective Wall Width

FCC	Face-Centred Cubic
FWHM	Full Width at Half Maximum
GMAW	Gas Metal Arc Welding
GTAW	Gas Tungsten Arc Welding
HAI	Heat Accumulation Index
HAZ	Heat-Affected Zone
HCF	High-Cycle Fatigue
HR15T	Rockwell Superficial Hardness (15 kgf load, T-scale)
HV	Vickers Hardness
ICDD	International Centre for Diffraction Data
ISO	International Organization for Standardization
KRL	KUKA Robot Language
L-PBF	Laser Powder Bed Fusion
LCD	Liquid Crystal Display
LEWD	Layer Edge to Water Distance
LH	Layer Height
LN₂	Liquid Nitrogen
MAG	Metal Active Gas (welding)
MAM	Metal Additive Manufacturing
MIG	Metal Inert Gas (welding)
ML	Machine Learning
NABL	National Accreditation Board for Testing and Calibration Laboratories
NC	Natural Cooling
NIAC	Near Immersion Active Cooling
OM	Optical Microscopy
PAW	Plasma Arc Welding
PBF	Powder Bed Fusion
PC	Passive Cooling

PDF	Powder Diffraction File
PSP	Process–Structure–Property
RTC	Real-Time Clock
SD	Secure Digital
SDAS	Secondary Dendrite Arm Spacing
SE	Secondary Electron
SEM	Scanning Electron Microscopy
SLM	Selective Laser Melting
SW	Surface Waviness
TB	Tensile Bottom (specimen orientation, longitudinal, bottom zone)
TC	Texture Coefficient (Harris method)
TCP	Tool Centre Point
TIG	Tungsten Inert Gas (welding)
TIN	Tensile Inclined (specimen orientation, 45°)
TIAC	Tandem Immersion Active Cooling
TIAC-BD	Tandem Immersion Active Cooling with Bidirectional Deposition
TT	Tensile Top (specimen orientation, longitudinal, top zone)
TV	Tensile Vertical (specimen orientation, transverse)
TWW	Total Wall Width
UTS	Ultimate Tensile Strength
WAAM	Wire Arc Additive Manufacturing
WEDM	Wire Electrical Discharge Machining
WFS	Wire Feed Speed
XCT	X-ray Computed Tomography
XRD	X-ray Diffraction
YTS	Yield Tensile Strength (0.2% offset proof stress)

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Chapter 1

Introduction

1.1 Background of Additive Manufacturing

Manufacturing technologies have evolved considerably over the past two centuries. The progression has moved from craft based hand fabrication to mechanised mass production, and more recently to digital systems that emphasise design freedom and material efficiency. Traditional manufacturing processes fall into three broad families. Subtractive processes like turning, milling, grinding, drilling and electrical discharge machining removes material progressively from a larger workpiece. Formative processes like casting, forging, rolling, extrusion and stamping reshapes bulk feedstock with negligible mass change through force, heat, or both. Additive processes, in contrast, build the component by depositing material in successive layers. This approach is exemplified historically by overlay welding and cladding, and in the modern context by the family of layer wise digital fabrication techniques known as additive manufacturing. The three families are compared in Figure 1.1.

While subtractive and formative routes remain the backbone of industrial production, they impose significant restrictions when the demand shifts towards complex geometries, internal channels, lattice and topology optimised structures, mass customization and rapid product development. Subtractive operations on expensive aerospace alloys often exhibit buy to fly ratios of 10:1 to 30:1, with the bulk of the purchased material removed as chip and lost to recycling streams [1, 2]. Casting and forging require dedicated tooling that is expensive to design, manufacture and modify, making them poorly suited to short production runs or iterative design cycles. Forming processes are further constrained by the achievable draft angles, undercuts and minimum wall thicknesses that the available die geometries permit. Conventional manufacturing chains also generate long lead times because the part travels through many separate operations, each requiring its own setup and quality control. These limitations span the

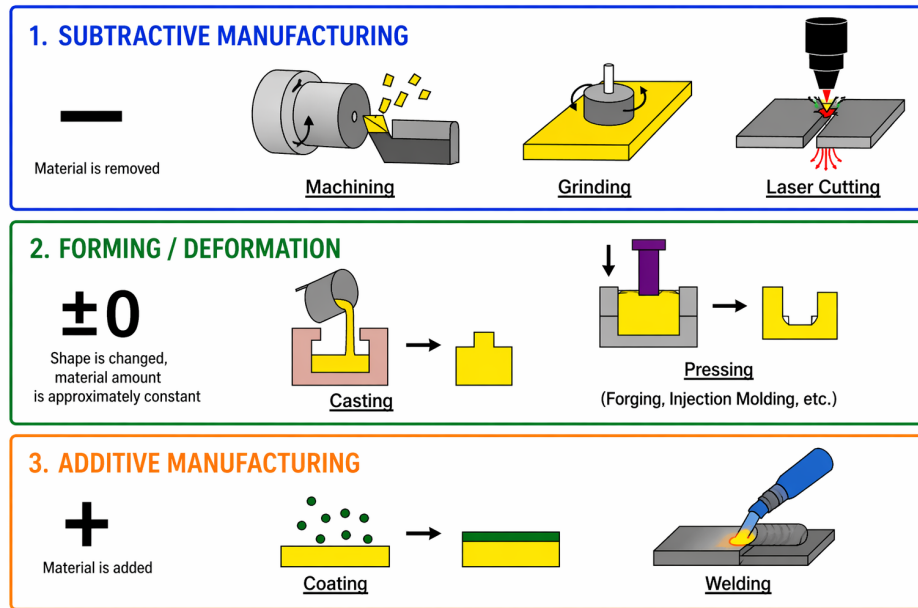


Figure 1.1: Classification of manufacturing processes based on the mode of material transformation: subtractive, formative and additive routes

aerospace, marine, biomedical, energy and defence sectors [3,4].

Additive manufacturing (AM), also known as three dimensional printing, fabricates physical components directly from a digital model by depositing or fusing material layer upon layer until the final part is produced. The process begins with a three dimensional CAD model that is converted into a stack of two dimensional slices through a slicing algorithm; each slice is then translated into machine instructions that direct the energy source and feedstock delivery system to reproduce the slice geometry in physical material. The defining characteristic of AM is that the part is built up rather than carved out, which removes many of the design constraints inherited from subtractive and formative routes [4,5]. Since its origins in rapid prototyping in the late 1980s, AM has matured into a credible production technology for metallic and polymeric components, driven by advances in laser sources, electron beam systems, arc welding electronics and computational toolpath planning.

To bring uniformity to the rapidly expanding additive landscape, the American Society for Testing and Materials, in collaboration with the International Organization for Standardization, formalised a categorisation of AM technologies in the ASTM F2792 and the subsequent ISO/ASTM 52900 standards [6,7]. The classification recognizes seven principal process families, illustrated in Figure 1.2: vat photopolymerisation, in which a liquid photocurable resin is selectively cured by ultraviolet light; material extrusion, in which a thermoplastic filament or paste is extruded through a heated nozzle; material jetting, in which droplets of build material are deposited and cured in a manner analogous to inkjet printing; binder jetting, in which a liquid binder is

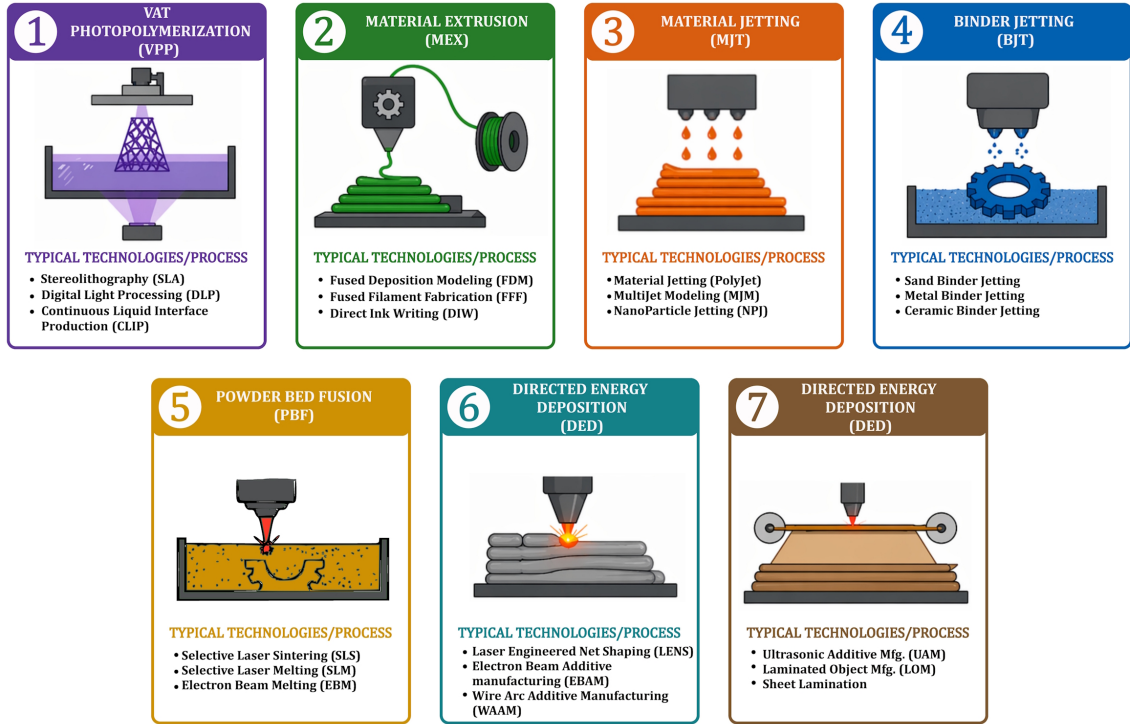


Figure 1.2: ASTM/ISO classification of additive manufacturing process families

selectively applied to a powder bed to consolidate the layer; powder bed fusion, in which a laser or electron beam selectively melts or sinters a layer of metallic or polymeric powder; directed energy deposition (DED), in which a focused energy source melts feedstock supplied in the form of powder or wire at the point of deposition; and sheet lamination, in which thin sheets of material are bonded and trimmed to form the final geometry. Each family encompasses a number of commercial implementations and supports a specific subset of materials, deposition rates, geometric resolutions and economic operating windows.

The advantages offered by additive manufacturing over conventional fabrication routes are well documented and continue to drive its industrial adoption. The ability to fabricate near net shape components directly from a digital model dramatically reduces material wastage, particularly when expensive aerospace alloys such as titanium and nickel based super alloys are involved [3,8]. Tooling requirements are minimised because the geometry is generated directly from the energy source and feedstock delivery; this in turn shortens lead times and enables iterative design refinement at a fraction of the cost of equivalent tool based development. AM permits the fabrication of internal channels, conformal cooling passages, lattice cores and topology optimised geometries that are either impossible or prohibitively expensive to produce by subtractive or formative routes [4,5]. Part consolidation is another important benefit, since assemblies

that would conventionally be produced from many individual parts joined by fasteners or welds can frequently be redesigned as a single AM built component, reducing weight, eliminating leak paths and improving overall reliability. The digital workflow underpinning AM also supports distributed and on demand manufacturing, which is of strategic interest for spare parts supply, defence logistics and remote operations where the inventory cost of a large parts catalogue is prohibitive [9].

Metal additive manufacturing (MAM) represents the most economically significant and technologically demanding subset of AM, because the engineering value of additively produced parts is greatest when high performance metallic components are involved [3,4]. MAM processes are dominated by two broad families: powder bed fusion (PBF), which includes selective laser melting (SLM), direct metal laser sintering (DMLS), and electron beam melting (EBM); and directed energy deposition (DED), which comprises laser, electron beam and arc based variants. PBF offers exceptional resolution and complexity but is constrained by build volume, slow deposition rates, high powder cost and the safety and handling burden associated with fine metallic powders. DED relaxes the build envelope constraint and supports significantly higher deposition rates, making it the preferred route for the fabrication and repair of medium and large scale metallic components [1, 5]. Among DED technologies, wire based variants stand out because the use of welding grade metallic wire as feedstock circumvents many of the cost and safety drawbacks of powder based processes, achieves near complete material utilisation and exploits well established welding infrastructure. Wire arc additive manufacturing (WAAM) has consequently emerged as the most industrially mature wire fed DED route and is the central subject of the present thesis.

1.2 Wire Arc Additive Manufacturing (WAAM)

Wire arc additive manufacturing (WAAM) is a directed energy deposition process in which metallic components are produced layer by layer using a continuously fed metal wire as feedstock and an electric arc as the heat source [2,3]. The process integrates the principles of arc welding, robotic motion control and digital manufacturing to fabricate three dimensional metallic parts directly from a CAD model. A pre programmed toolpath, generated from the sliced geometry, directs a welding torch over a substrate while metallic wire is fed into the arc generated molten pool; the molten droplets solidify on the previously deposited bead to build up a continuous wall, and successive walls and beads are layered upon one another to produce the final three dimensional shape [10,11]. The use of wire rather than powder as the feedstock confers near complete material utilisation (typically in excess of 98%), lower feedstock cost, simplified storage and handling and full compatibility with the welding consumables and qualifications that

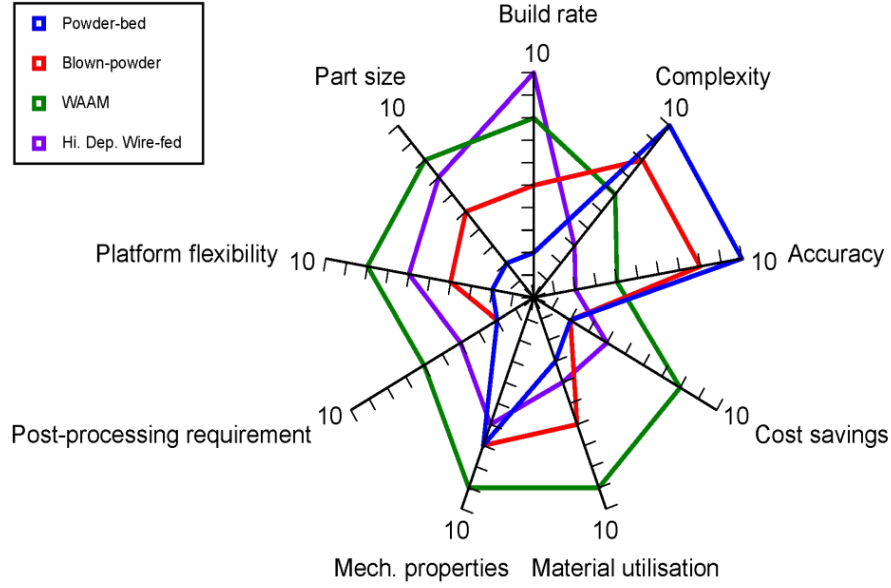


Figure 1.3: Comparative multi attribute process capability of major metal additive manufacturing processes

already exist within metal fabrication industries [12, 13].

The relative position of WAAM within the broader landscape of metal additive manufacturing is best appreciated through a multi criteria comparison with the competing process families. Figure 1.3 consolidates the principal performance attributes of laser powder bed fusion (L-PBF), electron beam powder bed fusion (EB-PBF), laser blown powder directed energy deposition (LP-DED), and WAAM along nine engineering axes: build rate, complexity, accuracy, cost savings, material utilisation, mechanical properties, post processing requirements, platform flexibility and part size. The radial profile of WAAM extends furthest along the deposition rate, material utilization and production flexibility axes, reflecting its high throughput arc based deposition, the near complete consumption of wire feedstock and the unconstrained build envelope offered by robotic or gantry mounted torches. Powder bed processes, by contrast, occupy the high accuracy and high complexity corners of the diagram but pay for this resolution with low deposition rates, limited build volumes and the safety and recycling burden associated with fine reactive powders. This complementary positioning explains why WAAM has consistently been advocated for medium and large scale structural components, while powder bed routes remain the route of choice for small, intricate, or topology optimised parts [3, 13].

The trade off between deposition rate and geometric resolution is examined in greater detail in Figure 1.4, which plots the operating windows of powder bed fusion, laser blown powder DED, conventional WAAM and emerging high deposition rate wire fed processes in a build rate versus layer height plane. Powder bed fusion delivers layer

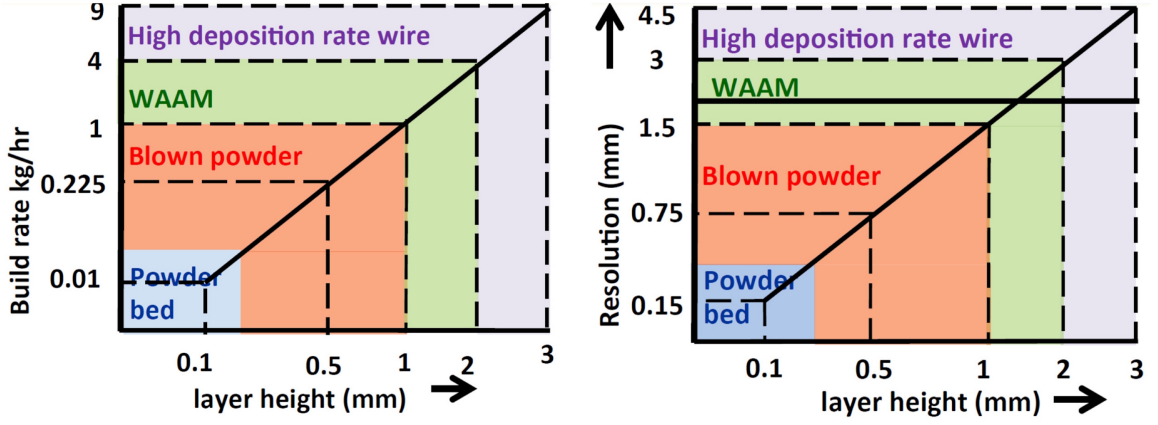


Figure 1.4: Comparison of metal additive manufacturing processes in terms of build rate, layer height and achievable resolution [1]

heights up to approximately 0.1 mm at deposition rates of typically up to 0.01 kg/h, supporting the fabrication of intricate features at the expense of throughput. Laser blown powder DED lifts the deposition rate to the 0.01 to 0.225 kg/h range while increasing the layer thickness to 0.1 to 0.5 mm. Conventional GMAW and CMT based WAAM occupies the high throughput regime at 0.225 to 4 kg/h with layer heights of 0.5 to 2 mm, and recent multi wire and tandem wire developments are pushing deposition rates up to 9 kg/h for large scale structural applications [11, 13]. No single process dominates the build rate resolution plane. The choice of route depends on the size, geometric detail and surface finish requirements of the component.

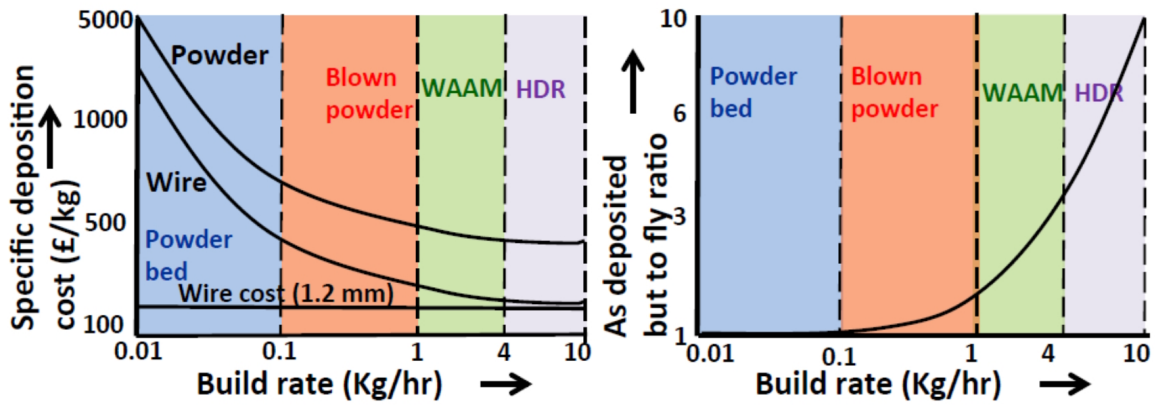


Figure 1.5: Comparison of metal additive manufacturing processes based on build rate, specific deposition cost and buy to fly ratio [1]

The economic case for WAAM, which complements the technical comparison above, is summarised in Figure 1.5. The figure correlates the build rate of each metal additive process with two production relevant cost indicators: the specific deposition cost (the per kilogram cost of as deposited material, including feedstock, consumables and machine amortisation) and the buy to fly ratio (the mass of raw feedstock required per

unit mass of finished part). Powder bed processes typically operate at specific deposition costs ranging from several hundred to around €5,000 per kilogram, dominated by the cost of atomised powder and the slow throughput of the equipment. WAAM, by contrast, exploits inexpensive welding grade wire and rapid deposition to reduce the specific deposition cost to approximately €100 to 150 per kilogram for aluminium and steel feedstocks, while the as deposited build to fly ratio of a WAAM built near net shape preform falls to values in the range of 2 to 5, compared with ratios of 10:1 to 30:1 typical of forged or machined aerospace components [2, 11]. These cost advantages explain the growing industrial adoption of WAAM for aerospace, marine and large structural applications.

A typical WAAM cell comprises an arc welding power source, a wire feed unit, a shielding gas delivery system, a robotic manipulator or computer numerically controlled gantry that positions the torch and a build substrate fixed to a worktable. The robotic motion system reproduces the slice by slice toolpath generated by dedicated path planning software, while the welding power source and wire feeder operate in synchronism to deliver a stable arc and a controlled volume of molten metal at each point along the toolpath [14, 15]. Specialised path planning algorithms have been developed to manage the geometric peculiarities of WAAM, including the requirement that toolpaths minimise start stop events (which are associated with localised defects), maintain a consistent bead width and height across the layer and accommodate the strong influence of inter layer temperature on bead morphology [14, 16]. Interpass cooling or temperature control is normally implemented through dwell times, forced gas convection, water cooling of the substrate, or more recently active cooling immersion strategies, with the objective of preventing progressive heat accumulation that destabilises the molten pool as the wall height grows [17, 18].

1.3 Aluminium–Magnesium Alloy and its Applications

Aluminium occupies a uniquely privileged position among engineering metals owing to its low density ($\rho \approx 2.70 \text{ g cm}^{-3}$), high specific strength when suitably alloyed, excellent corrosion resistance in atmospheric and marine environments and full recyclability [19, 20]. Wrought aluminium alloys are classified into eight series based on the dominant alloying element, as illustrated in Figure 1.6. Among these, the 5xxx series (Al–Mg) stands out for arc based manufacturing because it combines solid solution strengthening, outstanding corrosion resistance in chloride rich environments, and excellent weldability attributes that are critical for WAAM of structural components [21, 22].

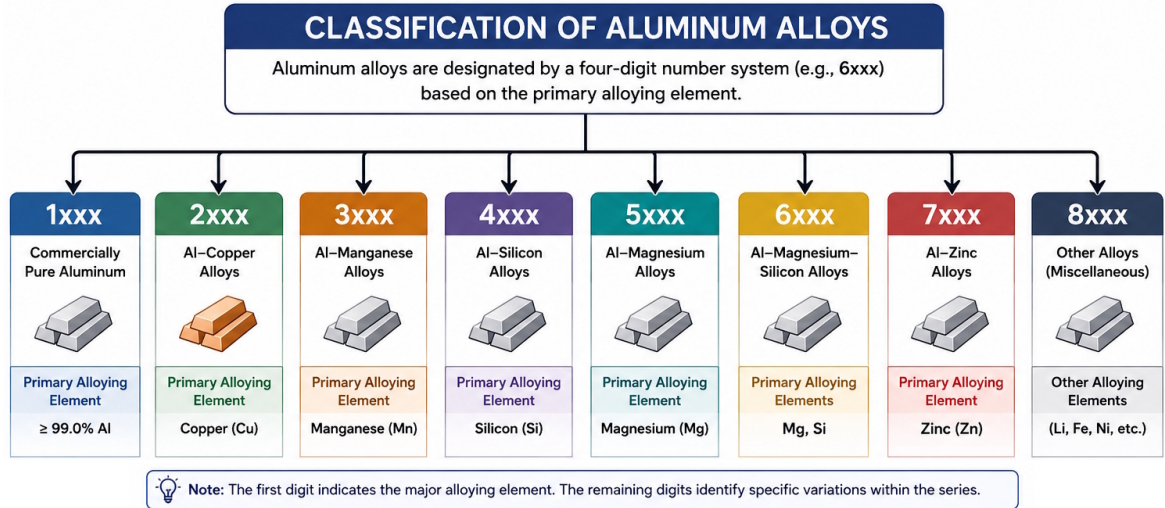


Figure 1.6: Classification of various aluminium alloys according to their number series

Aluminium magnesium (Al Mg) alloys of the 5xxx series typically contain 0.5 to 6 wt.% magnesium together with controlled additions of manganese, chromium, titanium and trace elements that contribute to grain refinement and microstructural stability. Magnesium is the most efficient solid solution strengthener available to aluminium, increasing the yield strength by approximately 15 to 30 MPa per weight percent in the dilute regime while imposing only a marginal penalty on ductility and weldability. In the commonly used welding consumables ER5183, ER5356 and ER5556, the magnesium content lies between 4.3 and 5.5 wt.% and is deliberately matched to the parent metal so that the deposited bead retains both the strength and the corrosion resistance of the surrounding wrought material [23, 24]. ER5356, with a nominal composition of Al 5Mg 0.1Ti, has emerged as the most extensively investigated 5xxx series wire for WAAM owing to its balanced combination of arc stability, mechanical performance, corrosion resistance and commercial availability and forms the basis of the experimental investigation reported in this thesis.

The mechanical properties of Al Mg alloys arise from solid solution strengthening and strain hardening, with the absence of any age hardening response meaning that as deposited properties are governed directly by the WAAM thermal history a practical advantage over the 2xxx and 7xxx families that require post deposition heat treatment [22, 25].

The combined attributes of low density, high specific strength, excellent corrosion resistance and good weldability have established Al Mg alloys and the ER5356 welding consumable in particular, across a broad range of industrial sectors. The principal application domains in which ER5356 deposits and weldments are routinely encountered are summarised in Figure 1.7 and span aerospace, marine, automotive, rail, defence and



Figure 1.7: Various applications of 5356 Al-Mg alloy across aerospace, marine, automotive, rail, defence and general structural sectors

general structural and process industry uses. In aerospace structures, 5xxx alloys are widely used as fuel tank walls, internal stiffeners and fairings and ER5356 is extensively employed for the fabrication of cryogenic propellant tanks and other welded aluminium subassemblies. In marine engineering, 5083, 5086 and 5383 plate joined with ER5356 filler dominate the construction of high speed catamarans, naval vessels, hull plating, superstructures and offshore platform modules, where corrosion resistance in salt water and good formability are decisive advantages [25]. In the automotive sector, magnesium bearing aluminium sheet has become the material of choice for body in white panels in lightweight vehicles particularly for hoods, doors and trunk lids because of its balance of formability, strength and dent resistance, with ER5356 widely used to join these panels. In rail transport, Al Mg alloys are employed for coach body panels, structural extrusions and welded subassemblies that exploit their light weight and durability. In defence applications, 5xxx alloys are encountered in armoured vehicle hulls, ballistic plating, naval structures and military vehicle bodies. Beyond transportation, the alloys are widely deployed in pressure vessels, cryogenic storage tanks for liquefied natural gas, chemical processing equipment, food processing machinery and architectural cladding, in each case exploiting one or more of their core attributes [20, 25].

The convergence of these industrial demands with the capability of WAAM to produce medium and large scale near net shape aluminium components has positioned 5xxx series alloys, and ER5356 in particular, as the most strategically important aluminium feedstock for current and future research in wire arc additive manufacturing.

1.4 WAAM of Aluminium Alloys

The wire arc additive manufacture of aluminium alloys has progressed from an exploratory research topic in the early 2010s to an actively investigated and partially industrialised manufacturing route by the mid 2020s. Early studies focused on demonstrating the feasibility of depositing thin walls in the 2xxx and 4xxx series, with subsequent work expanding into the 5xxx, 6xxx and 7xxx families. The 5xxx series alloys exemplified by ER5183, ER5356 and ER5556 have emerged as the most extensively studied group because their non heat treatable nature, favourable solidification range and excellent weldability render them well suited to the repeated thermal cycling intrinsic to WAAM [21,22].

A key challenge in WAAM of aluminium is heat accumulation: as the wall grows taller, the heat extraction path through the substrate lengthens, interpass temperature rises layer by layer and the molten pool widens and elongates, producing non uniform wall width, side wall sagging and geometric deviation [18,26]. Coupled with this is the susceptibility of aluminium to gas porosity driven by the high hydrogen solubility in the melt and the tenacious alumina film on wire and substrate surfaces which remains the single most pervasive defect in WAAM aluminium components [27,28]. For ER5356, the typical as deposited ultimate tensile strength is 240 to 290 MPa with yield strengths of 110 to 140 MPa, but these values are sensitive to the interpass temperature and cooling history [21,24].

Active cooling strategies that maintain the interpass temperature within a controlled window through forced gas convection, water spray, conductive fixtures, or near immersion in a liquid coolant have been progressively introduced to address heat accumulation and form the central technological contribution of the present thesis. A detailed review of the state of the art in WAAM process optimisation, thermal management and defect mitigation is presented in Chapter 2.

1.5 Major Challenges in WAAM of Aluminium Alloys

Despite the progress described above, WAAM of aluminium alloys remains technically demanding. The challenges arise from the thermophysical and chemical properties of aluminium, the repeated thermal cycling of layer wise deposition and the geometric constraints of large scale arc based manufacture.

The defect landscape encountered in WAAM is not uniform across the engineering alloys investigated to date; rather, each alloy family exhibits a characteristic spectrum

of dominant defects that is dictated by its thermophysical properties, solidification behaviour and chemical reactivity. The correlation between feedstock material and the predominant defect modes observed in arc deposited components is summarised in

Figure 1.8, drawing on the systematic survey reported by Wu et al. [29]. Steels and stainless steels are most commonly afflicted by residual stress, distortion and inter layer oxidation, but generally exhibit limited gas porosity owing to the relatively low hydrogen solubility of iron. Titanium alloys, by contrast, are highly susceptible to oxidation and nitridation in the heated state and require stringent shielding gas control, while coarse columnar prior β grain growth and the resulting Z direction anisotropy are nearly universal observations. Nickel based superalloys present the dual challenge of hot tearing and segregation of refractory elements during the non equilibrium solidification of the arc deposited bead. Aluminium alloys, which are the focus of the present thesis, occupy a particularly demanding position on this map: they accumulate hydrogen porosity at rates that far exceed those of any competing structural metal, they form tenacious surface oxides that disrupt inter layer bonding, and their high thermal conductivity and low melting point conspire to produce severe heat accumulation, side wall sagging and pronounced columnar anisotropy. Several of these defect modes are exacerbated in the magnesium bearing 5xxx series by the additional vapourisation of magnesium under the arc plume [23, 30]. The remainder of this section examines, in turn, the principal defect categories that govern the qualification of WAAM built aluminium structures.

Porosity is consistently identified as the most pervasive defect in WAAM built aluminium components, originating from the large difference in hydrogen solubility between molten and solid aluminium (20:1). Hydrogen sourced from moisture, surface hydrocarbons and hydrated oxides on the wire is rejected at the solidification front and entrapped as gas pores, producing the characteristic banded porosity distribution along layer interfaces that reduces the effective load bearing cross section and acts as the dominant fatigue crack initiation site [12, 27]. The mechanisms and mitigation strategies are reviewed in detail in Chapter 2.

Residual stress and distortion arise from the non uniform thermal expansion and contraction imposed by the repeated heating and cooling cycles; aluminium is particularly susceptible owing to its high coefficient of thermal expansion ($\sim 23 \times 10^{-6} \text{ K}^{-1}$) and large heat affected zone, with residual stresses capable of reaching a significant fraction of the yield strength and causing macroscopic distortion or substrate delamination [31, 32].

Surface finish in WAAM is intrinsically limited by the bead wise deposition, producing a stair stepped side wall geometry with typical as deposited roughness values of 200 to 500 μm R_a , which necessitates post deposition machining for any engineering tolerance

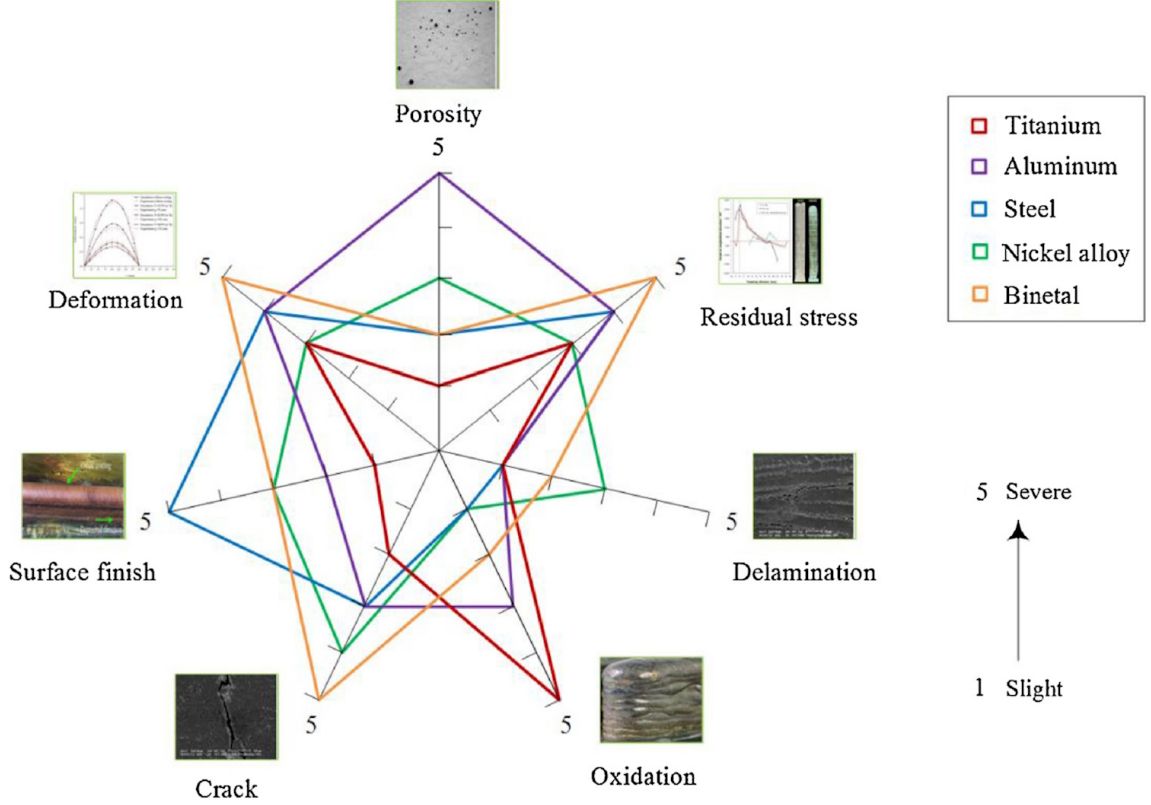


Figure 1.8: The correlation between materials and defects in WAAM processes [29]

surface [13,33].

Heat accumulation couples and amplifies all the foregoing defects: as the wall grows taller the interpass temperature rises progressively, widening the molten pool, coarsening the columnar grain structure, promoting magnesium vapourisation and increasing both porosity and dimensional deviation [18,26]. Managing interpass temperature through active cooling is therefore the central challenge addressed by the present thesis.

Dimensional inaccuracies (typically 0.5 to 2 mm on wall width and height) and mechanical anisotropy (5 to 20% difference in tensile and yield strength between longitudinal and build directions) are direct consequences of heat accumulation and the resulting columnar microstructure, and both are expected to respond to the cooling strategies investigated here [24,26].

1.6 Organization of the Thesis

This thesis is organised into five chapters that progress from the contextualisation of the research problem to the detailed experimental investigation of advanced cooling

strategies for the wire arc additive manufacture of the Al Mg alloy ER5356.

Chapter 1: Introduction establishes the broader engineering context within which the research is positioned. It surveys the evolution of manufacturing technologies, the emergence of additive manufacturing and its ASTM/ISO classification, the principles and variants of wire arc additive manufacturing, the properties and applications of aluminium magnesium alloys and the specific opportunities and challenges associated with WAAM of aluminium alloys. The chapter concludes with the identification of heat accumulation, porosity, anisotropy and dimensional inaccuracy as the principal degradation mechanisms that motivate the present research.

Chapter 2: Literature Review presents a critical review of the literature relevant to the thesis. It surveys the published work on WAAM of aluminium alloys, with particular emphasis on the 5xxx series, and reviews the various thermal management strategies that have been investigated to mitigate heat accumulation, including natural interpass cooling, forced gas convection, water cooling, conductive heat sink fixtures and near immersion and full immersion active cooling. The chapter critically evaluates the metallurgical, mechanical and geometric outcomes reported for each strategy, identifies the specific research gaps that the present work seeks to address, and states the research objectives of the thesis.

Chapter 3: Materials and Methodology describes the experimental methodology adopted in the thesis. It details the selection of the ER5356 feedstock wire and the AA6061 T6 substrate, the configuration of the GMAW based WAAM cell and the design and implementation of the seven interpass cooling strategies investigated, spanning natural cooling, passive water cooling, active immersion cooling at two LEWD values (20 mm and 10 mm), the novel Tandem Immersion Active Cooling (TIAC) architecture, its bidirectional deposition variant (TIAC BD), and cryogenic LN₂ cooling. The chapter describes the process parameters, deposition strategy and specimen preparation procedure for thin wall structures, and details the full characterisation programme covering thermal analysis, geometric measurement, mechanical testing (tensile, Charpy impact and hardness), optical microscopy, SEM/EDS, X ray diffraction and fractographic examination.

Chapter 4: Results and Discussion presents and interprets the experimental findings of the thesis across nine thematic sections. The thermal analysis quantifies the interpass temperature histories, heat accumulation index and water circuit temperatures for all seven conditions. The geometric characterisation reports layer height, total and effective wall width, surface waviness and deposition efficiency and examines the effect of bidirectional deposition. The mechanical properties section presents tensile strength, yield strength and elongation across four specimen orientations, Charpy V notch impact energy in two orientations and Rockwell superficial hardness at five

equally spaced locations along the build height, and discusses the evolution of mechanical anisotropy with cooling intensity. The chemical composition analysis reports the retained magnesium content and elemental distribution across all seven conditions. The microstructural analysis documents the columnar to equiaxed grain transition, porosity and SEM/EDS precipitate characterisation. The phase analysis reports XRD derived lattice parameter shifts and crystallographic texture indices. The fractographic analysis identifies the dominant failure mechanisms through SEM examination of tensile and impact fracture surfaces. The chapter concludes with a comparative analysis of all cooling techniques and a process structure property discussion that links the cooling strategy to the observed thermal, geometric, microstructural and mechanical responses.

Chapter 5: Conclusions and Future Scope summarises the principal findings of the research and draws overall conclusions regarding the effectiveness of the advanced cooling strategies in mitigating heat accumulation and improving the metallurgical, mechanical and geometric quality of WAAM produced ER5356 components. The chapter also outlines the contributions of the thesis and identifies promising directions for future investigation.

Chapter 2

Literature Review

2.1 Introduction

This chapter presents a critical review of the literature relevant to Wire Arc Additive Manufacturing (WAAM) of aluminium alloys, with particular emphasis on heat accumulation and its mitigation through interpass cooling strategies. The review is organised as a funnel: it begins with a survey of WAAM as a process and the hardware sub-systems that constitute a typical experimental rig, narrows to the use of aluminium alloys specifically ER5356 (Al–Mg, 5xxx series) as feedstock, and then focuses on heat accumulation as the dominant process bottleneck in aluminium WAAM. Subsequent sections review the principal mitigation strategies reported in the literature, including process parameter control, deposition path strategies, passive cooling, forced gas cooling, near immersion active cooling (NIAC), and cryogenic cooling using liquid nitrogen (LN₂). A critical comparison of these strategies, supported by tabular synthesis of the reported evidence, leads to the research gap and the research objectives that this thesis seeks to address.

2.2 Wire Arc Additive Manufacturing: Process Foundations and System Architecture

Metal Additive Manufacturing (AM) has matured into a family of processes that includes Powder Bed Fusion (PBF), Directed Energy Deposition (DED) and binder jetting variants [29, 34]. Within DED, Wire Arc Additive Manufacturing uses an electric arc as the heat source and a continuously fed metallic wire as feedstock, depositing material layer by layer to build near-net-shape components [13, 35]. Compared with powder bed processes, WAAM offers an order of magnitude higher deposition rate (typically 1 to 10 kg/h), lower feedstock cost, larger build envelopes and the ability to use

existing welding hardware [35, 36]. These attributes have made WAAM particularly attractive for large structural components in the aerospace, marine, defence and energy sectors [34, 36].

The trade off, however, is intrinsic: the arc deposits a large amount of energy into the workpiece per unit volume, resulting in coarse columnar microstructures, residual stresses, distortion and limited geometric resolution [37, 38]. These features of arc based deposition are the very reasons why thermal management has become a recurring theme in the WAAM literature [39, 40].

2.2.1 Process Variants for WAAM

WAAM is realised through three principal welding processes, each with a distinct interaction with feedstock and heat input [29, 41]. Their comparative attributes are summarised in Table 2.1.

- **Gas Metal Arc Welding (GMAW)** based WAAM, including its Cold Metal Transfer (CMT) variant, uses a consumable wire as both electrode and feedstock. It offers the highest deposition rates and the simplest single torch hardware, making it the most widely used variant for aluminium and steel [42, 43].
- **Gas Tungsten Arc Welding (GTAW)** based WAAM separates electrode and feedstock, allowing finer control of the arc and a cleaner deposit at the cost of lower deposition rate and a more complex wire feed geometry [44, 45].
- **Plasma Arc Welding (PAW)** based WAAM uses a constricted arc and is favoured for titanium and high value alloys due to its concentrated heat source and improved arc stability [2, 46].

For aluminium, GMAW–CMT has emerged as the dominant variant because the controlled short circuit metal transfer reduces heat input, suppresses spatter and helps to stabilise the molten pool of high thermal conductivity Al alloys [42, 47].

2.2.2 System Components of a WAAM Experimental Rig

The hardware architecture of a typical WAAM system, illustrated in Figure 2.1, reflects the integration of welding and motion technologies, and is therefore directly relevant to the first objective of this work (the design of a reliable experimental rig). The principal sub-systems reported in the literature are:

1. **Programmable arc power source** a digitally controlled welding power source (e.g. Lorch, Fronius, EWM or Lincoln) providing synergic current/voltage curves, with optional CMT or pulsed modes for low heat input deposition [42, 57].

Table 2.1: Comparison of WAAM process variants reported in the literature

Variant	Heat input	Deposition rate	Typical alloys	Representative studies
GMAW (conventional)	High	High (3–10 kg/h)	Steel, Ni alloys	[1, 29, 35]
GMAW–CMT (Cold Metal Transfer)	Low–Medium	Medium–High	Al alloys, Steel	[42, 43, 47–51]
GTAW	Medium	Low–Medium	Al alloys, Mg, Steel	[30, 44, 45, 52]
PAW	Medium	Medium	Ti, Ni superalloys	[2, 46]
Variable polarity GMAW	Low	Medium	Al alloys	[53–55]

2. **Wire feeder mechanism** a closed loop wire feeder, generally push pull for aluminium owing to its low column strength, mounted to minimise wire bending and feed disturbances [42, 57].
3. **Motion control system** realised either as a multi axis CNC gantry or, more commonly in current research, as a 6 or 7 axis industrial robot (KUKA, ABB, Panasonic, Yaskawa) coupled with offline path planning software [58, 59].
4. **Shielding gas arrangement** typically pure argon, or Ar–He, Ar–O₂–N₂O mixtures for aluminium, with controlled flow rate to avoid both oxidation and gas entrainment that can cause porosity [54, 60].
5. **Substrate clamping and base plate** rigid clamping is required to resist thermal distortion during long deposits, and pre heating or chilling of the substrate is often used to manipulate the initial thermal boundary condition [38, 61].
6. **Data acquisition (DAQ) system** thermocouples (typically K or N type), pyrometers, current/voltage transducers, and high speed cameras for arc imaging. Open hardware platforms (Arduino, NI–DAQ) are increasingly used to record thermal history at high sampling rates and low cost [57, 62].

These sub-systems together define the experimental envelope within which heat input, deposition geometry and thermal history can be manipulated. A clear understanding of their interactions is a prerequisite to designing the cooling strategies investigated in later chapters.

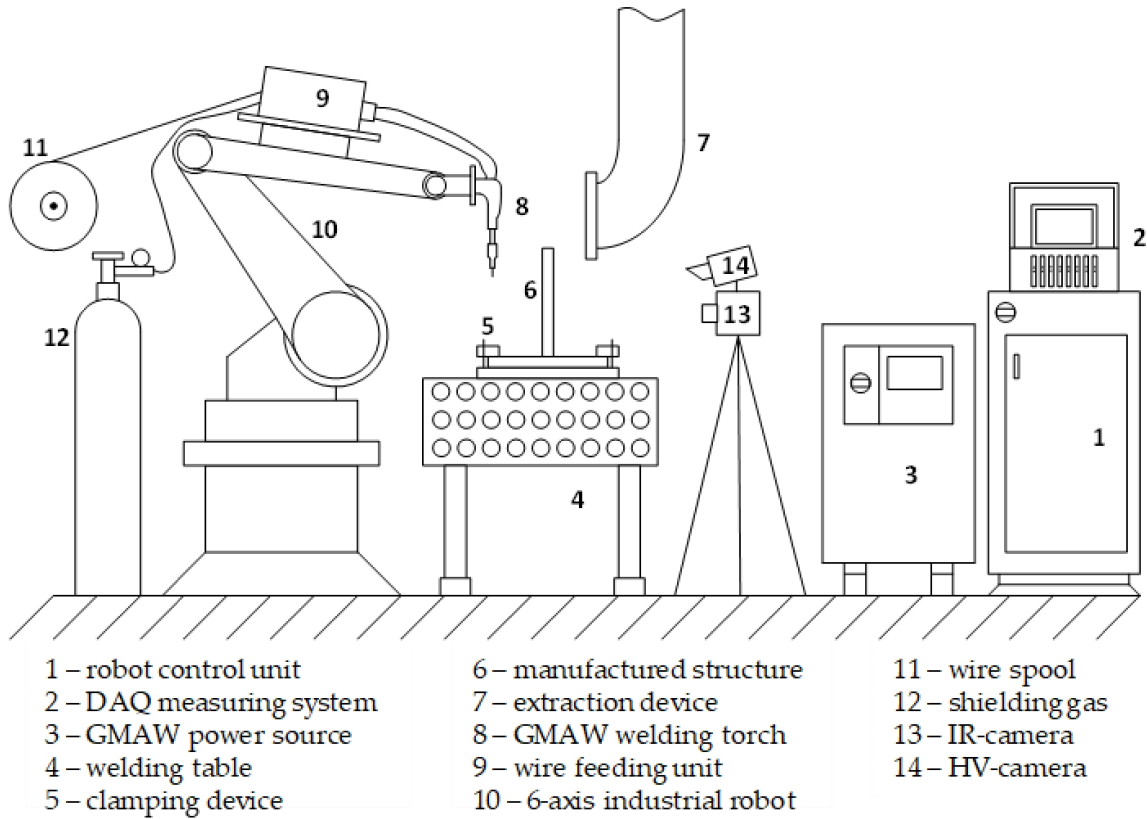


Figure 2.1: Typical Experimental setup of the wire arc additive manufacturing (WAAM) process [56]

2.3 WAAM of Aluminium Alloys with a Focus on ER5356

Among the metallic feedstocks investigated for WAAM, aluminium alloys have attracted particular research interest owing to their combination of low density, high specific strength and the strong industrial demand for lightweight near net shape components in aerospace, marine and defence sectors. Within this family, the 5xxx series Al–Mg alloys (ER5356 in particular) have emerged as the most extensively studied group for arc based deposition, on account of their excellent weldability, non-heat treatable character and commercial availability in standard wire form.

2.3.1 Aluminium Grades Reported in WAAM

Aluminium alloys investigated for WAAM cover almost the entire wrought designation system, with reported studies on the 2xxx (Al–Cu), 4xxx (Al–Si), 5xxx (Al–Mg), 6xxx (Al–Mg–Si), 7xxx (Al–Zn–Mg) and, more recently, scandium modified 5xxx variants [63, 64]. The commercially pure 1xxx and the manganese strengthened 3xxx series, although extensively used in wrought form, have received negligible attention in the

WAAM literature because their low mechanical strength offers no industrial incentive to fabricate them by an arc based route [12, 41]. The 8xxx (Al–Li and miscellaneous) family is similarly absent from mainstream WAAM studies, although isolated work on lithium containing wires has been reported in the context of aerospace structures [19]. The remaining five series therefore define the practical envelope of alloys explored for WAAM, and their relative position in this envelope is a strong function of their inherent weldability, solidification range and susceptibility to gas porosity.

The 2xxx series, dominated in WAAM studies by ER2319 (Al–6.3Cu) and to a lesser extent AA2024 (Al–4.5Cu–Mg), offers the highest as deposited specific strength among the families considered but is notoriously difficult to deposit [63, 65]. Its wide solidification range ($\sim 100^\circ\text{C}$ between liquidus and solidus) and the presence of low melting point Cu rich eutectic constituents promote solidification cracking and inter dendritic hot tearing, while the high hydrogen solubility of the molten phase results in characteristic banded porosity along the layer interfaces [27, 28]. Post deposition solution and ageing heat treatments are usually required to achieve usable strengths in 2xxx WAAM components, which partially erodes the economic advantage of the route [49, 65]. The 7xxx series (Al–Zn–Mg, including AA7075 and AA7050) is even more challenging: its high Zn content evaporates preferentially in the arc plume, its solidification range and shrinkage promote severe hot cracking, and reports of usable thin wall 7xxx WAAM deposits are limited to a handful of feasibility studies in which the alloy chemistry has been deliberately modified by inoculation [66, 67].

The 4xxx series, exemplified industrially by ER4043 (Al–5Si) and AlSi7Mg0.6, is widely used in WAAM because the silicon addition narrows the solidification range, lowers shrinkage, improves fluidity of the molten pool, and largely suppresses solidification cracking [68, 69]. The eutectic Al–Si microstructure provides good arc stability and acceptable as deposited surface finish but offers only modest tensile strength (160–200 MPa) and elongation (5–12%) when used as the sole feedstock, which restricts its application to non-structural geometries or to mixed feedstock strategies in which a 4xxx wire is co-deposited with a stronger consumable [68, 69]. Mixed Al–Si–Mg variants such as AlSi7Mg0.6 partially recover strength by introducing a Mg_2Si precipitation response, but they reintroduce some hot cracking sensitivity [43, 69].

The 5xxx series (Al–Mg, including ER5183, ER5356, ER5556, ER5087 and the Chinese 5A06 grade) combines weldability comparable to the 4xxx series with substantially higher strength and excellent corrosion resistance in chloride rich environments, and has therefore emerged as the most extensively investigated WAAM aluminium family [44, 68]. Magnesium contents in the range 4–6 wt.% provide solid solution strengthening that compensates for the loss of the wrought microstructure during deposition, and the absence of a precipitation hardening response means that the as deposited properties

depend almost exclusively on the deposition thermal history rather than on subsequent ageing treatments [45, 52]. The 6xxx series (Al–Mg–Si), although heat treatable in the wrought form, is generally deposited in WAAM with a 5xxx filler such as ER5356 in order to avoid the cracking sensitivity that the Mg_2Si precipitation reaction introduces under repeated thermal cycling [68, 69].

Recent research has further sought to broaden the WAAM alloy palette by inoculating 5xxx wires with grain refining agents such as scandium, zirconium, titanium or TiC nanoparticles [64, 70]. Scandium modified 5xxx alloys have been shown to undergo a columnar-to-equiaxed transition during arc deposition, with corresponding gains in mechanical isotropy and reductions in hot cracking susceptibility [19, 64]. Titanium powder additions to ER5356 reported by Wang et al. produced Al_3Ti heterogeneous nucleation sites that refined the inter-layer grain structure, increased the longitudinal and transverse tensile strengths and raised the inter-layer micro hardness by 5–10 HV [70]. These approaches remain at the research stage but represent an active frontier for the broader 5xxx WAAM community. Table 2.2 summarises the reported characteristics of all the major alloy families discussed in this subsection in the WAAM context.

2.3.2 Rationale for ER5356

ER5356 is the most widely investigated 5xxx feedstock in the WAAM literature, and its choice as the experimental consumable in this thesis is motivated by a combination of compositional, metallurgical, industrial and methodological considerations. ER5356 is supplied under the AWS A5.10 designation as a wrought Al–Mg welding consumable with a nominal chemistry of Al–5.0Mg–0.12Mn–0.12Cr–0.10Ti and a closely controlled iron and silicon residual content (typically below 0.40 wt.% combined). The magnesium content lies at the upper end of the practical range for hot worked Al–Mg alloys, providing maximum solid solution strengthening without crossing the threshold at which long term β -phase (Al_3Mg_2) precipitation along grain boundaries begins to compromise stress corrosion resistance [30, 45]. The minor additions of Mn, Cr and Ti serve to refine the as cast grain structure, retard recrystallisation during repeated thermal cycling, and improve resistance to inter granular corrosion attributes that carry directly into the WAAM context, where the deposit experiences a thermal history closer to that of multi pass welding than to a conventional cast or wrought product [50, 52].

The mechanical performance of WAAM deposited ER5356 is consistently reported to exceed that of the alternative 4xxx feedstock ER4043 by a substantial margin. Comparative studies have reported as deposited ultimate tensile strengths of 250–290 MPa for ER5356 against 170–210 MPa for ER4043 under broadly comparable deposition conditions, with elongations of 18–30% versus 8–12% respectively [68, 69]. This 30–50% strength advantage, combined with a two to three fold ductility advantage, makes

ER5356 the obvious choice when WAAM is to be used for medium or large scale structural components rather than for non load bearing demonstrators. The corresponding hardness profiles typically 70–85 HV for as deposited ER5356 walls are also more representative of the wrought 5xxx parent grades against which industrial qualification will ultimately be benchmarked [45, 51].

A further methodological advantage of ER5356 for thermal management research is that the Al–Mg system is non heat treatable in the conventional precipitation hardening sense. Variations in the as deposited tensile strength, hardness, elongation and corrosion resistance can therefore be attributed almost entirely to the deposition thermal history interpass temperature, cooling rate, residence time in the partially molten state rather than to a post deposition age hardening response that would otherwise convolve the cooling effect with a separate precipitation kinetics problem [45, 50]. This makes ER5356 ideal for cooling strategy research property changes can be attributed to thermal history without the confound of post deposition precipitation.

Finally, ER5356 is industrially manufactured to a tightly controlled chemistry and surface specification, is widely available at standard wire diameters (1.0, 1.2 and 1.6 mm), and is qualified for use with all major arc welding processes (GMAW, GMAW–CMT, GTAW and PAW). These attributes ensure reproducibility from one experimental batch to the next, allow direct cross comparison with the substantial body of published WAAM data on the same alloy, and ensure that the findings of the present work translate without modification to industrial WAAM users.

Because ER5356 sits at the demanding end of the aluminium WAAM range in respect of magnesium volatility and sensitivity to coarsening, a cooling strategy shown to control heat accumulation in ER5356 may reasonably be expected to control it in less thermally sensitive aluminium feedstocks. Findings obtained on ER5356 do not transfer to the precipitation hardenable systems such as ER2319 or the scandium modified grades, in which the property outcome depends on a precipitation response that the cooling strategy would itself alter, so that the thermal and the precipitation contributions could not be separated in the as deposited condition. This restriction defines the boundary of the present study and is carried forward into the recommendations for future work.

ER5183 belongs to the same non heat treatable aluminium magnesium family as ER5356 and is not excluded by the precipitation argument that rules out ER4043. ER5183 carries a slightly higher nominal magnesium content and additional manganese, and in the wrought condition offers marginally higher strength. Three considerations favour ER5356 over it here. The size of the comparison base: ER5356 is by a wide margin the most reported aluminium feedstock in the WAAM literature, so results obtained on it can be benchmarked against a substantial body of published data, whereas the ER5183 WAAM literature is comparatively thin and would leave the present find-

ings without an external reference. ER5356 being stocked in all common diameters and qualified across the major arc processes, which supports both reproducibility within this work and transferability outside it. The decisive one for a study of cooling, is that the higher magnesium content of ER5183 would increase the sensitivity of the deposit to magnesium loss without introducing any new mechanism into the investigation, so it would make the experiment harder to control without making it more informative. ER5356 therefore represents the most demanding yet best characterised member of the family, which is the combination a cooling study requires.

2.3.3 Reported Process Windows and Properties

The literature on WAAM of ER5356 and other 5xxx feedstocks has converged on a set of broadly consistent operating windows, although the specific values reported vary with the welding process (GMAW, CMT, GMAW-P, VP-GMAW, GTAW), the wall geometry (single bead thin wall, multi bead block, lattice), the substrate condition (room temperature, pre heated, water cooled), and the path strategy. Reported deposition currents fall in the range 80–160 A, wire feed speeds between 4 to 12 m/min, travel speeds between 4 to 12 mm/s, arc voltages between 14 to 22 V and shielding gas (pure argon, or argon rich Ar–He mixtures for high Mg alloys) flow rates between 12 to 22 L/min [44, 68]. Within this envelope, CMT and pulsed-CMT modes are preferred for thin wall structures because their controlled short circuit metal transfer reduces the average current and the corresponding heat input by 25–40% relative to conventional GMAW, while simultaneously suppressing spatter, smoothing the bead surface and mitigating the magnesium burn-off that otherwise depletes the deposit [47, 49].

The shielding gas plays a particularly important role in 5xxx WAAM. Pure argon is the most common choice and provides adequate arc stability and cathodic cleaning, but argon rich Ar–He mixtures (typically 25–50% He) are increasingly reported for thick wall deposits because the higher thermal conductivity of helium broadens the molten pool, improves interpass fusion and reduces the porosity originating from inadequate oxide disruption [12, 71]. Excessive shielding flow, however, induces turbulence in the arc plume that entrains atmospheric moisture and nitrogen and, paradoxically, increases the hydrogen porosity content of the deposit; a balance between 14 and 18 L/min has therefore emerged as the empirical optimum for ER5356 thin walls [50, 71].

Table 2.2: Aluminium alloy grades investigated in WAAM and their reported behaviour

Series	Typical wire	Strength point	Reported issues	Key studies
2xxx (Al–Cu)	ER2319	High strength after HT	Hot cracking, porosity, anisotropy	[28, 49, 55, 65, 72–74]
4xxx (Al–Si)	ER4043, AlSi7Mg0.6	Good weldability, low cracking	Modest strength and ductility	[43, 44, 68, 69]
5xxx (Al–Mg)	ER5356, ER5183, ER5087	Strength + corrosion resistance	Mg loss/burn-off; columnar grains	[30, 45, 47, 50, 51, 60, 75, 76]
5xxx inoculated (Sc/Ti/TiC)	ER5356+Ti, Sc-modified 5xxx, ER5183+TiC	Grain refinement; improved isotropy; CET	Limited industrial availability; cost	[19, 64, 66, 67, 70]
6xxx (Al–Mg–Si)	ER5356 used as filler	Heat treatable	Cracking sensitivity	[41, 68]
7xxx (Al–Zn–Mg)	— (limited)	Highest strength potential	Severe hot cracking; Zn loss	[41, 49]

Table 2.3: Reported deposition windows and as deposited properties of WAAM Al–Mg walls

Wire	Current (A)	WFS (m/min)	TS (mm/s)	UTS (MPa)	Elong. (%)	Ref.
ER5356	90–140	6–9	6–10	250–280	22–30	[30, 45]
ER5356 (CMT)	85–120	5–8	4–8	256–276	25–48	[47, 50, 76]
ER5356 thin wall	80–110	5–7	4–6	245–278	20–29	[60, 76]
ER5183 (CMT / VP–WAAM)	100–140	6–10	6–10	260–290	18–26	[53, 54, 75]
ER5556 (CMT)	90–130	5–9	6–10	270–382	22–26	[51]
ER5087 (rolled)	100–130	6–8	6–8	244–344	20–26	[77]

The effective heat input transferred to the workpiece, calculated as $HI = \eta \cdot V \cdot I/v$ where η is the process efficiency (typically 0.7–0.85 for GMAW and 0.5–0.7 for CMT modes), V is the arc voltage, I the current and v the travel speed, generally lies in the range 0.15–0.45 kJ/mm for stable ER5356 deposition [30, 49]. Heat inputs above this window aggravate side wall sagging, increase the interpass temperature drift up the wall, coarsen the columnar grain structure and reduce the achievable wall accuracy; heat inputs below the window compromise arc stability, increase the risk of lack of fusion defects between layers and produce excessive bead waviness [49, 76]. The interaction between heat input, interpass temperature and bead morphology is the principal lever through which thermal management strategies, including those investigated in the present thesis, exert their effect on the deposit.

Mechanical properties of as deposited ER5356 thin walls reported in the literature fall in a relatively narrow and reproducible band. Ultimate tensile strengths of 240–290 MPa, yield strengths of 110–145 MPa, and elongations of 18–30% are typical for CMT based deposition at moderate heat input, with Z-direction (build direction) values systematically 5–15% lower than X/Y-direction (deposition plane) values, reflecting the columnar grain structure that develops under slow cooling [44, 52]. Inter layer mechanical refinement strategies such as cold rolling, peening or burnishing have been demonstrated to raise the as deposited tensile strength to 320–380 MPa and to reduce the directional anisotropy below 5%, but at the cost of additional process steps and machine complexity [28, 77]. Hardness profiles along the wall height generally fall between 65 and 90 HV, with a gentle decrease from the substrate proximal region toward the upper layers that mirrors the rise in interpass temperature and the corresponding coarsening of the dendritic arm spacing [45, 51].

Among the other Al–Mg consumables included in Table 2.3, ER5183 (nominal Mg 4.3–5.2 wt.%) has been studied principally in variable polarity GMAW (VP–GMAW) configurations, where alternating electrode polarity provides enhanced oxide cleaning and improved inter-layer fusion; Arana et al., Scotti et al., and Derekar et al. report as deposited UTS values of 260–290 MPa and elongations of 18–26%, broadly consistent with ER5356 at comparable heat inputs [53, 54, 75]. ER5556, which carries a higher nominal Mg content of approximately 5.0–5.5 wt.%, delivers a wider UTS range of 270–382 MPa when deposited by CMT, with the upper end achievable only under optimised interpass cooling that suppresses Mg precipitation and retains solid solution strengthening [51]. Both consumables exhibit the characteristic columnar grain, inter layer banded microstructure under natural cooling conditions, confirming that heat accumulation driven microstructural degradation is a common feature of the Al–Mg 5xxx WAAM family and is not specific to any single wire composition [51, 75].

The geometric quality of ER5356 thin walls is governed primarily by the interpass

temperature, the heat input, the path strategy and the active cooling regime. As deposited surface waviness in the range 0.3–1.2 mm peak to valley, effective wall width drift of 5–25% along the build height, layer height reductions of 10–30% from the first to the final layer, and side wall sagging in the upper third of the wall are routinely reported and constitute the dominant geometric defects to be addressed by thermal management [49, 76]. The dominant defect chemistry, hydrogen driven porosity, varies from below 0.1% by volume in optimised CMT deposits to over 3% in deposits subjected to excessive heat accumulation, contaminated wire or unstable shielding [27, 60]. Reported process windows for ER5356 and related Al–Mg consumables, together with the resulting tensile properties, are collated in Table 2.3.

2.4 Heat Accumulation and their effects in Aluminium WAAM

2.4.1 Origin of Heat Accumulation

In WAAM, each deposited layer acts as a transient heat source while previously deposited layers, the substrate and the surrounding fixture act as heat sinks. The directional character of this heat extraction differs fundamentally from that in conventional joint welding: as shown in Figure 2.2, conventional welding allows conduction in multiple dimensions through the parent metal, whereas in WAAM the heat-dissipation path is essentially restricted to one direction, downward through the previously deposited layers and into the substrate [29, 34]. This one dimensional confinement of the heat extraction path is the structural cause of heat accumulation: the heat input per pass remains essentially constant while the available heat sink mass and the conductive cross section through the substrate become progressively less effective at extracting energy as the wall grows [17, 78].

The aluminium–magnesium system is particularly susceptible to this effect on account of its distinctive thermophysical properties. The thermal conductivity of ER5356 ($k \approx 117\text{--}130 \text{ W m}^{-1} \text{ K}^{-1}$ between 25 and 500°C), specific heat capacity ($c_p \approx 900\text{--}1100 \text{ J kg}^{-1} \text{ K}^{-1}$), density ($\rho \approx 2640\text{--}2700 \text{ kg m}^{-3}$), latent heat of fusion ($\Delta H_f \approx 390\text{--}400 \text{ kJ kg}^{-1}$) and a comparatively low solidus–liquidus interval ($575\text{--}640^\circ\text{C}$) jointly produce a system in which energy is rapidly transferred away from the molten pool during the early layers, but in which the rapid drop in cooling rate as the wall grows leads to a disproportionately large rise in the interpass temperature [38, 61]. The cumulative energy input across many layers therefore progressively raises the bulk temperature of the part, and the time spent above the partial melting threshold ($\sim 550^\circ\text{C}$ for ER5356) within the upper layers grows with each subsequent pass. This layer wise accumulation

of heat is illustrated in Figure 2.3, which shows how both the peak and the residual temperature of the deposit drift upward as successive layers are added [78,79]. Quantitatively, published thermal cycle measurements report interpass temperature drifts of 150–300°C above the substrate baseline for natural cooling WAAM walls of 20–30 layers, with the upper layers of long deposits routinely reaching 350–450°C at the start of the next pass when no active cooling is applied [17,40].

The dominant heat rejection mechanisms also shift as the wall grows. In the early layers, conduction into the cooled substrate accounts for the majority of energy removal; as the wall height increases, conduction along the build direction becomes progressively less efficient because the thermal resistance of the wall scales with its length and convection and radiation from the side walls take over as the limiting heat rejection mechanisms. Both convection and radiation are intrinsically slow for the small surface area of vertical thin walls and cannot match the energy delivered by the arc per unit time [39,80]. The interpass temperature therefore drifts upward typically with an exponential like signature in the first 10–15 layers and then with a slower asymptotic rise unless an external thermal management strategy is applied [81,82]. It is this rising interpass temperature, rather than the heat input of an individual pass, that drives the bulk of the defects characterised in the following subsections.

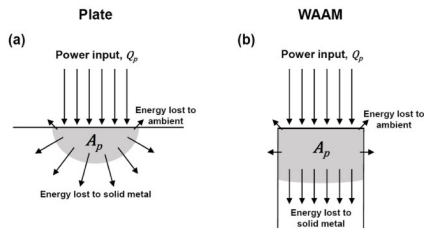


Figure 2.2: Heat transfer in joint welding and WAAM. Thermal conductivity takes place in (a) more dimensions; (b) one direction [34]

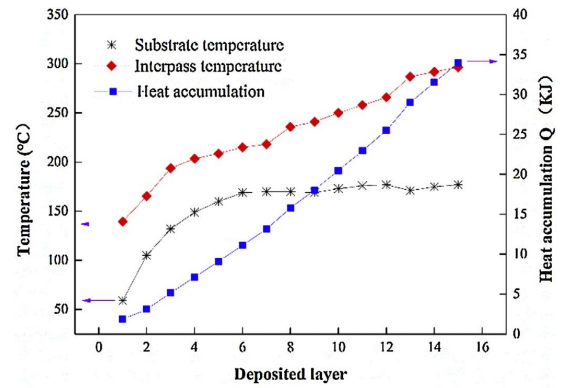


Figure 2.3: The variation in temperature and heat accumulation during fabrication [79]

2.4.2 Effect on Thermal Stability and Arc Behaviour

A rising interpass temperature reduces the temperature gradient between molten pool and substrate, which simultaneously destabilises three coupled aspects of arc operation: arc length, wire melting rate and droplet detachment [57,80]. The dedicated study by Wu et al. on the metal transfer behaviour of WAAM at elevated interpass temperatures demonstrated that, as the substrate temperature rises from room temperature toward 400°C, the arc length grows by 20–35%, the spatial scatter of droplet detachment

points increases by a factor of two to three, and the metal transfer mode itself can drift from controlled short circuit to globular and finally to free flight transfer, with each transition introducing additional process noise [78, 79].

For aluminium, the consequences of this arc destabilisation are particularly severe because the low melting point and high thermal conductivity of the alloy already place the molten pool close to its dimensional stability limit. As the interpass temperature rises, the molten pool widens by 20–40%, becomes shallower and elongates in the deposition direction, with consequent intermittent sagging, reduced reproducibility of bead width and increased layer to layer dimensional scatter as deposition progresses up the wall [80, 82]. The same studies report that the standard deviation of bead width grows from 0.1–0.2 mm in the first few layers to 0.4–0.8 mm in the upper layers under natural cooling. At still higher interpass temperatures, undercut, humping and discontinuous beads have been documented for both pulsed-GMAW and CMT modes on ER5356 and related Al-Mg consumables [47, 49]. The arc destabilisation also alters the spatter rate typically increasing by a factor of 3–5 once the interpass temperature exceeds 250°C and degrades the cathodic cleaning action that is essential for disrupting the tenacious surface oxide on the previous bead, with downstream consequences for inter layer bonding and porosity that are discussed in subsequent subsections [48, 80].

2.4.3 Effect on Bead Geometry and Surface Waviness

Heat accumulation directly affects bead geometry through the coupling between molten pool viscosity, surface tension and the cooling rate of the deposit. As the substrate temperature rises, the time available for the bead to solidify before partial collapse under gravity grows and the layer height (LH) decreases by 10–30% from the first to the final layer of typical natural cooling deposits, while the bead width (BW) increases by 5–25% and the surface waviness (R_a , R_z or peak to valley) grows from 0.1–0.3 mm in the lower layers to 0.3–1.2 mm in the upper layers [62, 82]. The mechanism is well established: at high interpass temperature, the residual heat in the previously deposited bead reduces the cooling rate at the solidification front, the molten pool spreads laterally before solidification arrests its flow and the new bead deposits onto an incompletely solidified surface that itself sags under its own weight.

The geometric consequence is a progressive deviation between the target and the actual wall profile. The effective wall width (EWW) defined as the largest interior width that can be machined from the as deposited wall while removing the side wall waviness is the most widely reported metric for this deviation, and falls from 90–95% of the nominal bead width in the first few layers to 60–75% in the upper layers of long natural cooling deposits [73, 82]. The deposition efficiency, defined as the ratio of net deposited material remaining in the final machined part to the gross material fed into

the arc, drops correspondingly from 90–95% to 55–70%. This loss of material and the associated post deposition machining requirement erode the principal economic advantage of WAAM relative to subtractive routes and have become a dominant driver of active cooling research in aluminium WAAM [83, 84].

Beyond the gross geometric drift, heat accumulation also degrades the local surface finish of WAAM aluminium walls. Documented surface roughness values for as deposited ER5356 walls fall in the range 200–500 $\mu\text{m } R_a$, with the worst readings concentrated in the upper third of the wall where the interpass temperature is highest [76, 85]. Mathematical models that correlate the layer roughness with the deposition parameters and the cooling history including those of Yaseer and Chen using machine learning techniques confirm that the interpass temperature is the single most influential variable, followed by the travel speed and the wire feed speed [82, 85]. The cumulative effect on the geometric fidelity of the part is therefore an inseparable consequence of the underlying thermal history.

2.4.4 Effect on Microstructure

Slower cooling rates at higher interpass temperatures promote (i) coarsening of primary α -Al grains, (ii) the formation of long columnar grains aligned with the build direction due to epitaxial growth, (iii) widening of the secondary dendrite arm spacing (SDAS) from 8–12 μm at the substrate proximal layers to 20–35 μm at the top of the wall, and (iv) coarsening or partial dissolution of β -phase (Al_3Mg_2) particles at grain boundaries [52, 86]. The SDAS variation is itself a sensitive proxy for the cooling rate at the solidification front a coarser SDAS corresponds to slower solidification and therefore more time for solute segregation along the dendritic boundaries and provides a quantitative microstructural fingerprint of the underlying thermal history [30, 87]. The sensitivity of the as deposited microstructure to the linear energy of deposition is evident in Figure 2.4, in which SEM micrographs of Al–Mg deposits at 1440 and 1800 J/cm reveal coarser dendritic arms and more continuous grain boundary networks at the higher energy input. Equivalent observations have been reported for ER5356, ER5183 and ER5556 by independent groups, all of whom found that the columnar-to-equiaxed transition (CET), which is the principal mechanism for breaking the directional grain texture, is progressively suppressed as the interpass temperature rises [45, 52].

The spatial distribution of microstructural features is itself heterogeneous along the build direction. The substrate proximal region, where the cooling rate is highest, generally exhibits a fine equiaxed microstructure with SDAS $\sim$ 8–12 μm and a high density of finely dispersed β -phase particles. The middle of the wall which experiences the most thermal cycles and the most prolonged residence time at elevated temperature

displays the coarsest columnar structure, extensive solute partitioning and the highest density of continuous grain boundary β -phase networks. The top of the wall, which experiences only the most recent thermal cycle before deposition ends, exhibits a partially equiaxed fine grained structure that resembles the substrate proximal region but with intermediate SDAS values [21, 24]. This three zone heterogeneity carries directly into the property variation through the wall height and constitutes one of the most important challenges for the qualification of WAAM aluminium components for safety critical applications. Bands of remelted material at each layer interface, corresponding to regions that have been partially remelted and rapidly resolidified by the subsequent pass, exhibit a locally refined microstructure that interrupts the dominant columnar pattern but does not eliminate the overall anisotropy [45, 86].

Post deposition mechanical refinement is able to partially counteract the heat accumulation driven coarsening: Figure 2.5 compares as deposited WAAM 5087 walls with samples subjected to inter layer rolling at 15, 30 and 45 kN, demonstrating progressive grain refinement and a denser distribution of secondary phases with increasing rolling load [73, 77]. Active interpass cooling, which is the principal subject of the present thesis, similarly reduces the time available for grain growth and SDAS coarsening, with reported reductions of 30–60% in average SDAS and a corresponding shift toward equiaxed morphology in the upper layers of NIAC treated walls [53, 83]. These microstructural features and their gradients along the build direction underpin the directional dependence of mechanical properties characterised in the next subsection.

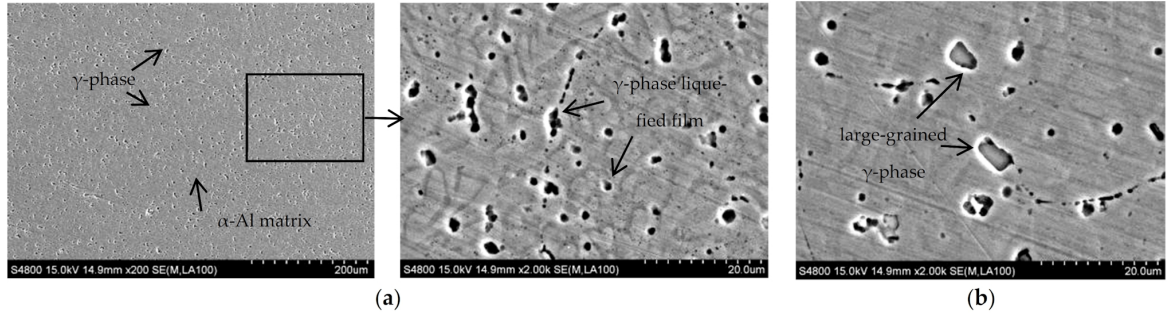


Figure 2.4: SEM microstructure of Al-Mg alloyed additive under different linear energies (a) 1440 J/cm line energy; (b) 1800 J/cm line energy [51]

2.4.5 Effect on Mechanical Properties and Z-Direction Anisotropy

The columnar microstructure produced by heat accumulation translates into a systematic and reproducible mechanical anisotropy. Tensile measurements on as deposited ER5356 walls consistently report ultimate tensile strengths (UTS) of 250–290 MPa along the deposition (X) and transverse (Y) directions but only 220–260 MPa along the

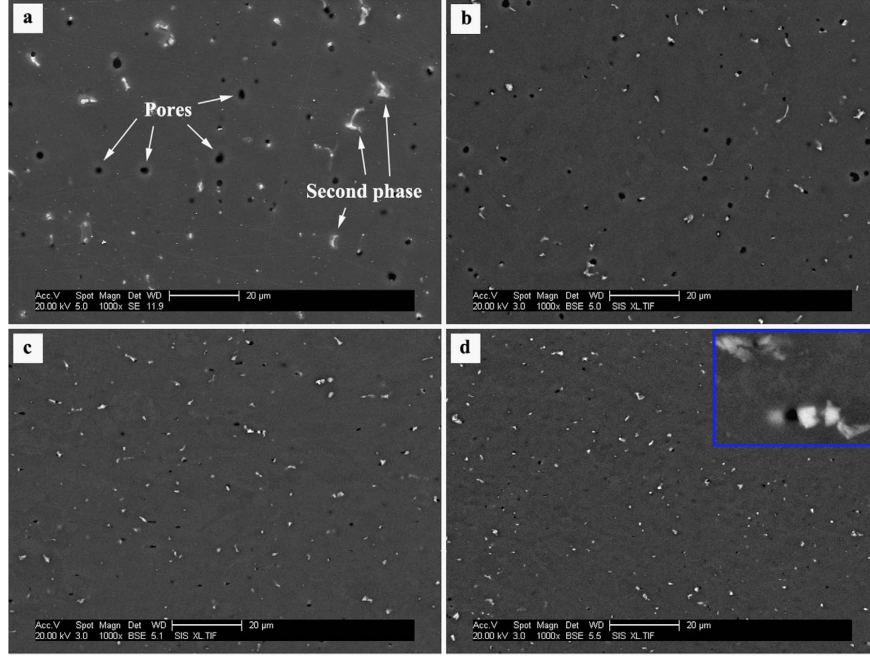


Figure 2.5: SEM images for WAAM 5087 alloys with inter layer rolling in states of (a) as deposited; (b-d) 15 kN, 30 kN, 45 kN inter layer rolled [77]

build (Z) direction, with elongation at fracture values of 22–30% along X/Y dropping to 15–22% along Z [44, 52]. The corresponding yield strengths show smaller absolute differences but the same directional ordering, with X/Y values of 120–145 MPa against Z values of 105–130 MPa. The reduction of 5–15% in tensile strength and the more pronounced reduction of 20–40% in elongation along the build direction reflect the elongated columnar grain morphology, the orientation of the grain boundary β -phase networks perpendicular to the loading axis in Z-specimens, and the preferential alignment of porosity bands at the layer interfaces, all of which act as preferential paths for crack propagation under Z-direction loading [88, 89].

Hardness profiles along the wall height also exhibit a gradient, with lower hardness in the upper layers where the time spent at elevated temperature is longest. Reported Vickers hardness values for as deposited ER5356 fall in the range 65–90 HV, with a typical drop of 8–15 HV from the substrate proximal region to the upper third of the wall [45, 47]. The gradient correlates directly with the SDAS variation described in the previous subsection and with the coarsening of the grain boundary β -phase precipitates. Inter layer mechanical refinement strategies such as cold rolling, hammer peening or burnishing partially flatten this gradient and can raise the average hardness by 25–40%, but they cannot eliminate the underlying compositional and microstructural heterogeneity [90, 91].

The fatigue performance of WAAM aluminium walls is even more sensitive to the heat accumulation driven defect spectrum than is the quasi static tensile response. Fatigue

cracks preferentially nucleate at hydrogen pores, oxide inclusions and the columnar grain boundary networks formed under high interpass temperature, with the result that the high cycle fatigue strength of as deposited ER5356 is typically 60–75% of the wrought 5083 reference value [88, 89]. The strong dependence of the fatigue life on pore size, pore distribution and inter layer oxide entrapment has been quantified for titanium, nickel and aluminium WAAM systems and forms the principal motivation for the porosity mitigation strategies discussed in subsequent sections of this chapter. The combined effect of tensile anisotropy, hardness gradient and degraded fatigue performance therefore makes heat accumulation a first order design concern for any structural application of WAAM aluminium components.

2.4.6 Residual Stress and Distortion

Repeated heating and cooling cycles lock in residual stresses through the constrained thermal contraction of each newly deposited layer: as a layer cools toward the interpass temperature its contraction is resisted by the material beneath it and by the substrate clamping, producing tensile longitudinal stresses in the upper wall balanced by compressive stresses near the substrate. Neutron diffraction and contour method measurements on WAAM aluminium walls report peak longitudinal values of 80–140 MPa, some 60–90% of the as deposited yield strength, concentrated in the upper layers [92, 93]. On unclamping, this drives visible bending distortion of the order of 1–5 mm per 100 mm of wall length, frequently necessitating straightening or additional machining [37, 94]. The magnitude is governed chiefly by the interpass temperature and the deposition sequence, both central variables in the present work, with the high coefficient of thermal expansion of aluminium amplifying the strain per unit temperature swing; interpass rolling, peening and post deposition stress relief are reported as complementary mitigations [2, 90, 93].

2.4.7 Hydrogen Porosity and Hot Cracking

Hydrogen porosity is the most pervasive defect in WAAM aluminium and originates from the roughly twentyfold fall in hydrogen solubility between molten and solid aluminium: hydrogen introduced from moisture in the shielding gas, surface hydrocarbons, or hydrated oxides on the wire is rejected at the solidification front and entrapped as pores, whose size and volume fraction grow with the time spent at elevated interpass temperature [27, 28]. Reported porosity ranges from below 0.1% in optimised deposits to over 3% under excessive heat accumulation, and X-ray computed tomography reveals two populations, fine spherical hydrogen micropores through the bulk and coarser irregular pores banded at the layer interfaces, the latter acting as dominant fatigue crack initiation sites [28, 60, 89]. Hot cracking is the second chemistry driven defect,

but is markedly less severe in the 5xxx family: the moderate magnesium content and favourable solidification range of ER5356 suppress the low melting eutectic films responsible for inter granular cracking under most conditions, although very high heat inputs and elevated interpass temperatures can reintroduce susceptibility [23, 51, 95].

2.4.8 Inter-Layer Bonding Defects and Magnesium Burn-Off

Inter-layer bonding describes the metallurgical continuity between successive deposited layers and is governed by the local energy balance, the cleanliness of the previous bead surface and the dynamics of the molten pool during the next pass. Under stable deposition conditions, the arc remelts a thin layer ($\sim 100\text{--}400\ \mu\text{m}$) of the underlying bead, mechanically disrupts the surface oxide film through the cathodic cleaning action, and produces a metallurgical bond that approaches the strength of the surrounding material. Heat accumulation degrades this bonding by two coupled mechanisms: first, the rising interpass temperature reduces the temperature gradient between the new bead and the previously deposited material, which decreases the depth of remelting and increases the likelihood that the alumina film on the previous bead is incompletely disrupted; second, the elevated dwell time at high temperature promotes oxide growth on the previously deposited surface, increasing the volume and tenacity of the film that must be broken by the subsequent arc [48, 72]. Lack of fusion defects, manifesting as longitudinal cracks along the layer interfaces or, more subtly, as embedded oxide stringers visible only in metallographic cross sections, constitute a documented failure mode for WAAM aluminium walls deposited above the heat accumulation threshold [48, 72].

Magnesium burn-off is a defect category that is specific to 5xxx series WAAM and arises from the high vapour pressure of magnesium at the molten pool temperature. The vapour pressure of magnesium at 700°C is approximately 2–3 orders of magnitude greater than that of aluminium, and at the typical arc plume temperatures of WAAM ($1500\text{--}2500^\circ\text{C}$ in the immediate vicinity of the droplet) magnesium evaporates preferentially from the molten droplet and the surface of the molten pool [23, 72]. Quantitative chemical analysis of ER5356 deposits has reported magnesium losses of 5–15% of the nominal wire composition under conventional GMAW deposition, and the loss is amplified at elevated interpass temperatures because the molten pool dwells for a longer fraction of each pass above the threshold for significant evaporation [23, 24]. The chemical drift away from the nominal Al–5Mg composition has a direct consequence on the strength of the deposit: each 1 wt.% reduction in magnesium content reduces the solid solution strengthening contribution by approximately 15–25 MPa, with secondary effects on the corrosion resistance and on the susceptibility of the deposit to stress corrosion cracking. Cold metal transfer modes substantially mitigate the burn-

off by reducing the average heat input and the dwell time at elevated temperature, and represent one of the principal reasons for the dominance of CMT based variants in 5xxx WAAM [49, 72].

A further consequence of heat accumulation, occasionally observed for very thin walls (single bead, height below 50 mm) deposited without active cooling, is local burn through of the upper layers. When the interpass temperature exceeds the partial melting threshold of ER5356 ($\sim 550^\circ\text{C}$), the deposited bead can lose mechanical integrity under its own weight, leading to local collapse and discontinuous deposition [49, 87]. Although burn through is a less frequent failure mode than the porosity, residual stress and surface quality defects above, it represents the extreme manifestation of unchecked heat accumulation and is one of the strongest demonstrations of the need for active thermal management.

2.4.9 Summary of Heat Accumulation Driven Defects

Interpass rolling [2, 90] and interpass cooling [39, 40] are the two principal mitigation routes reported in the literature for the suite of heat accumulation driven defects described in the preceding subsections, and their detailed evaluation forms the central technical contribution of this thesis. The principal manifestations of heat accumulation, the reported magnitudes of each effect for WAAM aluminium, and the representative studies that quantify them are consolidated in Table 2.4.

2.5 Thermal Management Strategies

This section reviews the principal heat mitigation strategies reported in the WAAM literature in order of increasing process intervention.

2.5.1 Process Parameter Based Control

The simplest and most widely adopted heat mitigation route is to reduce the energy delivered to the workpiece by manipulating the welding parameters themselves, without any modification to the rig hardware. Three classes of parameter based intervention are reported: switching from conventional GMAW to low heat input pulsed or cold metal transfer modes; tuning the current/voltage/wire feed/travel speed combination within a given mode to operate near the lower bound of the stable deposition envelope; and increasing the interpass dwell time to allow the wall to cool between layers [42, 43]. Each acts on a different point in the energy balance of the deposit but all share the same fundamental limitation: they reduce the deposition rate or the build throughput in direct proportion to the heat input reduction.

Table 2.4: Manifestations of heat accumulation in aluminium WAAM and reported magnitudes

Manifestation	Observed effect	Reported magnitude	Representative studies
Interpass temperature drift	T_{ip} rises with each layer	150–300°C above baseline	[17, 40, 81–84]
Arc / metal transfer instability	Arc lengthening, droplet scatter, mode drift	20–35% arc-length growth; 2–3× detachment scatter	[49, 57, 78–80]
Bead widening / sagging	EWV increases up the wall	5–25% width drift	[49, 53, 76, 84, 87]
Surface waviness / roughness	Valleys deepen with wall height	0.3–1.2 mm peak to valley; 200–500 μm R_a	[50, 73, 76, 85]
Layer height variability	LH decreases with height	10–30% reduction	[71, 76, 84, 87]
Deposition efficiency loss	Drop in EWV / net deposit ratio	90–95% $\rightarrow$ 55–70%	[1, 41, 84, 87]
Columnar grain growth	Strong $\langle 100 \rangle$ texture along Z	Aspect ratio 5–15	[45, 51, 54, 86, 87]
SDAS coarsening	Increase with wall height	8–12 $\rightarrow$ 20–35 μm	[30, 45, 51, 87]
Microstructural heterogeneity	3-zone gradient bottom/middle/top	SDAS, β -phase content, hardness all gradient	[21, 24, 30, 71, 87]
Z-direction tensile anisotropy	UTS / elongation lower along build	5–15% UTS; 20–40% elongation	[30, 44, 50, 52, 54]
Hardness gradient	Lower hardness in upper layers	8–15 HV drop bottom $\rightarrow$ top	[45, 47, 51, 76]
Fatigue performance degradation	Lower S–N curve vs wrought; pore initiated cracks	60–75% of wrought HCF strength	[50, 88, 89, 96, 97]
Hydrogen porosity	Pore growth and banding at high T_{ip}	0.1–3% vol. fraction	[12, 28, 54, 60, 65, 75, 86]

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Table 2.4 – continued from previous page

Manifestation	Observed effect	Reported magnitude	Representative studies
Hot / solidification cracking	Inter granular cracks via eutectic films	Rare in 5xxx; severe in 2xxx/7xxx	[27, 65, 75, 95, 98, 99]
Residual stress	Tensile along build direction	80–140 MPa (60–90% of YTS)	[37, 38, 73, 90, 92, 93]
Substrate distortion	Out of plane bending after unclamping	1–5 mm per 100 mm wall length	[38, 73, 87, 94, 100]
Magnesium burn-off	Mg loss from molten pool / arc plume	5–15% of nominal Mg; ~15–25 MPa per wt.%	[12, 23, 24, 30, 48, 72]
Inter-layer bonding defects	Lack of fusion, embedded oxide stringers	Occurs above T_{ip} threshold	[12, 48, 71, 72, 75]
Local burn through	Collapse of upper layers at very high T_{ip}	Thin walls (<50 mm height) most vulnerable	[1, 12, 49, 87]

Cold metal transfer (CMT) is the single most influential parameter based development for aluminium WAAM. By mechanically retracting the wire during the short circuit phase of metal transfer, CMT decouples the metal transfer event from the arc burning and reduces the average heat input by 25–40% relative to conventional pulsed-GMAW at the same wire feed speed, while simultaneously suppressing spatter, smoothing the bead surface and mitigating the magnesium burn off described in Section 2.4.8 [47, 72]. Within the CMT family, the advanced variants CMT-P (pulsed CMT), CMT-ADV (variable polarity, also marketed as CMT-Advanced), and CMT-PADV (pulsed advanced) offer progressive reductions in heat input and additional levers for tailoring the deposit. The systematic comparison by Cong et al. on Al-6.3%Cu reported that CMT-PADV produced the lowest porosity and the most stable bead among the four variants tested, an observation that has been broadly corroborated for Al-Mg alloys by subsequent authors [49, 72]. Variable polarity CMT further reduces the heat transferred to the substrate by adjusting the electrode positive (EP) to electrode negative (EN) ratio: an EN-biased balance reduces the energy deposited in the substrate by 30–50%, although a sufficient EP fraction must be retained for cathodic cleaning of the alumina film [54, 55].

Interpass dwell time is the second principal lever. Empirical studies on ER5356 and other Al-Mg alloys have shown that increasing interpass dwell from 30 s to 120 s reduces peak interpass temperature by 30–50% under natural cooling conditions, but extends the total build time by a factor of two or more [76, 80]. Quantitative correlations between dwell time and the resulting wall geometry have been reported by da Silva et al. and Yu et al., who showed that the EWW drift can be reduced from 25% to below 10% if the dwell is extended to the point at which the wall returns to within 50°C of the substrate baseline between passes [50, 76]. The penalty, however, is severe: at the typical deposition rate of 1–3 kg/h for CMT-ER5356, a doubling of the cycle time effectively halves the throughput, eroding the principal economic advantage that motivates the adoption of WAAM in the first place [50, 84].

Travel speed optimisation, wire feed speed trimming and dual electrode tandem GMAW have also been investigated as parameter level interventions. Increasing the travel speed reduces the heat input per unit length but narrows the stable bead window and increases the risk of humping defects above approximately 12 mm/s for ER5356 [80, 87]. Tandem GMAW with two synchronised wires increases the deposition rate by 60–90% but introduces additional thermal asymmetry that must be managed by the path strategy [101, 102]. Parameter based strategies are a necessary starting point, not a complete solution. They reduce heat input and extend dwell times, but cannot suppress heat accumulation entirely or meet industrial throughput requirements on their own.

2.5.2 Deposition Strategy as a Thermal Control Lever

The path along which the torch moves and the order in which beads and layers are deposited has a strong influence on local heat input and, consequently, on the thermal history of each point in the part. The deposition strategy acts on the thermal history through two coupled mechanisms: the spatial distribution of energy along a layer (which determines whether heat is concentrated at the start, the centre, or the end of the bead) and the temporal sequencing of beads within and between layers (which determines how long each region has to cool before being heated again by the next pass). Both mechanisms can be exploited as a thermal management lever, independently of the absolute heat input [58, 59].

Reported strategies include uni-directional, bi-directional (raster), oscillating, switch-back, contour, hatching and adaptive slicing variants; representative torch path families investigated by Arana et al. for Al-Mg walls are shown in Figure 2.6. For thin wall structures, the alternation of deposition direction layer by layer (commonly called bi-directional or zig zag strategy) is generally found to produce more symmetric thermal fields and reduce wall waviness by 30–50% compared with uni-directional deposition, because the cumulative bias of the start up and shutdown transients is cancelled out from layer to layer [62, 82]. The corresponding reduction in the start up bulge and the end of layer dip is the principal geometric benefit of the zig zag strategy. Switch-back (cyclic forward-backward) deposition, in which each layer is itself constructed from short alternating segments rather than from a single continuous pass, was demonstrated by Yehorov et al. to break the unidirectional epitaxial grain growth, reduce the columnar texture and lower the porosity of aluminium WAAM walls by 20–35% under matched heat input conditions [33, 86]. Hatching and circling deposition strategies have also been explored for Al-Mg alloys, with circling producing thicker walls and reduced anisotropy in 5xxx deposits when the shielding gas envelope is appropriately maintained [54, 55].

For more complex thin walled geometries, dedicated path planning algorithms are required to convert the sliced cross section into a continuous toolpath that minimises start-stop events and maintains a constant bead width and height across the layer. The contour offset methodology proposed by Ding et al. for thin walled WAAM components is illustrated in Figure 2.7, in which the boundary curves are progressively offset inward to generate the deposition pattern, with seam locations alternated between layers to distribute the start-stop defect over the wall surface [35, 58]. Adaptive slicing approaches, in which the layer thickness is varied locally to follow the geometric features of the part, have been shown to support modest overhangs without dedicated support material and to reduce the post processing burden for non-vertical thin walls [58, 59]. Hybrid path strategy and process monitoring approaches that adap-

tively replan the toolpath in response to real time thermal feedback are an active area of research and are reviewed in §2.5.7 alongside the other emerging approaches [82, 85]. Table 2.5 summarises the principal deposition strategies and their reported effects.

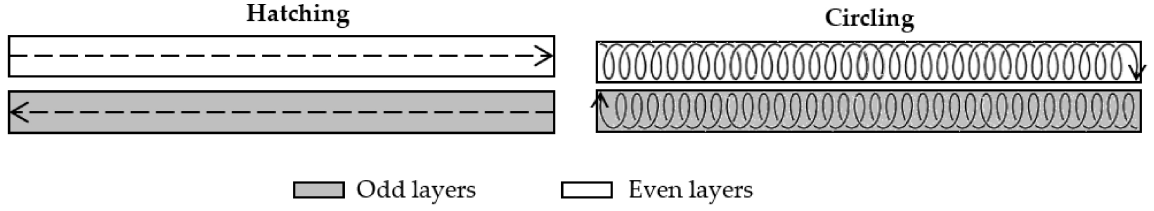


Figure 2.6: Deposition strategies of the study carried by Arana et al. (2021) [54]

Recent work has demonstrated that combining a bi-directional or switchback path with active interpass cooling produces complementary improvements thermal symmetry from the path strategy and reduced bulk temperature from the cooling with notable gains in wall uniformity and reductions in distortion of the order of 25–40% compared with the same path strategy without active cooling [82, 87]. The interaction between path strategy and active cooling for ER5356, however, remains incompletely characterised in the published literature, with most reports addressing either the path strategy or the cooling strategy in isolation. A systematic factorial study that varies both axes simultaneously is one of the explicit objectives of the present thesis.

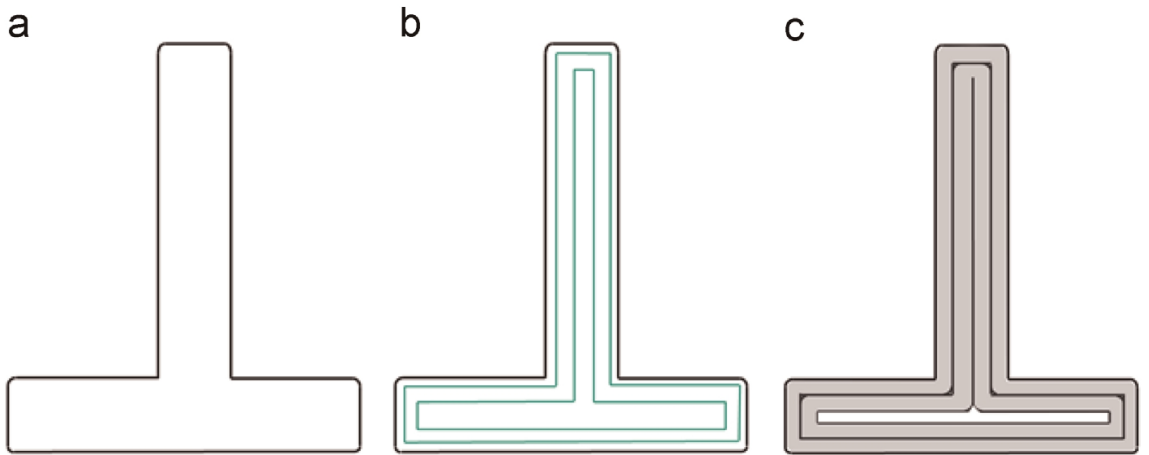


Figure 2.7: Path planning methodology for WAAM of thin-walled structures by Ding et al. (2015) (a) A cross-section of a simple thin-walled structure; (b) contour path patterns generated by offsetting the boundary curves; (c) deposition pattern [35]

Table 2.5: Deposition strategies reported in WAAM and their reported effects on aluminium thin wall components

Strategy	Description	Reported effect	Studies
Uni-directional	Same start, same direction every layer	Unidirectional grain growth; high anisotropy	[35, 71, 86]
Bi-directional (zig-zag)	Reverse direction layer by layer	Symmetric thermal field, reduced waviness	[50, 62, 71, 79]
Switchback / oscillating	Cyclic forward-backward motion	Break unidirectional grains; reduced porosity	[86]
Contour / circling	Closed loop torch path	Thicker walls; lower anisotropy in 5xxx	[54, 55]
Hatching	Parallel raster fill within layer	Suitable for wide walls; porosity sensitive	[54, 55]
Adaptive slicing	Layer thickness varies with geometry	Supports overhangs without support material	[58, 81]

2.5.3 Passive Thermal Management

Passive approaches enhance the heat rejection capacity of the deposit by improving the conductive path from the molten pool to a fixed thermal sink, without requiring any actively pumped coolant flow. The most commonly reported configurations are (i) thick or chilled base plates that act as a large volume heat sink for the early layers; (ii) copper or aluminium backing plates inserted between the substrate and the table to exploit the high thermal conductivity of these materials; (iii) pillar based or finned substrates that increase the effective surface area available for heat rejection to the surrounding air; and (iv) thermally conductive clamping fixtures that draw heat away laterally through the build platform [61, 103].

For aluminium WAAM specifically, the use of a thick aluminium or copper base plate provides a useful baseline mitigation in the first 5–10 layers, with reported interpass temperature reductions of 25–40% relative to a standard thin steel substrate [61, 80]. Pre chilling the base plate to sub-ambient temperatures (typically 0–10°C using a water cooled fixture or a thermo electric Peltier stage) extends this benefit by 5–10°C of additional T_{ip} headroom but introduces condensation related hydrogen pickup concerns analogous to those discussed for cryogenic cooling in §2.5.6 [12, 81]. Pillar substrate

designs proposed by Khan and Madhukar for mild steel walls increase the convective surface area by a factor of three to four and have been shown to reduce the post deposition machining allowance by 30–40%, although direct evidence for ER5356 is limited [81].

The fundamental limitation of all passive approaches is that the heat rejection rate is proportional to the temperature difference between the deposit and the sink, and decreases as the sink itself heats up over the course of a long deposit. Equally important, the effectiveness of passive substrate cooling decays rapidly with increasing wall height as the conductive path from the molten pool to the substrate lengthens beyond the thermal diffusion length of a single deposition cycle: typical 1-D thermal diffusion calculations show that the substrate proximal benefit is essentially exhausted within the first 15–25 mm of wall height for ER5356 deposits [40,80]. Passive measures provide an inexpensive starting point but must be combined with active strategies forced gas, water immersion, or cryogenic cooling to achieve sustained thermal control over a full build.

2.5.4 Forced Gas Cooling

Forced cooling by directed jets of compressed air, CO₂, N₂ or argon has been investigated as a non-contact mitigation that extracts heat from the deposit through convective and, in the case of CO₂, evaporative mechanisms. The principal attraction of forced gas cooling lies in its hardware simplicity, its compatibility with existing WAAM rigs and the absence of any liquid coolant near the live arc [104,105]. A representative in-situ gas cooling (ISGC) apparatus, developed by Ma et al. for carbon steel WAAM, is shown in Figure 2.8; the schematic illustrates how an auxiliary cooling jet is mounted concentrically with the welding torch so that the cooling gas envelopes the freshly deposited bead immediately behind the arc, taking advantage of the highest available temperature gradient for heat extraction.

The choice of cooling gas determines both the available cooling capacity and the chemistry compatibility with the deposit. Ma et al. [40] compared CO₂, N₂ and compressed air in their ISGC configuration on carbon steel and reported that CO₂ delivered the highest heat extraction approximately 60% greater than compressed air at the same volumetric flow together with a 4% improvement in ultimate tensile strength and a refined grain structure attributable to the increased cooling rate at the solidification front. Liu et al. [104] applied follow weld forced gas cooling on a CMT-WAAM nickel-aluminium bronze and reported a 20% increase in yield strength and a doubling of the grain refinement metric. Hackenhaar et al. [39] reported that active interpass gas cooling reduced the thermal distortion of Ti-6Al-4V WAAM walls by 35–50% relative to natural cooling under matched deposition conditions. Montevecchi et al. [105]

demonstrated that air jet impingement positioned 30–50 mm behind the torch could maintain the interpass temperature of a steel wall below 150°C through 20+ layers of continuous deposition, validating the principle of follow weld gas impingement as a continuous operation strategy. For aluminium specifically, the published evidence base is markedly thinner, and the available studies have identified two practical constraints. First, the improvements in interpass temperature for aluminium under forced gas cooling are typically limited to 10–20% relative to natural cooling, well below the 60–85% reductions reported for water immersion strategies, because the convective heat transfer coefficient of a gas jet on a smooth metal surface is intrinsically two to three orders of magnitude lower than that of a water film [40, 82]. Second, gas jets directed at the molten pool or the still hot upper layer can disturb the shielding envelope of the arc, entraining atmospheric moisture and nitrogen and, paradoxically, increasing the hydrogen porosity content of the deposit. The phenomenon has been documented for both ER5183 and ER5356 deposits and represents a real failure mode if the cooling jet is not carefully positioned and timed [48, 54]. Adequately mitigated designs typically position the cooling jet 30–80 mm behind the arc and activate it only during the interpass dwell, accepting the reduced cooling capacity in exchange for arc shielding integrity.

The use of cryogenic gas streams typically vaporised liquid nitrogen or vaporised liquid argon combines forced gas cooling with the very low source temperature of the cryogenic supply, and has been demonstrated by Xian et al. for Ti-6Al-4V to reduce the columnar prior- β grain coarseness and narrow the heat affected zone band to approximately 70 μm [106]. This sub-zero variant is reviewed in §2.5.6 alongside the other cryogenic strategies. For aluminium ER5356, however, the systematic comparison of room temperature versus sub-zero forced gas cooling under identical deposition conditions remains a clear gap in the published literature, with the few attempts to date focused on mild steel or titanium feedstocks [40, 106].

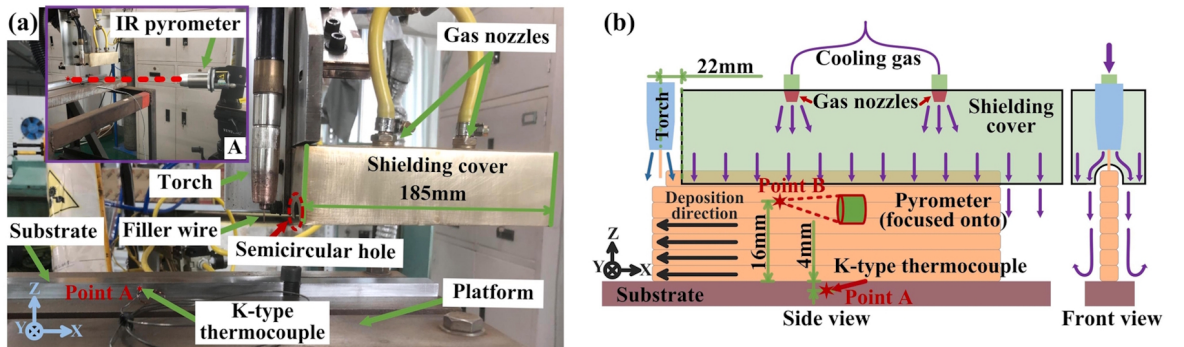


Figure 2.8: (a) Apparatus diagram of ISGC, (b) schematic diagram of ISGC, used by Ma et al. (2021) to carry out investigation of the in-situ gas cooling of carbon steel during WAAM [40]

2.5.5 Liquid Based Near Immersion Active Cooling (NIAC)

Near Immersion Active Cooling (NIAC) was proposed by da Silva and co-workers as a means of dramatically increasing the heat rejection capacity of WAAM by partially immersing the workpiece in a water bath, with only a fixed gap maintained between the water surface and the current deposition layer [83, 84]. The concept exploits the very high convective heat transfer coefficient of water (typically $500\text{--}3000\text{ W m}^{-2}\text{ K}^{-1}$ in still water and rising to $10\,000\text{--}100\,000\text{ W m}^{-2}\text{ K}^{-1}$ under nucleate boiling on a hot wall) to extract energy from the immersed portion of the wall at a rate two to three orders of magnitude greater than that achievable by natural air convection [83, 107]. The concept is compared with the natural and forced cooling baselines in Figure 2.9: in NIAC, the dominant heat rejection path is conduction through the immersed portion of the wall into the surrounding water, Q_{cond} , supplemented by convection Q_{conv} from the wall above the waterline. As deposition proceeds, water is incrementally added to keep the gap referred to as the Layer Edge to Water Distance (LEWD) constant, so that the immersed fraction of the wall grows with each successive pass while the cooling action remains targeted at the high gradient region immediately below the current deposition layer.

The concept and validation paper of da Silva et al. [83] demonstrated, on Al–5Mg single pass multi layer walls, that NIAC achieved an interpass temperature reduction of 60–85% relative to natural cooling, kept the wall temperature essentially independent of wall height, produced virtually constant wall width (EWW drift below 5% over 50 layers), and reduced the mechanical anisotropy from the typical 10–15% Z-direction debit to below 5% all without a measurable increase in the porosity content of the deposit. The thermal cycle measurements reported in that work showed that, under NIAC, the cooling rate at the solidification front was elevated by a factor of two to four relative to natural cooling, with the corresponding refinement of the SDAS and the equiaxed biased grain morphology described in §2.4.4. A follow up study by da Silva et al. [84] addressed the throughput question explicitly and showed that, for ER5356 thin walls, when the same interpass temperature is achieved under natural cooling (using long dwell times) and NIAC (using continuous deposition), the part geometry is comparable but NIAC achieves the same geometry in 50–70% shorter total deposition time. The 200 mm walls deposited under natural cooling and NIAC reported in that study are compared in Figure 2.10, where the NIAC wall exhibits a more uniform width profile and reduced side wall waviness over the full build height.

Scotti et al. [53] combined NIAC with CMT–Advanced (variable polarity) and conducted a factorial investigation that decoupled the contributions of the LEWD parameter and the electrode polarity ratio. They reported that LEWD primarily controls heat accumulation (and therefore wall widening), with the optimum lying between 25

and 40 mm for the geometries tested, while the EP/EN ratio primarily affects layer dimensions and surface waviness, with the optimum being EN-biased for the lowest substrate heat input. This decoupling of the two control variables broadens the optimal parameter window for WAAM and represents one of the most significant practical contributions of the NIAC research stream. Feucht et al. [107] demonstrated a comparable water spray cooling strategy on steel bridge components, achieving mechanical properties equivalent to natural cooling at substantially shorter cycle times and confirming that the throughput quality trade off can be broken by water based cooling on materials other than aluminium. Shi et al. [102] reported on a thermo electric in process cooling configuration for tandem-GMAW deposits that achieved 9–15% increases in the permissible wire feed speed and 42–54% reductions in the required interpass dwell, representing a related but distinct route to the same throughput quality trade off.

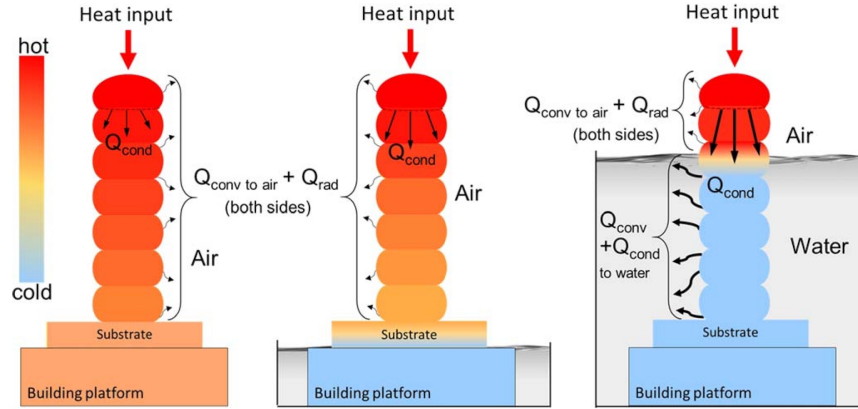


Figure 2.9: Schematic illustration of the NIAC concept versus the natural and forced cooling approaches in WAAM [83]

The principal limitations reported for NIAC are: (i) the cooling action is essentially uniform around the immersed perimeter and cannot be locally tailored to regions of high heat accumulation immediately behind the arc, leaving a portion of the high gradient region of the freshly deposited bead unexploited [83, 84]; (ii) the LEWD is a sensitive process parameter too small a gap (typically below 15–20 mm) induces splatter, steam interference with the arc and contamination of the deposit by water vapour, while too large a gap (typically above 50–60 mm) diminishes the cooling efficiency to levels close to those of natural cooling [53, 84]; (iii) the bath itself is not isothermal, with reported temperature rises of 15–35°C from the start to the end of a long single wall deposit, requiring active recirculation or scheduled exchange to maintain the cooling effectiveness over a multi hour build [53, 83]; and (iv) the immersed wall is exposed to a high humidity environment that can introduce hydrogen pickup if the substrate or wire surface preparation is inadequate, although the published evidence indicates that this risk can be controlled by appropriate pre cleaning protocols [12, 75].

The first limitation NIAC cannot localise cooling to the high gradient region immedi-

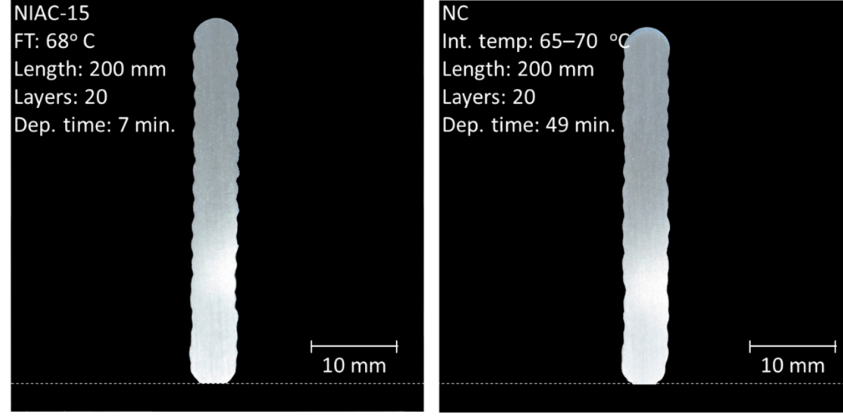


Figure 2.10: Comparison between 200 mm long walls deposited with the NIAC and the NC (natural cooling) approaches [84]

ately behind the arc points directly to the approach taken in this work. An additional, localised jet based mechanism was developed to target this region before the freshly deposited bead equilibrates with the bulk wall. This extension is evaluated and compared against NIAC and LN₂ based approaches in Chapter 4. A summary of reported water based active cooling studies in WAAM is provided in Table 2.6.

Table 2.6: Reported water based / immersion active cooling studies in WAAM

Study	Alloy	Cooling mode	Reported outcome	Ref.
da Silva et al. 2020	Al5Mg (Al-Mg)	NIAC (partial water immersion)	Heat accumulation suppressed; constant wall width; reduced anisotropy	[83]
da Silva et al. 2021	ER5356 thin walls	NIAC vs natural cooling at matched T_{ip}	Equivalent geometry at shorter total time	[84]
Scotti et al. 2020	Al alloy walls	NIAC + CMT Advanced (variable polarity)	LEWD controls heat accumulation; polarity controls waviness	[53]
Shi et al. 2019	Tandem GMAW deposits	Thermoelectric in process cooling	+9–15% wire feed speed; –42–54% dwell	[102]
Feucht et al. 2020	Steel (bridge)	Water spray cooling	Equivalent properties; shorter cycle time	[107]
Khan & Madhukar 2021	Mild steel	Pillar-substrate (passive + air)	Adequate inter-layer cooling; reduced machining	[81]

2.5.6 Cryogenic Cooling Using Liquid Nitrogen and Related Strategies

Liquid nitrogen (LN_2) offers a substantially higher cooling potential than water owing to its very low boiling temperature (-196°C , equivalent to a ΔT of approximately 800°C against an Al–Mg molten pool versus the $\sim 540^\circ\text{C}$ ΔT available in a 100°C water bath) and its substantial latent heat of vaporisation (199 kJ kg^{-1}). The combination of an order of magnitude larger temperature difference and the additional vaporisation enthalpy allows, in principle, heat rejection rates several times those achievable with water immersion at the same coolant flow rate [106, 108]. Three distinct sub-zero strategies have been reported in the WAAM literature: cryogenic vaporised Ar or N_2 shielding/cooling jets directed at the deposit immediately behind the arc; sub-zero interpass quenching, in which the deposit is briefly dipped or sprayed with LN_2 between passes; and direct immersion of the build platform in an LN_2 bath analogous to the NIAC concept but with the bath replaced by a cryogenic medium [106, 108].

The most extensively reported approach is the cryogenic gas variant. Xian et al. [106] demonstrated that cryogenic vaporised Ar shielding and cooling of a Ti–6Al–4V WAAM wall reduced the columnar prior- β grain coarseness by 40–60%, narrowed the heat affected zone band to approximately $70 \mu\text{m}$, and improved the isotropy of the tensile properties between the deposition and building directions. The proposed mechanism is that the cryogenic gas envelope around the bead suppresses the time spent in the β -phase field after solidification, freezing in a finer prior- β structure that translates into a more uniform $\alpha + \beta$ lamellar microstructure on subsequent cooling. Cottam et al. [108] characterised the dislocation density variation under different cooling regimes for Ti–6Al–4V WAAM and confirmed that the cryogenic gas treatment produced the highest dislocation density and the corresponding hardness improvement among the cooling strategies tested.

For aluminium WAAM, the published evidence base on cryogenic cooling is markedly thinner and presents a set of distinct challenges that are not present in titanium. The first concern is thermal shock cracking: the very large temperature gradient between the molten pool and the LN_2 cooled wall, combined with the high coefficient of thermal expansion of aluminium ($\sim 23 \times 10^{-6} \text{ K}^{-1}$, approximately 2.5 times that of titanium), can in principle induce contraction strains large enough to nucleate cracks at grain boundaries or oxide inclusions [65, 106]. The second concern is hydrogen pickup due to atmospheric moisture condensation on the cooled wall: any moisture deposited on the surface from the surrounding air is dissociated by the next arc pass and contributes additional hydrogen to the molten pool, with direct consequences for the porosity content of the deposit as described in §2.4.7 [27, 75]. The third concern is the film boiling

regime: when a cryogenic liquid contacts a metal surface above the Leidenfrost temperature, a vapour film forms between the liquid and the surface and dramatically reduces the heat transfer coefficient, so that the apparent cooling capacity of LN₂ can be much lower than its theoretical value in the first 5–10 s of each cooling cycle [41, 106]. Finally, the operational and safety considerations associated with handling LN₂ during multi hour deposits asphyxiation risk in enclosed spaces, cold burns, embrittlement of fixture materials and the recurring cost of cryogen replenishment constrain the industrial adoption of the strategy and explain why systematic studies for ER5356 remain limited.

A controlled comparison of LN₂ based interpass cooling against water based active cooling under identical deposition conditions, addressing precisely these aluminium specific concerns, is one of the explicit objectives of the present thesis. Table 2.7 summarises the principal cryogenic and forced gas cooling studies in WAAM that inform the design of the LN₂ fixture used in this work.

Table 2.7: Reported cryogenic and forced gas cooling studies in WAAM

Study	Alloy	Cooling medium	Reported outcome	Ref.
Xian et al. 2023	Ti-6Al-4V	Cryogenic vaporised Ar	Narrow HAZ band; isotropic tensile gains	[106]
Hackenhaar et al. 2019	Ti-6Al-4V	Active interpass cooling (gas)	Reduced thermal distortion	[39]
Ma et al. 2021	Carbon steel	In-situ gas (CO ₂ , N ₂ , air)	CO ₂ best heat extraction; +4% UTS	[40]
Liu et al. 2024	NAB (Cu alloy)	Follow weld forced gas cooling	+20% YTS; grain refinement	[104]
Cottam et al. 2019	Ti-6Al-4V	Multi process comparison	Different dislocation density vs cooling rate	[108]

2.5.7 Hybrid and Emerging Approaches

Hybrid approaches combine two or more of the mitigations reviewed above to capture complementary effects, typically thermal control from one mechanism together with microstructural refinement or residual stress redistribution from another. The most developed is interpass cold rolling, applied by vertical rolling of the top layer or by side rolling of the wall flanks as shown in Figure 2.11, each imposing a distinct strain field on the deposit. Inter layer rolling of aluminium magnesium WAAM walls has been

reported to raise hardness, yield stress and tensile strength substantially and to reduce side wall waviness, and when combined with active cooling the benefits are additive, the cooling lowering the interpass temperature while the rolling refines the columnar grains that cooling alone cannot fully suppress [73, 77, 90]. Mechanical peening, in ultrasonic, hammer, shot and laser shock variants, introduces compressive surface stresses and refines the grain structure through severe plastic deformation, and has been reported to reduce hydrogen porosity and improve high cycle fatigue strength in aluminium WAAM, at the cost of cycle time and a line of sight requirement [74, 91].

Other emerging strategies pair thermal control with metallurgical conditioning or with real time process control. In process induction pre heating combined with post layer water cooling delivers a more uniform thermal history through the wall height, at the expense of added rig complexity and the energy cost of operating two thermal sources [17, 105]. A distinct class relies on closed loop monitoring, using infrared thermography, arc imaging, current and voltage transient analysis and machine learning pattern recognition to detect incipient defects and adaptively adjust the heat input, dwell or cooling parameters [62, 85]. These approaches show clear promise but increase system complexity and have rarely been demonstrated for aluminium ER5356 thin wall components, so the gap between demonstrated potential and industrial qualification remains a recurring theme in the recent review literature [11, 50].

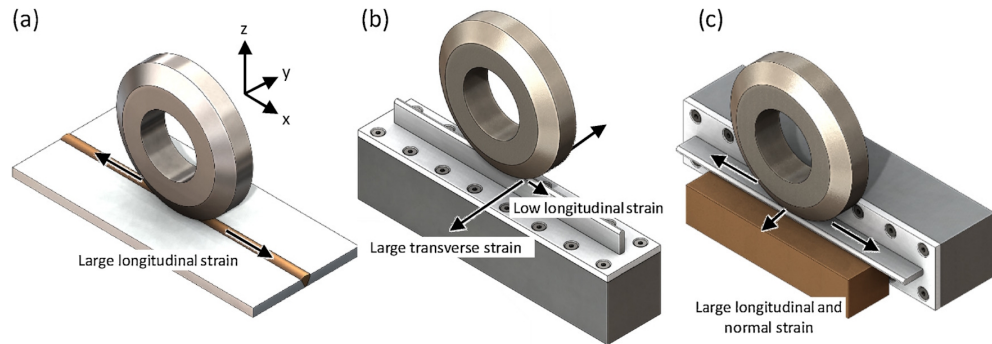


Figure 2.11: Main directions of strain induced by cold rolling: a) vertical rolling of butt welds; b) vertical rolling and c) side rolling of WAAM wall [73]

2.6 Comparative Assessment of Existing Mitigation Approaches

A consolidated comparison of all the heat mitigation strategies reviewed in this section covering the alloys studied, the reported reductions in interpass temperature, the strength of the aluminium–magnesium specific evidence base, the principal benefit and the principal limitation of each approach is presented in Table 2.9.

The reported reductions in interpass temperature have converged: process parameter control and passive heat sinking cluster in the 25 to 50% band, forced gas cooling in the 10 to 20% band for aluminium, and liquid immersion in the 60 to 85% band [39,61,83]. The choice of cooling medium, rather than the refinement of any single strategy within a medium, therefore sets the achievable ceiling.

This convergence in thermal performance is not matched by a convergence in reported mechanical outcomes. Studies reporting comparable interpass temperature reductions report as deposited tensile strengths differing by more than 50 MPa and elongations differing by more than 10% points [68,69,84]. This dispersion indicates that the interpass temperature alone is an incomplete descriptor of the thermal state, and that the location at which heat is extracted, together with the cumulative time the deposit spends above the coarsening range, are the variables that actually govern the metallurgical result. A cooling strategy cannot therefore be selected on the reported interpass temperature reduction alone.

The recent aluminium specific literature remains dominated by single strategy studies executed on differing rigs, with differing feedstock lots and differing characterisation suites, so that the reported numbers cannot be placed on a common axis [34,41]. No study reports a within study comparison of passive, immersion and cryogenic cooling on one rig for one alloy, and it is this comparison that the present work provides.

A critical reading of the foregoing literature reveals a number of consistent shortcomings that constrain the practical application of existing thermal management strategies to aluminium WAAM, and to ER5356 in particular.

First, the majority of interpass cooling studies have been performed on titanium and steel rather than aluminium [104,108]; the high thermal conductivity, high reflectivity to arc radiation and tendency for hydrogen porosity of aluminium make the direct transfer of these findings to ER5356 unreliable.

Second, NIAC the strategy with the strongest published evidence base for aluminium delivers a uniform, perimeter wide cooling action that cannot be tuned to localised hotspots or to differential cooling between the freshly deposited layer and the bulk wall [53,83]. There is therefore an opportunity to extend NIAC with an additional, localised mechanism that targets the high gradient region immediately behind the arc.

Third, the interaction between deposition strategy (path, sequence, dwell) and active cooling has only been partially studied for aluminium [55,79]. In particular, it is not clear whether bi-directional, switchback or circling strategies that produce more symmetric thermal fields without cooling retain their advantage when superimposed with high cooling rate strategies such as NIAC or LN₂.

Fourth, LN₂ cooling of aluminium WAAM is reported only in isolated studies, with no

systematic comparison against water based active cooling under identical deposition parameters [41, 106]. Questions about thermal shock cracking, brittle zone formation and grain coarsening at very low cooling temperatures remain open for aluminium.

Fifth, almost no published study has applied the *same* experimental rig, the *same* ER5356 feedstock and the *same* characterisation suite (thermal, geometric, mechanical, microstructural and fractographic) to a panel of cooling strategies under identical deposition conditions [34, 41]. This absence of a comparative framework is the single most important gap because it prevents an evidence based selection of cooling strategy for a given application.

Taken together, these five shortcomings mark the boundary of what the current literature has established. Liquid immersion is the most capable cooling medium so far demonstrated for aluminium WAAM, and the approximate parameter window within which immersion is stable is known. It remains unestablished whether the cooling action can be localised to the high gradient region immediately behind the arc, whether a cryogenic medium confers any advantage over water once the location of heat extraction is accounted for, or how any of these strategies rank against one another when the rig, the feedstock and the characterisation protocol are held fixed. These questions are addressed by the research gaps stated in Section 2.7 and by the objectives that follow.

A condensed mapping of these critical shortcomings to specific reported evidence is presented in Table 2.8.

Table 2.8: Mapping of critical shortcomings in the WAAM thermal management literature to specific reported evidence

Critical shortcoming	Evidence in literature	Representative studies
Imbalance toward Ti/Steel over Al–Mg	Most active cooling studies are on Ti/Steel; aluminium underexplored	[39, 40, 104, 106, 108]
Uniform but non-localised NIAC cooling	NIAC reduces bulk T_{ip} but cannot target hotspots; LEWD only global control	[53, 83, 84]
Path strategy and active cooling treated separately	Path/strategy studies and cooling studies rarely overlap for Al alloys	[50, 55, 71, 79, 86]
Limited LN_2 data on aluminium	Cryogenic studies dominantly on Ti; few on Al–Mg	[41, 106]
Absent comparative framework	No published study uses identical rig + feedstock + characterisation across cooling strategies	[1, 29, 34, 41]

Table 2.9: Comparative summary of heat mitigation strategies reported in WAAM literature

Strategy	Alloys studied	T _{ip} reduction	Al–Mg evidence	Key benefit	Principal limitation	Ref.
Increased inter-pass dwell	Steel, Al, Ti	30–50%	Strong	Simple; no hardware change	Long build time; partial mitigation	[12, 49, 80, 84]
Pulsed / CMT (low heat input)	Al, Steel	25–40%	Strong	Reduced HAZ, less spatter, less Mg burn-off	Lower deposition rate	[42, 43, 47, 48, 53, 72]
Variable polarity CMT	Al alloys	30–50% (substrate)	Moderate	Reduced substrate heating	Sensitive to EP/EN ratio	[53–55]
Deposition path optimisation	All	0–15%	Moderate	Symmetric thermal field; reduced anisotropy	Limited stand alone effect	[50, 71, 79, 82, 86]
Passive substrate cooling	Al, Steel	25–40% (first 5–10 layers)	Limited	No moving parts; cheap	Decays with wall height	[17, 61, 80, 81]
Forced gas cooling	Ti, Steel, NAB	10–20% (Al); 30–45% (Ti)	Limited	Non-contact, easy integration	Low cooling capacity; disturbs shield	[39, 40, 104, 105]
NIAC (water immersion)	Al–Mg, ER5356, Steel	60–85%	Strong	High cooling rate; reduced anisotropy; faster cycle	Uniform cooling only; LEWD sensitive; bath heating	[53, 83, 84, 107]
In-process induction + water	Al, Steel	Configurable	Limited	Uniform thermal history	Two thermal sources; energy cost	[17, 53, 105]

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Table 2.9 – continued from previous page

Strategy	Alloys studied	T _{ip} reduction	Al–Mg evidence	Key benefit	Principal limitation	Ref.
Cryogenic / LN ₂ / cryo–Ar	Ti, limited Al	Theoretically very high	Present thesis	Very high cooling capacity; HAZ narrowing	Thermal shock; H pickup; safety; cost; limited Al data	[41, 106, 108]
Hybrid (rolling + cooling)	Ti, Al–Mg, Steel	30–50%	Moderate	Microstructure refinement + low T _{ip}	Complex rig; limited Al data	[2, 73, 77, 90, 91]
Peening (US / hammer / LSP)	Ti, Al	N/A (post layer)	Moderate	Compressive surface stress; reduced porosity	Cycle time; line of sight	[74, 91]
Closed loop monitoring + ML	Ti, Al, Steel	Configurable (adaptive)	Emerging	Real time defect avoidance	Integration with welder controller	[11, 13, 82, 85, 87]

2.7 Research Gap

The critical analysis presented in Section 2.6 can be distilled into the following research gaps. Each gap is framed so that it maps directly onto one of the research objectives stated in Section 2.8.

1. **Gap-1: Need for an integrated WAAM rig dedicated to thermal management research.** Most published thermal management studies use ad-hoc retrofits to commercial systems. A dedicated WAAM rig integrating a programmable GMAW source, robotic motion, controlled wire feeding, shielding gas and a synchronised data acquisition system is required to support systematic comparison of cooling strategies under identical conditions.
2. **Gap-2: Inconsistent baseline data on thin wall ER5356 components.** Although ER5356 is the leading 5xxx WAAM consumable, baseline deposition parameters and thermal histories for thin wall vertical structures are reported inconsistently across the literature, hindering benchmarking and study to study comparison.
3. **Gap-3: Absence of a localised cooling mechanism complementary to perimeter immersion.** NIAC delivers uniform cooling around an immersed perimeter but cannot target the high gradient region immediately behind the arc. A strategy that combines perimeter immersion cooling with an additional localised, jet assisted cooling mechanism has not been reported in the literature for aluminium WAAM.
4. **Gap-4: Insufficient coupling of deposition strategy with active cooling.** The interaction between path/sequence strategies (uni-directional, bi-directional, switchback, oscillating) and interpass active cooling has not been systematically characterised for ER5356.
5. **Gap-5: Limited and inconsistent data on LN₂ interpass cooling of aluminium WAAM.** A controlled investigation of LN₂ based interpass cooling benchmarked under identical conditions against water based active cooling approaches is required to clarify the trade offs between cooling capacity, thermal shock risk and resulting material properties.

These five gaps motivate the objectives presented in the following section.

2.8 Research Objectives

The primary aim of this research is to develop and evaluate advanced interpass cooling strategies to mitigate heat accumulation in WAAM of ER5356 aluminium alloy, and

to establish a comparative understanding of their effects on thermal, geometric, mechanical and microstructural properties. Building on the research gaps identified in Section 2.7, the specific objectives of this thesis are:

1. To design and develop a reliable WAAM system for experimentation by integrating key components such as a programmable welding power source, wire feeder mechanism, motion control system (CNC/robotic arm), shielding gas arrangement and data acquisition system. (*Addresses Gap-1.*)
2. To fabricate vertical wall components by deposition of ER5356 aluminium alloy to build thin wall vertical structures using the developed WAAM setup. (*Addresses Gap-2.*)
3. To investigate the effect of inter-layer (interpass) active cooling on the material properties using water as a coolant, and to develop a new water based cooling approach that extends the existing *Near Immersion Active Cooling* (NIAC) configuration with a localised cooling action, with a quantitative comparison between the two. (*Addresses Gap-3.*)
4. To investigate the effect of deposition strategy on the material properties along with active cooling. (*Addresses Gap-4.*)
5. To investigate the effect of inter-layer (interpass) active cooling on the material properties using liquid nitrogen as a coolant. (*Addresses Gap-5.*)

The experimental design, characterisation methodology and the interpretation of results in the remainder of this thesis are organised so that each chapter contributes evidence toward one or more of these objectives.

Chapter 3

Materials and Methodology

3.1 Research Framework

The present chapter describes the complete experimental methodology adopted to systematically investigate the influence of seven distinct interpass thermal management strategies on the WAAM process behaviour, deposited geometry, microstructural evolution and mechanical performance of ER5356 aluminium alloy thin wall structures. The methodology was structured around a two phase comparative approach. In Phase 1 (NC, PC, A20, A10), a fixed set of process parameters was used while the interpass cooling technique was varied as the sole independent variable. In Phase 2 (TIAC, TIAC-BD, LN₂), a second parameter set was employed, selected to accommodate the physical constraints of the TIAC hardware configuration and to evaluate cooling effectiveness under a more thermally demanding deposition scenario; the rationale is given in Section 3.5. For six of the seven conditions the deposition path strategy was also held constant (unidirectional); for the seventh condition (TIAC-BD) the deposition direction was the sole additional variable, enabling its effect to be isolated through direct comparison with TIAC under otherwise identical thermal and process conditions.

The seven cooling strategies evaluated in this work are:

1. **Natural Cooling (NC)** — baseline condition with no external cooling medium or device.
2. **Passive Cooling (PC)** — tap water maintained just below the substrate underside, providing passive conductive heat dissipation.
3. **Active Cooling, LEWD = 20 mm (A20)** — substrate partially submerged in tap water, with the Layer Edge to Water Distance (LEWD) maintained at 20 mm by periodic fresh water addition.
4. **Active Cooling, LEWD = 10 mm (A10)** — variant of A20 with the water

level raised to a LEWD of 10 mm to increase heat extraction intensity.

5. **Tandem Immersion Active Cooling (TIAC)** — novel architecture developed in this research, combining substrate immersion (LEWD = 10 mm) with a localised water impingement jet directed at the previously deposited layer.
6. **TIAC with Bidirectional Deposition (TIAC-BD)** — identical to TIAC in all cooling and process parameters; deposition direction is the sole variable changed (unidirectional replaced by bidirectional).
7. **Cryogenic Cooling (LN₂)** — liquid nitrogen at $-196\text{ }^{\circ}\text{C}$ applied at each inter-pass interval via a pressurised delivery nozzle.

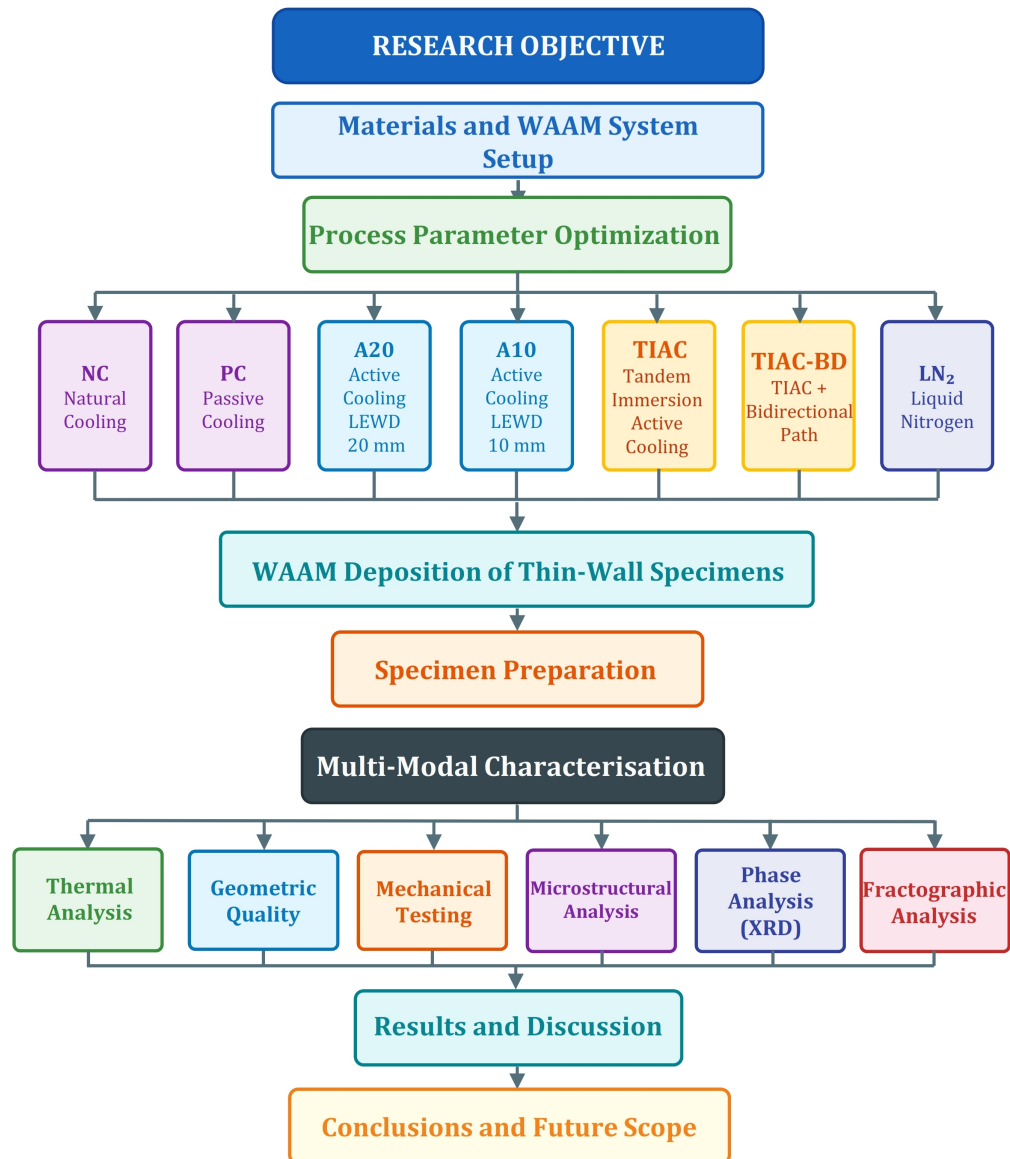


Figure 3.1: Schematic overview of the experimental methodology adopted

For each condition, thin wall specimens of identical geometry were fabricated under a

controlled robotic WAAM setup. Thermal histories were recorded in real time during deposition. After deposition, specimens were machined into standardised coupons and subjected to geometric, thermal, mechanical, microstructural, phase and fractographic characterisation.

The overall research workflow is summarised schematically in Figure 3.1.

3.2 Materials and Consumables

3.2.1 Feedstock Wire (ER5356 Aluminium Alloy)

ER5356 is an Al–Mg alloy (approximately 5 wt.% Mg) conforming to AWS A5.10/A5.10M classification. As discussed in Chapter 1, it was selected for its combination of arc stability, solid solution strengthening, corrosion resistance and suitability for WAAM of structural thin wall components [12, 13]. The wire was supplied as a 7 kg spool with a diameter of 1.2 mm. The nominal composition limits as specified in AWS A5.10/A5.10M and the certified composition values from the manufacturer’s certificate of conformance supplied with the wire are given in Table 3.1.

Table 3.1: Chemical composition of ER5356 welding wire (wt.%)

Element	Al	Mg	Cr	Mn	Si	Fe
Nominal wt.%	Bal.	4.7–5.5	0.05–0.20	0.05–1.00	≤0.25	≤0.40
Certificate wt.%	Bal.	4.83	0.18	0.157	0.110	0.285

3.2.2 Substrate Material

Thin wall specimens were deposited onto substrate plates of AA6061-T6 with nominal dimensions of $300 \times 300 \times 25$ mm. The substrate surface was cleaned immediately prior to deposition by stainless steel wire brushing to disrupt and remove the native aluminium oxide layer, followed by acetone degreasing to remove surface oils and organic contaminants, both of which could otherwise cause porosity or lack of fusion defects in the first deposited bead. The substrates were clamped rigidly to the work tank base plate to minimise thermal distortion during deposition.

3.2.3 Shielding Gas

Pure argon (Ar, 99.99% purity, industrial grade) was used as the shielding gas throughout all deposition experiments. Argon is the preferred shielding gas for aluminium WAAM because its higher density provides effective atmospheric exclusion, its inert chemistry avoids oxidation and nitridation of the molten pool, and its lower thermal

conductivity relative to helium promotes arc stability and a smooth metal transfer mode [12]. A constant flow rate of 20 L/min was maintained using a calibrated flow meter attached to the gas line.

3.2.4 Scope of the Experimental Programme

The experimental programme is bounded in three respects.

The material is confined to a single feedstock, ER5356, for the reasons set out in Section 2.3.2. The governing consideration is that the alloy is non heat treatable, so that every difference in strength, ductility, hardness and toughness measured across the seven cooling conditions is traceable to the deposition thermal history rather than to a precipitation response that the cooling strategy would itself have altered.

The geometry is confined to single bead thin walls deposited vertically on a flat substrate. This is the benchmark geometry used throughout the WAAM cooling literature, and it isolates the thermal variable from the path planning and inter bead overlap effects that a multi bead build would introduce. Multi bead walls, curved and closed section geometries and overhanging features are outside the present scope.

The process is confined to a single GMAW configuration with one fixed parameter set within each of the two experimental phases, so that the interpass cooling strategy remains the only deliberately varied factor within a phase. Low heat input arc modes such as CMT and pulsed GMAW, and post deposition treatments of any kind, are not examined.

3.3 Development of WAAM System

The WAAM system was assembled from commercially available components to reduce capital cost while achieving the industrial grade process stability required for reproducible experiments. The integrated system consists of a robotic manipulator, a welding power source with wire feeder and a shielding gas delivery circuit. The hardware configuration is shown in Figure 3.2.

3.3.1 Welding Power Source

The welding power source used in this research is the Lorch S5-RoboMIG XT, a dedicated robotic MIG/MAG welding package designed for integration with KUKA robot controllers. The system supports both conventional MIG and advanced pulse/SpeedPulse^{XT} transfer modes, with a current output range of 25–400 A and a corresponding voltage range of 14.5–34 V. The duty cycle is rated at 50% at the maximum output of 400 A and rises to 100% at 320 A, ensuring sustained operation

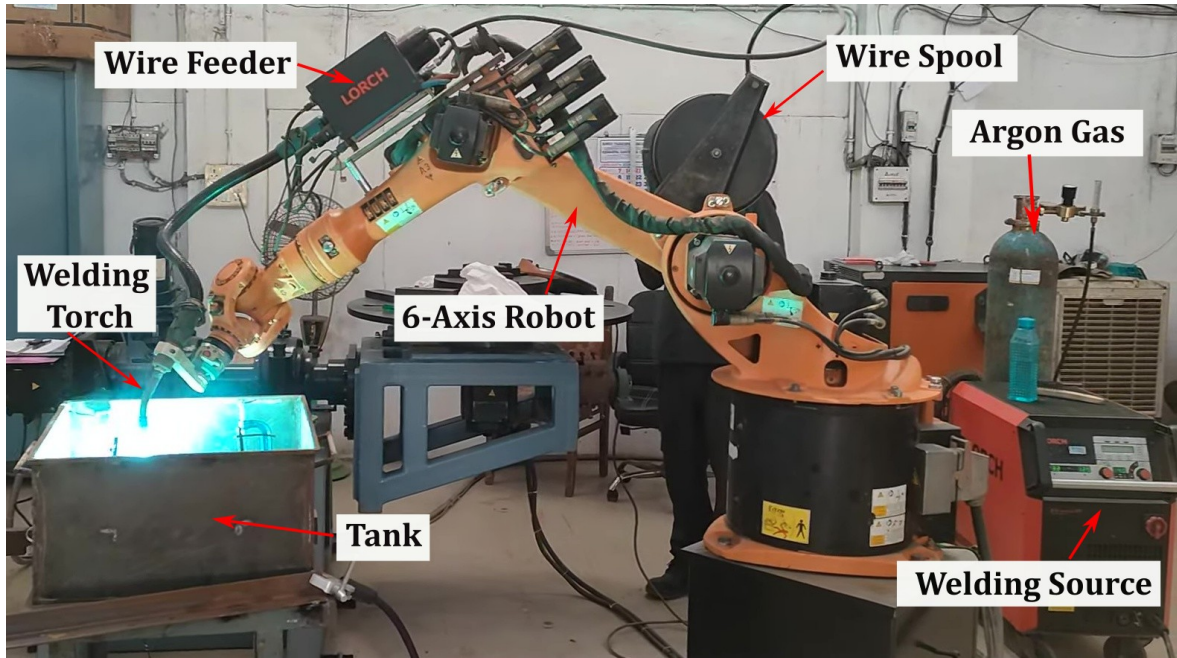


Figure 3.2: Integrated WAAM system used for experimentation

during multi-layer WAAM builds without thermal overload. The wire feed speed is adjustable from 1.5 to 25 m/min and the unit is compatible with aluminium wire diameters of 1.0–1.2 mm, accommodating the ER5356 wire (1.2 mm) used in this study. The power source operates on a 3-phase, 400 V, 50 Hz supply with an open-circuit voltage of 50 V. Communication between the robot controller and the welding power source is established via a digital interface, allowing synchronised motion and welding parameter control from a single programme. The principal specifications of the power source are summarised in Table 3.2.

Table 3.2: Specifications of the Lorch S5-RoboMIG XT welding power source

Parameter	Value / Range
Welding process	MIG / Pulse MIG (SpeedPulse ^{XT})
Current range	25–400 A
Voltage range	14.5–34 V
Duty cycle at rated current	50% at 400 A, 100% at 320 A
Wire diameter compatibility	1–1.2 mm for Aluminium
Wire feed speed	1.5–25 m/min
Input supply	3-phase, 400 V, 50 Hz
Open-circuit voltage	50 V

3.3.2 Robotic System

Robot guided torch motion was provided by a KUKA KR 8 R1420 arc HW six-axis industrial robot, paired with a KR C4 controller. The robot was programmed using KUKA Robot Language (KRL) to execute the layer by layer deposition toolpaths. The torch was mounted on the robot flange via a rigid aluminium adapter bracket to ensure zero play between the tool centre point (TCP) and the programmed path. A coordinate system was defined at the substrate clamping fixture to provide a fixed reference frame for all depositions. The robot's positional repeatability of 0.04 mm ensured consistent inter layer registration throughout multi layer builds. The key technical specifications of the KUKA KR 8 R1420 arc HW robot are summarised in Table 3.3.

Table 3.3: Technical specifications of the KUKA KR 8 R1420 arc HW industrial robot

Parameter	Specification	
Rated payload	8 kg	
Maximum reach	1421 mm	
Number of axes	6	
Pose repeatability (ISO 9283)	±0.04 mm	
Controller	KR C4 smallsize-2 / KR C4 compact	
Operating temperature	5 °C to 45 °C	
<i>Axis Motion Range and Maximum Speed (rated payload)</i>		
Axis	Motion Range	Max Speed
A1	±170°	220 °/s
A2	−185° / +65°	210 °/s
A3	−120° / +180°	270 °/s
A4	±165°	430 °/s
A5	−115° / +140°	430 °/s
A6	±350°	630 °/s

3.3.3 Wire Feed Mechanism

The wire feed system is integrated within the Lorch S5-RoboMIG XT platform, as shown in Figure 3.3. It employs a four roller drive mechanism in which two pairs of opposing grooved rollers grip the wire from above and below, providing positive traction that prevents slippage and wire buckling during continuous deposition. This roller configuration ensures smooth and consistent delivery of the 1.2 mm ER5356 aluminium wire at the programmed feed rate without deforming the relatively soft aluminium wire. The wire spool is mounted directly on the robotic arm, keeping the free wire length short and minimising the risk of feed fluctuations caused by spool

inertia or conduit friction. A conduit liner guides the wire from the spool through the cable assembly to the torch contact tip, maintaining a controlled and reproducible wire feed throughout the multi layer build.

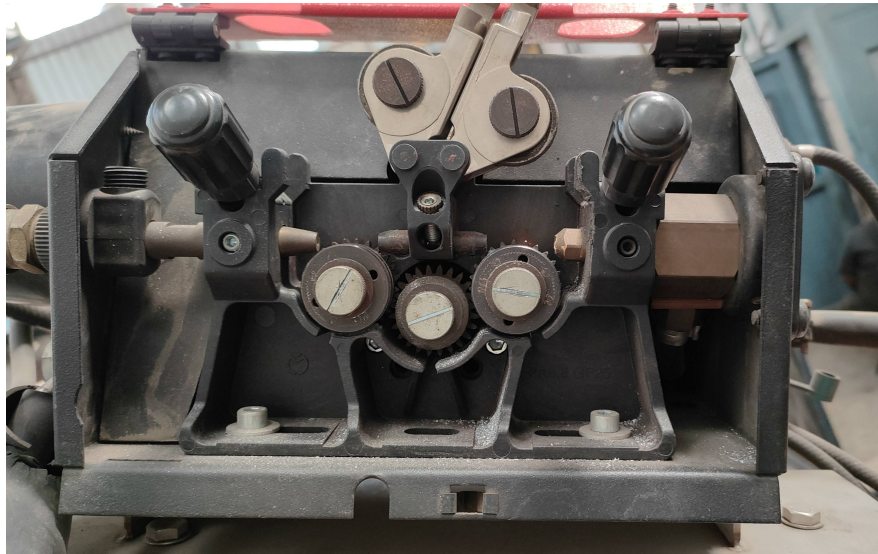


Figure 3.3: Four roller wire feed unit of the Lorch S5-RoboMIG XT platform

3.3.4 Shielding Gas Delivery

Shielding gas was delivered through the concentric annular nozzle of the GMAW torch. A precision pressure regulator and calibrated rotameter were used to maintain a steady argon flow rate of 20 L/min throughout each deposition run. The gas line was purged for 30 s before arc ignition to flush atmospheric moisture from the hose and torch body. Post flow shielding was maintained for 10 s after arc extinction to protect the solidifying weld pool from oxidation.

3.4 Experimental Setup and Instrumentation

3.4.1 Substrate and Work Tank

A custom fabricated mild steel work tank was designed and manufactured to serve as the primary containment structure for the cooling medium in the active cooling experiments. The overall construction and dimensions of the tank are shown in Figure 3.4. The tank was dimensioned to accommodate the substrate plate with clearance on all sides and a depth sufficient to achieve the required immersion level for the active cooling conditions. When filled with water to the level of the substrate underside, the effective water volume in the tank is approximately 47 L. The substrate plate was mounted on a raised mild steel support grid inside the tank, allowing the cooling medium to act

on the substrate underside and lateral faces. Four C-clamps were used to secure the substrate to the support grid and to maintain a fixed datum for the robot's TCP. The internal substrate mounting arrangement is illustrated in Figure 3.5.

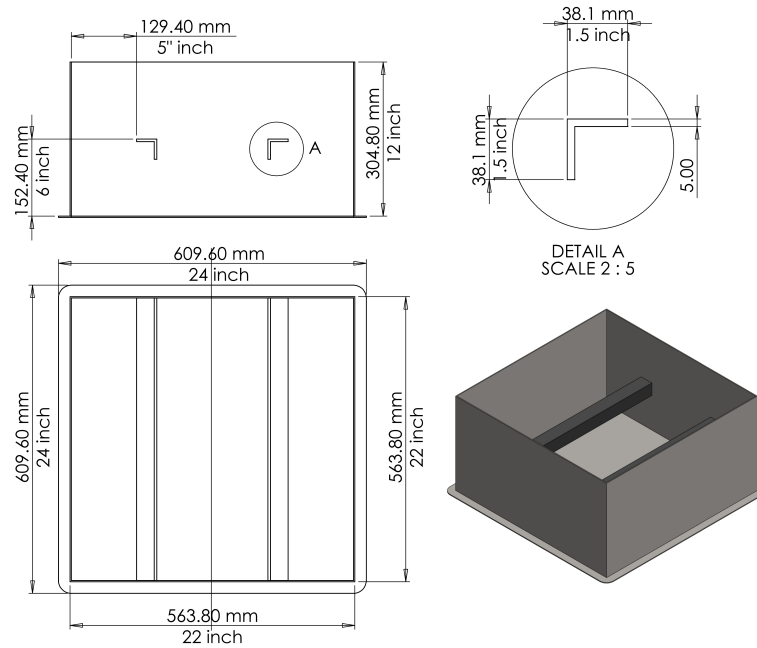


Figure 3.4: Construction and overall dimensions of the work tank

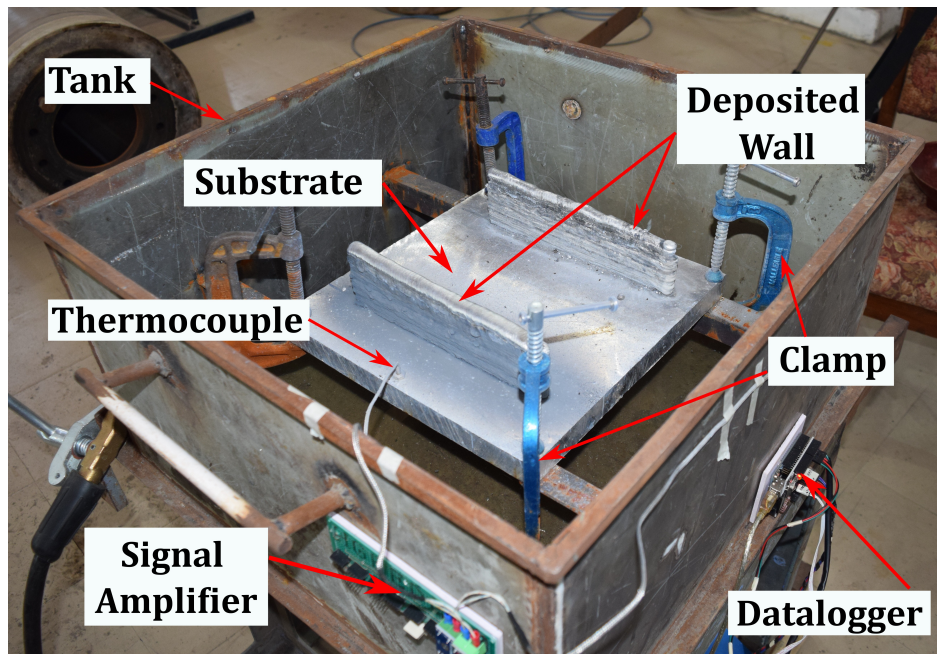


Figure 3.5: Substrate mounting arrangement inside the mild steel work tank

3.4.2 Thermocouple Arrangement

Thermal histories during WAAM deposition were recorded using K-type thermocouples (chromel–alumel, wire diameter 0.5 mm, accuracy ± 1.5 °C or $\pm 0.4\%$ of reading, whichever is greater). Although the custom datalogger is capable of recording four thermocouple channels simultaneously, two thermocouples were employed in this study. TC-A was used to record the substrate temperature during deposition. It was inserted into a blind hole drilled in the substrate plate, located 50 mm laterally from the centreline of the deposition path and at the mid length of the deposition path, ensuring the junction was in intimate contact with the substrate material. It is noted that this measurement represents the thermal response of the substrate to heat conducted from the growing deposit, not the temperature of the deposited layer itself. TC-B was immersed in the cooling water to monitor the bulk water temperature during deposition, applicable to the PC, A20, A10, TIAC and TIAC-BD conditions; for the NC and LN₂ conditions, TC-B was not deployed. The K-type thermocouple and datalogger assembly were formally calibrated by the comparison method at four reference points of 20, 100, 400 and 700 °C using a traceable S-type thermocouple reference standard (ISO/IEC 17025 accredited laboratory, NABL India). The maximum observed deviation of the unit under calibration from the reference master was -1.08 °C at 700 °C, and the expanded measurement uncertainty across the calibration range was ± 0.99 to ± 2.79 °C (coverage factor $k = 2$, confidence level 95%). The calibration was carried out without adjustment; all reported thermal data therefore carry this stated uncertainty. Each thermocouple was connected to a dedicated MAX6675 thermocouple to digital signal amplifier module, shown in Figure 3.6(c), which provides cold junction compensation and 12-bit digital output.

3.4.3 Custom Arduino Based Datalogger

A custom multi channel temperature datalogger was designed and fabricated to record the signals from the four K-type thermocouples simultaneously. The hardware components of the system are shown in Figure 3.6. The datalogger is built around an Arduino Uno microcontroller board mounted with a dedicated data logging shield (Figure 3.6a), which integrates an SD card module for onboard data storage and a DS3231 real-time clock (RTC) module that provides a hardware stamped date & time reference for each recorded sample, enabling precise synchronisation of the thermal history with the robot toolpath log. Each of the four K-type thermocouples is interfaced through a MAX6675 signal amplifier module (Figure 3.6c), providing 12-bit cold junction compensated temperature readings over the range 0–1024 °C with a resolution of 0.25 °C. A JHD 204A alphanumeric LCD display (Figure 3.6b) provides real-time onsite readout of all four channel temperatures during deposition. Temperature data are logged to the SD card

as comma separated values (CSV) files with columns for elapsed time, RTC timestamp and individual thermocouple readings (TC-A through TC-D). The complete bill of materials and firmware source code for the datalogger are provided in Appendix A.

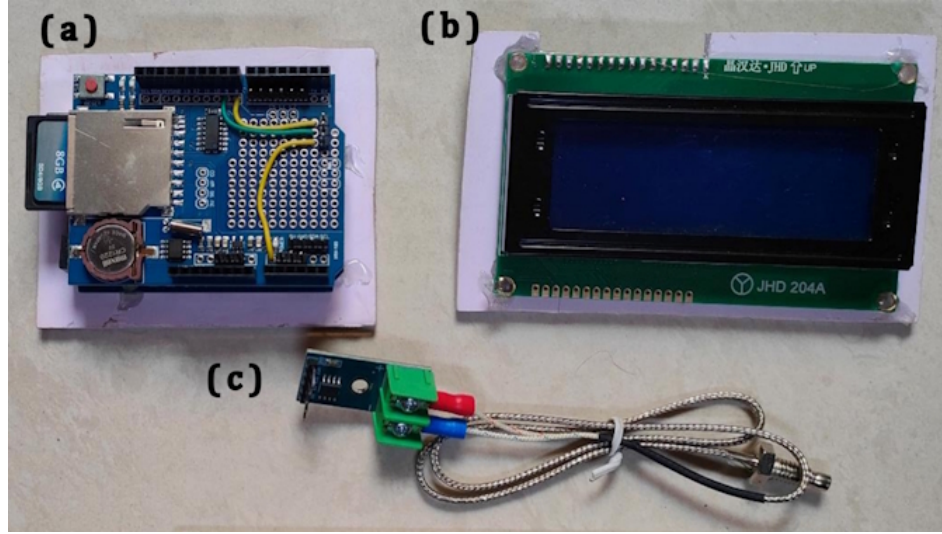


Figure 3.6: Hardware components of the temperature datalogger:
 (a) Arduino Uno with data logging shield; (b) LCD display; (c) MAX6675 thermocouple signal amplifier with K-type thermocouple

3.5 Process Parameters

The experimental programme was structured into two phases, each employing a dedicated set of process parameters established through systematic preliminary single bead and multi layer trials. **Phase 1** (NC, PC, A20, A10) used a parameter set optimised for stable deposition under standard conditions and forms the basis of the published results. **Phase 2** (TIAC, TIAC-BD, LN₂) employed a different parameter set whose rationale is discussed below. Parameters common to both phases like wire material, substrate, wire diameter, CTWD, shielding gas, flow rate, programmed layer height, target wall dimensions and interpass dwell were held constant. The process parameters for both phases are listed in Table 3.4.

Rationale for Phase 2 Parameter Selection

The Phase 2 travel speed was reduced from 12.5 mm/s to 4.0 mm/s for three interrelated reasons. First, the TIAC impingement jet nozzle is positioned 50 mm behind the welding torch. At 12.5 mm/s, the jet contacts any given point on the freshly deposited surface for only 4 s ($t = 50 \text{ mm}/12.5 \text{ mm/s}$); at 4.0 mm/s this increases to 12.5 s, providing sufficient dwell time for effective inter-layer heat extraction. This kinematic constraint is analogous to trailing gas cooling systems, where nozzle offset relative to the torch and travel speed together govern the cooling contact time on the deposit

Table 3.4: WAAM process parameters for Phase 1 (NC, PC, A20, A10) and Phase 2 (TIAC, TIAC-BD, LN₂) experimental conditions

Parameter	Phase 1	Phase 2	Unit
Wire material	ER5356	ER5356	—
Substrate	AA6061-T6 (300×300×25 mm)		—
Wire diameter	1.2	1.2	mm
Welding current	123	122	A
Arc voltage	22.1	20.8	V
Wire feed speed	7.6	7.3	m/min
Travel speed	12.5	4.0	mm/s
CTWD	10	10	mm
LEWD ¹	10–20	10	mm
Shielding gas	Argon (99.99%)		—
Shielding gas flow rate		20	L/min
Heat input Q (Eq. 3.1)	0.174	0.508	kJ/mm
Layer height (programmed)	2	3	mm
Target wall length		200–225	mm
Target wall height		60–65	mm
No. of deposited layers ²	32–35	25–29	—

¹Phase 1: 20 mm (A20), 10 mm (A10); Phase 2: 10 mm (TIAC, TIAC-BD); N/A for NC, PC, LN₂

²NC:35, PC:35, A20:32, A10:33 (Phase 1); TIAC:25, TIAC-BD:25, LN₂:29 (Phase 2)

surface [18, 39]. Second, the proximity of the water surface (LEWD = 10 mm) requires a more stable, shorter arc to prevent water vapour from entering the shielding zone; the arc voltage was correspondingly reduced to 20.8 V [53, 83]. Third, systematic preliminary trials confirmed that 4.0 mm/s was the minimum travel speed for stable bead formation with the TIAC jet active, consistent with the established practice of reoptimising deposition parameters whenever a new thermal management hardware configuration is introduced [109, 110].

The resulting increase in heat input per unit length (0.508 kJ/mm versus 0.174 kJ/mm, a factor of 2.9×) means Phase 2 conditions represent a significantly more thermally demanding deposition scenario than Phase 1. The slower travel speed deposits thicker layers per pass (2.60–2.83 mm versus 1.72–1.90 mm), reducing the total number of layers required to achieve the target wall height from 32–35 (Phase 1) to 25–29 (Phase 2) and thereby reducing the total number of interpass interruptions. This is consistent with reported strategies for improving WAAM build productivity through active cooling [84, 110]. Within-phase comparisons (NC vs. PC vs. A20 vs. A10; TIAC vs. TIAC-BD vs. LN₂) are fully parameter controlled. Cross phase comparisons are interpreted in the context of the respective heat inputs and are discussed in Chapter 4.

A uniform interpass dwell of approximately 10 s was applied between successive layers across all seven cooling conditions. This brief pause was necessitated by the requirement to inspect and, where applicable, adjust the water level to maintain the target LEWD in the active cooling conditions. To eliminate interpass timing as a confounding variable, the same 10 s dwell was retained identically in the NC and PC conditions, ensuring that any differences in thermal history or mechanical response are attributable solely to the cooling medium.

The heat input per unit length was calculated using the standard formula:

$$Q = \frac{\eta \cdot V \cdot I}{v_t} \quad (3.1)$$

where Q is the net heat input (J/mm), η is the arc thermal efficiency for GMAW (taken as 0.8 [13]), V is the arc voltage (V), I is the welding current (A), and v_t is the torch travel speed (mm/s).

3.6 Cooling Techniques Implemented

Seven interpass thermal management strategies were implemented to provide a systematic progression from purely passive heat dissipation to aggressive active cooling, with additional investigation of the effect of deposition path on thermal behaviour. A unified schematic illustrating the physical cooling arrangement for all seven conditions is shown in Figure 3.7. Each strategy is described in the subsections below, and the key distinguishing features are summarised in Table 3.5.

3.6.1 Natural Interpass Cooling — NC (Baseline)

In the natural cooling condition, no external cooling medium or heat extraction device was applied to the deposit or the substrate. Deposition was performed layer by layer under ambient laboratory conditions (temperature ≈ 25 °C, still air) with no interpass temperature threshold; a uniform 10 s interpass dwell was applied consistent with all other conditions (see Section 3.5). This condition represents the most common industrial WAAM practice and serves as the reference against which all active cooling strategies are benchmarked. Heat accumulation under natural cooling is inherently progressive because the thermal resistance of the growing deposit retards heat conduction to the substrate, resulting in a steady rise in the peak interpass temperature over successive layers [10, 13].

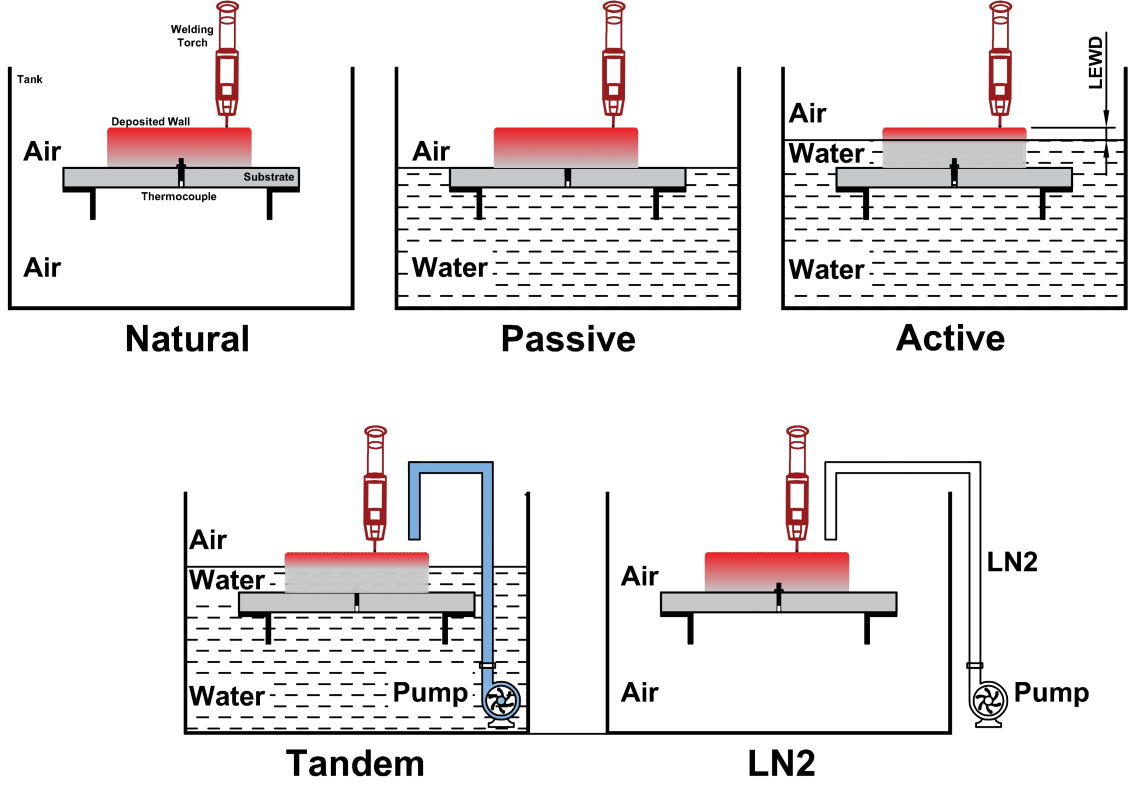


Figure 3.7: Schematic representation of all seven interpass cooling strategies implemented

3.6.2 Passive Cooling — PC

In the passive cooling (PC) condition, no external fins, heat sinks, or forced convection devices were attached to the substrate or deposit. The mild steel work tank was filled with tap water to a level just below the underside of the substrate plate, such that the substrate bottom was in close proximity to the water surface without direct immersion of the deposition wall. As in all other conditions, a uniform 10 s interpass dwell was observed before initiating each successive layer. The presence of water beneath the substrate provides a modest enhancement in conductive heat dissipation compared to still air while avoiding the more aggressive immersion cooling of the A20 and A10 conditions, making PC representative of a minimal passive thermal management intervention.

3.6.3 Active Cooling with LEWD = 20 mm — A20

In the A20 condition, the mild steel work tank was filled with tap water to a level at which the Layer Edge to Water Distance (LEWD) is defined as the vertical distance between the top surface of the substrate and the water surface which was maintained at 20 mm. The LEWD values of 20 mm (A20) and 10 mm (A10) were selected based on the near immersion active cooling (NIAC) framework established by Scotti et al. [53,88],

who demonstrated that decreasing LEWD progressively suppresses heat accumulation in aluminium WAAM by reducing the solid conduction resistance between the deposit base and the cooling water surface, with LEWD = 20 mm representing a moderate cooling level and lower values producing more aggressive heat extraction. The water level was monitored throughout deposition and fresh water was added periodically as required to maintain the target LEWD; no recirculating pump circuit was employed. Under this arrangement, the underside and lateral faces of the substrate plate were in direct contact with the cooling water, while the top deposition surface remained above the waterline to prevent direct interaction between the liquid and the molten pool. The significantly higher thermal conductivity of water relative to still air ($k_{\text{water}} \approx 0.6 \text{ W m}^{-1}\text{K}^{-1}$ versus $k_{\text{air}} \approx 0.026 \text{ W m}^{-1}\text{K}^{-1}$ at ambient temperature [111]) produces a substantially higher rate of heat extraction from the substrate compared to NC and PC, suppressing heat accumulation in the deposit.

3.6.4 Active Cooling with LEWD = 10 mm — A10

The A10 condition was identical in principle to A20 but with the water level raised so that the LEWD was reduced to 10 mm. Fresh water was again added periodically to maintain this reduced distance throughout the build. A smaller LEWD shortens the solid conduction path between the hot deposit base and the cooling water interface, increasing the effective heat transfer coefficient and producing a more aggressive heat extraction rate than A20. A LEWD of 10 mm was identified as the practical lower limit for stable deposition with the present process parameters; further reduction risked direct contact between the water surface and the active molten pool, which would cause hydrogen porosity, spatter and arc instability. The A10 condition therefore represents the most aggressive purely immersion-based cooling strategy in the experimental matrix.

3.6.5 Tandem Immersion Active Cooling — TIAC

The Tandem Immersion Active Cooling (TIAC) strategy is a novel thermal management architecture developed in the present research and constitutes the primary experimental contribution of this work. TIAC combines the substrate level immersion heat extraction of the active cooling conditions (A20/A10) with a secondary localised impingement jet of water directed at the sidewall of the most recently deposited layer during each interpass interval. The tank was filled with tap water to a LEWD of 10 mm, the same level used for the A10 condition. The jet water is drawn from the same water already present inside the mild steel work tank by a submersible recirculation pump; no external water source is used for the jet circuit. The pump directs the recirculated tank water through a nozzle positioned 50 mm behind the welding torch nozzle and

oriented at 90° to the deposited wall surface. The 50 mm offset was selected to ensure the jet remained outside the arc shielding envelope, preventing water vapour from entering the shielding zone and causing hydrogen pick-up or porosity in the aluminium deposit, consistent with the minimum safe cooling distances reported for active cooling in WAAM [53]. The water flow rate through the impingement nozzle was maintained at 1 L/min, established through preliminary calibration trials as a conservative rate that provides effective heat extraction without causing thermal shock or excessive splashing near the arc zone. To compensate for evaporative and splashing losses, the tank water level is monitored and fresh water is added periodically to maintain the target LEWD of 10 mm throughout the build. The tandem action of immersion cooling at the substrate and jet impingement cooling at the deposited wall creates a dual heat extraction pathway hypothesised to be more effective than either mechanism in isolation; this hypothesis is evaluated through the thermal and mechanical characterisation presented in Chapter 4. The TIAC configuration is shown schematically in Figure 3.7.

The tandem architecture differs from near immersion active cooling (NIAC) in the number of independent heat extraction paths available to it. NIAC extracts heat through a single path, namely conduction from the immersed portion of the wall and the substrate into the surrounding bath, and the only control the operator has over that path is the layer edge to water distance. That single control is subject to an inherent conflict: reducing the distance raises the heat extraction rate but brings the water surface toward the arc, where steam interference, spatter and hydrogen pickup impose a floor below which the distance cannot be reduced. Because the water surface cannot approach the arc, the region of the deposit with the steepest thermal gradient, the sidewall of the layer just deposited, is precisely the region the bath cannot reach; it equilibrates with the bulk wall before the immersion path can act on it.

The tandem architecture resolves this conflict by separating the two functions onto two independent paths. The immersion path continues to remove bulk heat from the substrate and the wetted wall at a layer edge to water distance of 10 mm, while the jet path removes heat from the freshly deposited sidewall directly, at a nozzle offset of 50 mm behind the torch which places it outside the shielding envelope. The high gradient region is therefore cooled without the water surface itself having to approach the arc, so that the constraint limiting NIAC is circumvented rather than merely optimised against. The two paths are independently controllable, the immersion path through the layer edge to water distance and the jet path through its flow rate and offset, which gives the process a second degree of freedom that immersion alone does not possess. The experimental consequence is that TIAC attains a heat accumulation index of $55.0\text{ }^\circ\text{C}$ while operating at $2.9\times$ the heat input of the immersion only Phase 1 conditions, so that normalised by heat input it extracts more heat per unit energy

delivered than immersion alone.

3.6.6 TIAC with Bidirectional Deposition — TIAC-BD

The TIAC-BD condition employed an identical cooling arrangement to TIAC in which substrate immersion with the tank water level maintained at the target LEWD by periodic tap water addition, combined with inter layer wall impingement jet cooling delivered via the same submersible recirculation pump drawing from the tank water, but replaced the unidirectional deposition path with a bidirectional (alternating) deposition strategy. All process and cooling parameters were identical to TIAC; the deposition direction was the sole variable changed, enabling a direct assessment of the effect of torch path strategy on wall geometry and property uniformity under otherwise identical thermal conditions. In the bidirectional strategy, odd numbered layers are deposited left-to-right and even numbered layers right-to-left, distributing the arc start/stop heat input symmetrically and eliminating the end-taper effect associated with unidirectional builds. The motivation for TIAC-BD was to investigate whether the combined benefit of maximum active heat extraction (TIAC) and geometric uniformity (bidirectional path) could further reduce mechanical anisotropy and improve spatial homogeneity of the deposited structure relative to unidirectional TIAC. TIAC-BD therefore represents the most comprehensive cooling and deposition control strategy in the experimental matrix.

3.6.7 Cryogenic Cooling — LN₂

In the cryogenic cooling condition, liquid nitrogen (LN₂) at a temperature of $-196\text{ }^{\circ}\text{C}$ was applied to the inter layer surface at each interpass interval by directing a controlled flow through a stainless steel cryogenic delivery nozzle. The LN₂ was supplied from a pressurised 35 L Dewar vessel. The nozzle was positioned to distribute the cryogen across the full width of the previously deposited bead. Based on comparable LN₂ interpass cooling studies in WAAM [112, 113], a controlled LN₂ flow rate of 1.0 L/min was applied continuously at the interpass stage; this flow rate provides effective inter layer heat extraction without inducing excessive thermal contraction or cold cracking in the aluminium deposit. Cryogenic cooling represents the most aggressive heat extraction condition in the experimental matrix and was included to investigate the effects of extreme inter layer thermal gradient on microstructural refinement and mechanical performance.

Table 3.5: Comparative summary of the seven interpass cooling strategies evaluated in this work

Label	Cooling Condition	Cooling Medium	Cooling Applied To	Cooling Mechanism	LEWD	Deposition Direction
NC	Natural	Air	Substrate + wall	Free convection	N/A	Unidirectional
PC	Passive	Water	Substrate	Conduction	N/A	Unidirectional
A20	Active	Water	Substrate + partial wall	Conduction	20 mm	Unidirectional
A10	Active	Water	Substrate + partial wall	Conduction	10 mm	Unidirectional
TIAC	Active	Water	Substrate + wall	Forced conduction	10 mm	Unidirectional
TIAC-BD	Active	Water	Substrate + wall	Forced conduction	10 mm	Bidirectional
LN ₂	Active	LN ₂	Wall	Forced conduction	N/A	Unidirectional

3.7 Deposition Strategy

Two torch motion strategies were evaluated for their effect on heat distribution and geometric consistency of the deposited walls.

3.7.1 Unidirectional Deposition

In the unidirectional strategy, the torch traverses from a fixed start point to a fixed end point for every layer. At the end of each pass, the torch is retracted, repositioned at the original start point and incremented upward by one layer height before initiating the next pass. This strategy is simple to programme and produces a predictable thermal gradient along the deposition direction; however, it introduces a systematic asymmetry in the heat distribution because the start end of the wall accumulates more heat from frequent arc restarts. The resulting differential thermal history across the wall length may produce a slight taper in the wall width that must be accounted for in the geometric analysis.

3.7.2 Bidirectional Deposition

In the bidirectional (alternating) strategy, the torch direction is reversed at the end of each layer: odd numbered layers are deposited left-to-right and even numbered layers right-to-left. This oscillating pattern distributes the start/stop heat input more symmetrically, reducing the end taper effect and promoting a more uniform interpass temperature distribution along the wall length. The bidirectional strategy is gener-

ally preferred in WAAM research because it yields more consistent geometric and microstructural properties [10, 13]. In the present work, unidirectional deposition was used for NC, PC, A20, A10, TIAC and LN₂; bidirectional deposition was applied exclusively to TIAC-BD to investigate the effect of torch path strategy on the deposited structure under identical TIAC cooling conditions.

3.8 Specimen Preparation

After deposition, each as-built wall was detached from the substrate plate by hacksaw cutting along the substrate-wall interface. Both ends of the wall were subsequently removed by hacksaw to expose cross sectional surfaces for metallographic examination and to provide planar reference surfaces for the subsequent milling operation. Both lateral faces of the wall were then machined on a vertical milling machine to produce flat, parallel surfaces and to reduce the wall thickness to a uniform 5 mm. A wall thickness of 5 mm was selected because it is the common critical dimension for both the tensile sub-size specimens (thickness dimension, ASTM B557M) and the Charpy V-notch impact specimens (width dimension, ASTM E23), enabling all mechanical test coupons to be extracted from a single standardised wall section. The milled walls were subjected to dimensional inspection before specimen sectioning.

Specimens were then machined from the milled walls using wire electrical discharge machining (WEDM) to preserve the as-built microstructure and avoid heat induced microstructural alteration that would arise from conventional sawing. Ten coupons per wall were extracted to cover tensile testing in four orientations and Charpy V-notch impact testing in two orientations. The extraction layout is shown in Figure 3.8, and each specimen is identified by an alphanumeric code defined in Table 3.6.

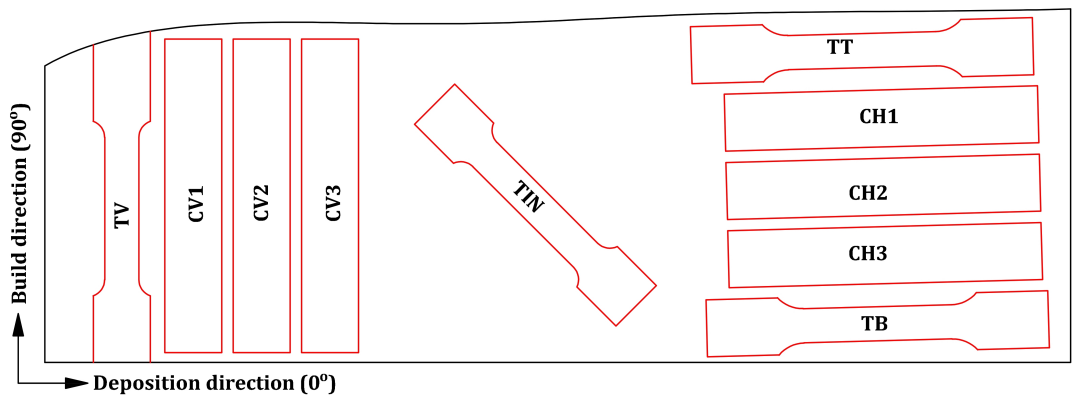


Figure 3.8: Specimen extraction plan from an as-built WAAM thin wall showing the location and orientation of all test coupons. Build direction (90°) is vertical; deposition direction (0°) is horizontal

Tensile specimens were machined to the sub-size geometry specified in ASTM B557M. Four orientations were extracted to quantify mechanical anisotropy: transverse to the deposition direction (TV, 90°, along the build direction), longitudinal from the top zone (TT, 0°, along the deposition direction), longitudinal from the bottom zone (TB, 0°), and inclined (TIN, 45°). Charpy V-notch impact specimens were machined in accordance with ASTM E23 with sub-size dimensions of 55 × 10 × 5 mm, consistent with the 5 mm wall thickness after milling. Three specimens were extracted in the transverse orientation (CV1–CV3, 90°, along the build direction) and three in the longitudinal orientation (CH1–CH3, 0°, along the deposition direction), each set sampling different height positions within the wall to assess the variation in impact toughness with build height. All impact energy results were normalised to unit cross sectional area (J/cm²) for cross condition comparison. Metallographic examination was carried out on cross sections obtained from the material adjacent to the Charpy blanks. These sections were mounted in conductive Bakelite, and polished through successive grades of SiC abrasive paper (240, 400, 600, 800, 1200, 2500 grit) followed by 1 µm and 0.05 µm alumina suspension on a vibratory polisher to achieve a mirror finish. Specimens were etched with Keller’s reagent (2.5 mL HNO₃, 1.5 mL HCl, 1.0 mL HF, 95 mL H₂O) for 15 to 20 s to reveal grain boundaries.

Table 3.6: Specimen naming convention used for all WAAM walls

Code	Specimen Type	Orientation corresponding to deposition direction / Zone
TV	Tensile — Vertical	Transverse (90°)
TB	Tensile — Bottom layers	Longitudinal (0°), bottom zone
TT	Tensile — Top layers	Longitudinal (0°), top zone
TIN	Tensile — Inclined	45°
CV	Charpy — Vertical	Transverse (90°)
CH	Charpy — Horizontal	Longitudinal (0°)

3.9 Characterisation Techniques

3.9.1 Thermal Analysis

Temperature histories recorded by the Arduino datalogger during deposition were subsequently analysed to extract thermal metrics characterising each cooling condition. The following parameters were evaluated from each thermocouple record: peak interpass temperature (T_{peak}), cooling rate from peak to 100 °C, and the layer-to-layer evolution of steady state temperature. No interpass temperature threshold was enforced in any condition. A uniform 10 s interpass dwell was applied across all conditions (Sec-

tion 3.5), and the recorded thermal data were used solely for post process comparative analysis. Thermal profiles were plotted as a function of layer number to visualise heat accumulation trends across the wall height.

The thermal characterisation in this work is measurement based. The seven conditions differ not only in the magnitude of heat extraction but in its location: immersion acts on the substrate and the wetted wall, the impingement jet acts on the sidewall of the freshly deposited layer and cryogen acts on the upper surface of the deposit alone. Representing these three heat extraction paths in a single transient model would require boundary conditions that are themselves the object of the enquiry, in particular the effective heat transfer coefficient at a boiling water to hot aluminium interface and the film boiling behaviour of nitrogen against a surface above the Leidenfrost temperature. The instrumented thermocouple record instead captures the thermal history as it occurred under each condition, and provides the boundary condition data such a model would require. Development of a validated transient model is recommended as future work in Chapter 5.

3.9.2 Geometric Measurement

The geometric quality of the deposited walls was assessed by measuring the total wall width (TWW), effective wall width (EWW), surface waviness (SW), and deposition efficiency (DE) on cross sectional images. The geometric parameters are illustrated schematically in Figure 3.9. All dimensional measurements were performed on cross sectional images using ImageJ image analysis software.

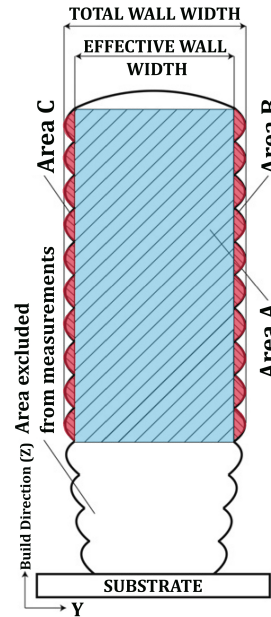


Figure 3.9: Schematic cross sectional view of a WAAM thin wall illustrating the geometric parameters [46]

Surface waviness quantifies the average lateral undulation of the wall surface and is defined as:

$$SW = \frac{TWW - EWW}{2} \quad (3.2)$$

where TWW is the total wall width measured at the outer peaks of the surface undulations and EWW is the effective (flat) wall width enclosed within the undulations.

Deposition efficiency is defined as the fraction of the total deposited cross sectional area that constitutes the effective usable wall section:

$$DE = \frac{\text{UsefulArea}}{\text{TotalArea}} = \frac{A}{A + B + C} \times 100\% \quad (3.3)$$

where A is the effective cross sectional area within the EWW boundary, and B and C are the surface undulation areas on the right and left sides of the wall respectively, as defined in Figure 3.9.

3.9.3 Mechanical Testing

Tensile Testing

Uniaxial tensile tests were conducted at ambient temperature at a crosshead displacement rate of 1 mm/min with a hold time of 1 s, in accordance with ASTM E8/E8M, consistent with the requirements of ASTM B557M for aluminium alloy wrought products. The initial gauge length was 25 mm for all specimen orientations. The flat sub-size specimen geometry used in this study is shown in Figure 3.10. One specimen per orientation per cooling condition was tested across four orientations: transverse to the deposition direction (TV, 90°), longitudinal from the top zone (TT, 0°), longitudinal from the bottom zone (TB, 0°), and inclined (TIN, 45°), giving four tensile specimens per wall. The restriction to one specimen per orientation is a consequence of the limited usable wall volume available after all characterisation coupons were extracted from the 5 mm thick milled wall. The following mechanical properties were determined from the engineering stress–strain curves: ultimate tensile strength (UTS), 0.2% offset yield tensile strength (YTS), percentage elongation at fracture (%El), percentage elongation at yield (%El_y), and elastic modulus (E). Two anisotropy metrics were derived: the anisotropy ratio ($AR_t = UTS_{TB}/UTS_{TV}$), which quantifies the horizontal and vertical strength balance using the bottom zone horizontal specimen most directly influenced by the interpass cooling strategy; and the anisotropy index ($AI_t = UTS_{\max}/UTS_{\min}$), defined as the ratio of the maximum to minimum UTS across all four orientations, providing a broader measure of total directional dependence. Formal definitions and equations for AR and AI are given in Chapter 4 (Equations 4.2 and 4.3).

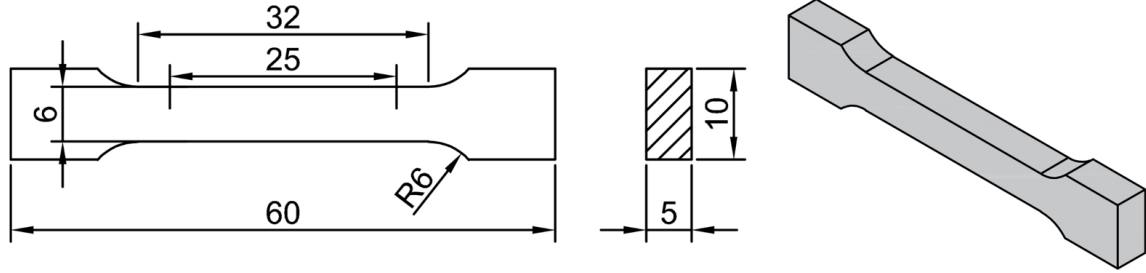


Figure 3.10: Geometry and dimensions of the flat tensile test specimen used in this study, as per ASTM B557M. All dimensions are in mm

Impact Testing

Charpy V-notch impact tests were performed on a pendulum impact testing machine at ambient temperature in accordance with ASTM E23. The specimen geometry and notch dimensions are shown in Figure 3.11. Tests were carried out in two orientations: transverse to the deposition direction (CV1–CV3, 90°, along the build direction) and longitudinal (CH1–CH3, 0°, along the deposition direction), with three specimens per orientation per cooling condition sampling the wall at different height positions. All impact energy results were normalised to unit cross sectional area (J/cm^2) for cross condition comparison.

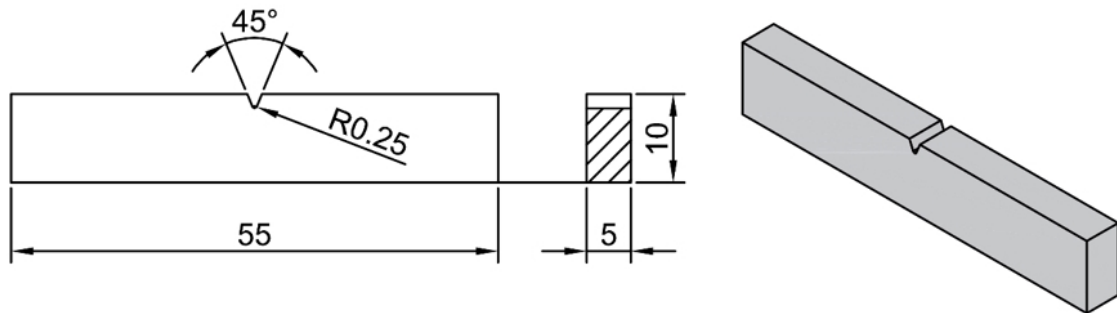


Figure 3.11: Geometry and dimensions of the Charpy V-notch impact test specimen used in this study, as per ASTM E23. All dimensions are in mm

Hardness Testing

Hardness measurements were performed on polished cross sections using a Rockwell cum superficial hardness tester (HR15T scale) in accordance with ASTM E18. Measurements were taken at five locations equally spaced along the build height, from the bottom layer to the top layer of each wall, to assess the variation in hardness with build height across all seven cooling conditions.

3.9.4 Chemical Composition Analysis

The bulk chemical composition of the as deposited walls was determined by optical emission spectrometry (OES) in accordance with ASTM E 1251-17a. Tests were carried out using a Bruker optical emission spectrometer. One flat specimen per cooling condition (seven in total) was prepared from the milled wall and submitted with the parent ER5356 wire as a reference. The reported elements were Al, Mg, Cr, Mn, Si and Fe. Results were compared against the AWS A5.10/A5.10M (ASME SFA 5.10) composition limits for ER5356 to verify that the deposited material conforms to the alloy specification, and against the parent wire composition to quantify any elemental losses, particularly Mg burn off, attributable to the WAAM thermal cycle.

3.9.5 Microstructural Analysis

Optical Microscopy (OM)

Optical micrographs were acquired at magnifications of $100\times$ and $200\times$ across all seven cooling conditions. For the NC, PC and AC conditions (where AC refers collectively to A20 and A10), an Olympus GX51 inverted metallurgical microscope was used; for the TIAC, TIAC-BD and LN_2 conditions, imaging was performed on an Olympus DSX1000 digital microscope. Grain size was measured by the linear intercept method (ASTM E112) on etched specimens. The following microstructural features were assessed: grain morphology (equiaxed vs. columnar), grain size distribution, layer band visibility, porosity and oxide inclusion content and the extent of the heat affected zone at the substrate interface.

Scanning Electron Microscopy and Energy Dispersive Spectroscopy (SEM/EDS)

High magnification imaging of microstructural features was performed on a Hitachi SU3800 scanning electron microscope (Hitachi High Tech India Pvt. Ltd., Vadodara) at the Micro Nano Research and Development Centre, Parul University. The instrument was operated in both secondary electron (SE) and back scattered electron (BSE) modes. Energy dispersive X-ray spectroscopy (EDS) was performed using the integrated Oxford Xplore EDS system; point analyses and elemental mapping were conducted at inter dendritic regions to characterise the distribution of Mg, Al and minor alloying elements and to identify any intermetallic or secondary phase precipitates.

3.9.6 Phase Analysis (XRD)

Phase identification was carried out by X-ray diffraction (XRD) on polished flat specimens using a Bruker D6 PHASER X-ray diffractometer (Bruker India Scientific Pvt. Ltd.) at the Micro Nano Research and Development Centre, Parul University, Vadodara. The instrument is equipped with a 1.2 kW X-ray tube source operating with Cu-K α radiation ($\lambda = 1.5406 \text{ \AA}$) at a measurement precision of $\pm 0.01^\circ$. Scans were performed over a 2θ range of 20° – 90° at a step size of 0.029° and a count time of 0.4 s per step. Phase identification was performed using the International Centre for Diffraction Data Powder Diffraction File release 2 (ICDD PDF-2) database. Lattice parameter shifts of the principal Al matrix reflection were used to assess the degree of Mg solid solution retention under each cooling condition.

3.9.7 Fractographic Analysis

Fracture surface examination was carried out by scanning electron microscopy (SEM) on selected tensile and Charpy V-notch impact specimens to identify the operative failure mechanisms under each interpass cooling condition. Specimens from conditions NC, A10, TIAC, TIAC-BD and LN₂ were selected to represent the full range of thermal histories investigated in this study. All fractographic imaging was performed on the Hitachi SU3800 scanning electron microscope at the Micro Nano Research and Development Centre, Parul University, in secondary electron (SE) imaging mode at an accelerating voltage of 15 kV.

For tensile specimens, fracture surfaces of both vertical (TV) and horizontal bottom zone (TB) orientations were examined to characterise the orientation dependence of the failure mode. For impact specimens, fracture surfaces from horizontal (CH) and vertical (CV) orientations were examined. Faces from horizontal (CH) and vertical (CV) orientations were examined.

Chapter 4

Results and Discussion

This chapter presents results from WAAM of ER5356 aluminium alloy thin wall structures under seven interpass cooling conditions: NC, PC, A20, A10, TIAC, TIAC-BD and LN₂. TIAC, TIAC-BD and LN₂ are novel contributions of this thesis. The programme ran in two phases: Phase 1 (NC, PC, A20, A10; $Q = 0.174$ kJ/mm) and Phase 2 (TIAC, TIAC-BD, LN₂; $Q = 0.508$ kJ/mm). Within phase comparisons are parameter controlled. Cross phase comparisons are discussed in Section 4.8. Results are organised into seven sections such as thermal analysis, geometric characteristics, mechanical properties, chemical composition analysis, microstructural analysis, phase analysis and fractographic analysis followed by a comparative synthesis and a process–structure–property discussion.

4.1 Thermal Analysis

4.1.1 Temperature Evolution During Deposition

The thermocouple data acquired during multi layer deposition provide a continuous record of the thermal environment experienced by the growing deposit under each cooling condition. Figure 4.1 shows the substrate temperature v/s time histories recorded at thermocouple TC-A (positioned in a drilled hole at the mid-length of the substrate, 50 mm from the deposition centreline) for NC, PC, A20 and A10 over the full deposition sequence. In all cases, the thermal signal exhibits a characteristic cyclic pattern: a rapid rise in temperature as the arc passes overhead, followed by an exponential decay during the interpass dwell. At the first deposited layer, all four conditions show similar peak substrate temperatures in the range 5 to 75 °C, as the substrate is still near ambient and the cooling strategies have yet to exert a differential influence. The divergence between conditions becomes increasingly pronounced with each successive layer as the differing heat extraction rates govern the accumulation of heat in the substrate.

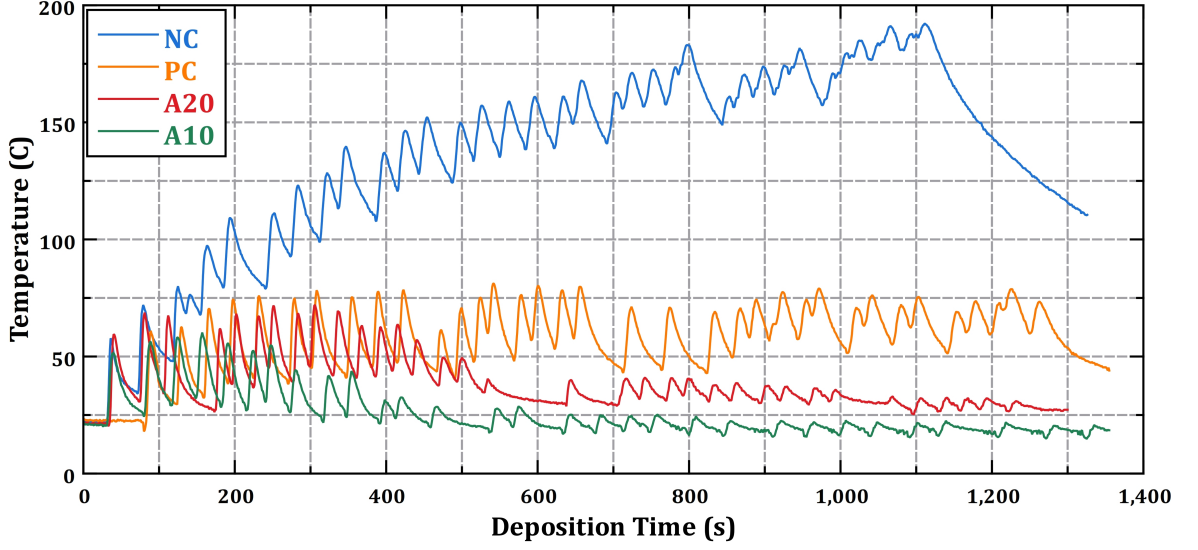


Figure 4.1: Temperature v/s time histories recorded during multi layer WAAM deposition for NC, PC, A20 and A10 cooling conditions

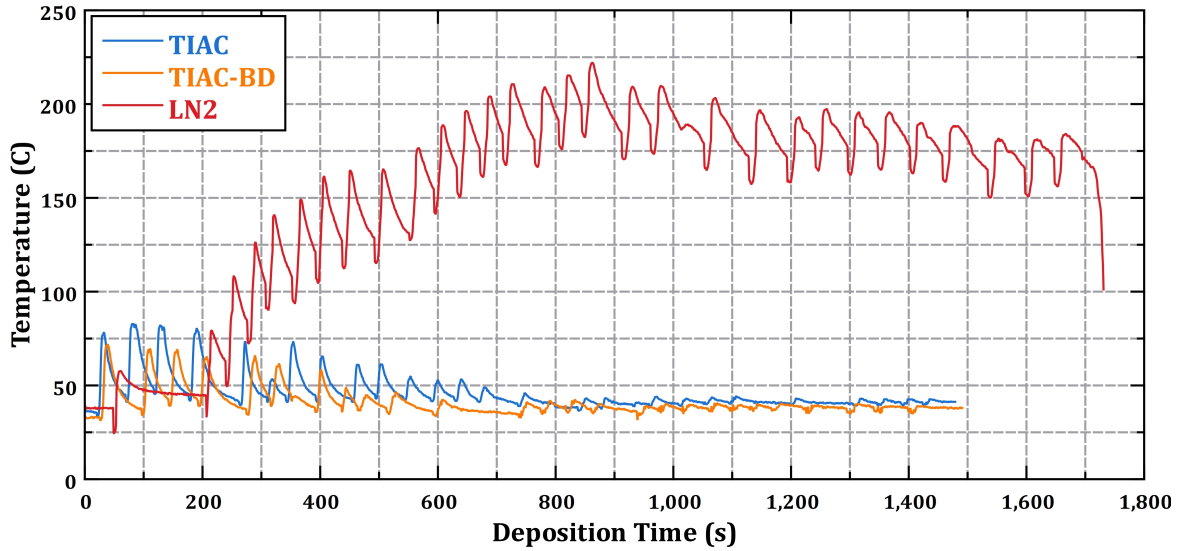


Figure 4.2: Temperature v/s time histories recorded during multi layer WAAM deposition for TIAC, TIAC-BD and LN₂ cooling conditions

Under the Natural Cooling (NC) condition, the peak substrate temperature at TC-A rose progressively from approximately 75 °C after the first layer, reaching a quasi-steady-state plateau of approximately 185 °C beyond layer 20. This progressive rise reflects the increasing thermal resistance of the growing deposit, which lengthens the conduction path between the heat source and the ambient surroundings [10,13]. With 35 deposited layers and a uniform 10 s interpass dwell, the total NC build time was approximately 1300 s. The Passive Cooling (PC) condition substantially reduced the peak substrate temperature relative to NC, with the interpass peaks stabilising in

the range 60 to 80 °C for most of the build. However, as visible in Figure 4.1, the PC baseline drifts upward layer by layer, the inter-peak trough temperature rising from ≈ 22 °C to ≈ 50 °C over 35 layers, confirming that static water without active replenishment cannot prevent progressive heat accumulation in the substrate. The total build time under PC was approximately 1350 s, the same as NC.

Both active cooling conditions (A20 and A10) produced substantially lower and spatially stable substrate temperatures than NC and PC. A20, with fresh water replenished to maintain a LEWD of 20 mm, reduced the plateau peak substrate temperature to approximately 40 °C, while A10 with the water raised to a LEWD of 10 mm achieved a further reduction to approximately 27 °C, near ambient. Critically, both profiles are essentially flat across all layers (Figure 4.1), demonstrating that continuous fresh water replenishment eliminates progressive heat accumulation entirely. The significantly higher thermal conductivity of water ($k \approx 0.6 \text{ W m}^{-1}\text{K}^{-1}$) relative to still air ($k \approx 0.026 \text{ W m}^{-1}\text{K}^{-1}$) accounts for the substantially enhanced heat extraction at the substrate underside. Build times were approximately 900 s for both A20 (32 layers) and A10 (33 layers), shorter than NC due to the smaller layer count resulting from the higher effective layer heights achieved under active cooling.

The TIAC and TIAC-BD conditions (data presented in Figure 4.2) combined substrate immersion at LEWD = 10 mm with a directed inter-layer impingement jet. Despite operating at the higher Phase 2 heat input ($Q = 0.508 \text{ kJ/mm}$), both conditions achieved stable substrate temperatures. TIAC settled to approximately 42 to 50 °C and TIAC-BD to 38 to 42 °C after an initial transient of five to six layers. In contrast, the LN₂ condition also under Phase 2 heat input produced the highest substrate level thermal accumulation among all Phase 2 conditions, with the TC-A profile rising progressively to approximately 200 to 220 °C and exhibiting the largest layer to layer amplitude variation of all seven conditions. This behaviour arises because TC-A, drilled into the substrate underside, receives no direct benefit from cryogen applied to the top of the growing deposit; the large cycle to cycle oscillations reflect the batch wise, manual nature of cryogen delivery and the rapid boil-off of LN₂ upon contact with the hot deposit surface.

4.1.2 Effect of Cooling Techniques on Heat Accumulation

To quantify heat accumulation, the Heat Accumulation Index (HAI) is defined as the arithmetic mean of the peak interpass temperatures recorded at TC-A over all deposited layers:

$$\text{HAI} = \frac{1}{N} \sum_{i=1}^N T_{\text{peak},i} \quad (4.1)$$

where $T_{\text{peak},i}$ is the peak interpass temperature recorded at TC-A after layer i and N is the total number of deposited layers. The HAI values for all seven conditions are compared in Figure 4.3.

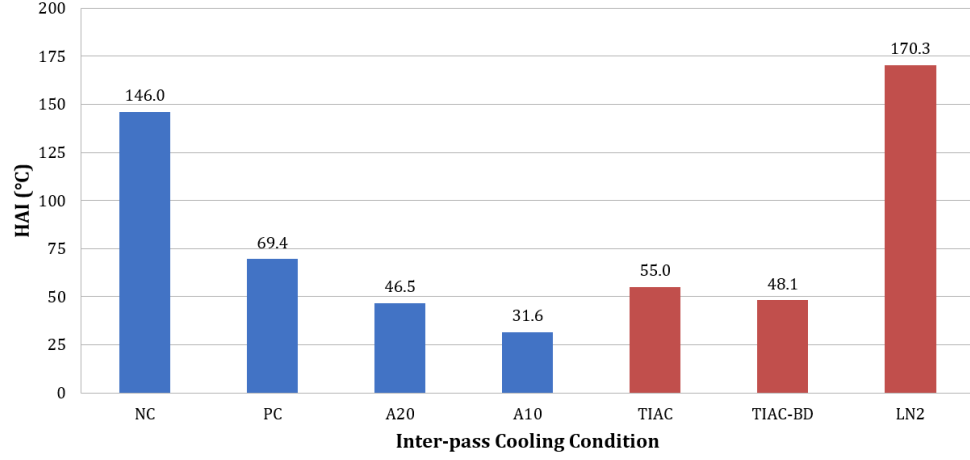


Figure 4.3: Heat Accumulation Index (HAI) for all seven interpass cooling conditions. Lower HAI indicates more effective interpass heat management

The computed HAI values are: NC = 145.9 °C, PC = 69.4 °C, A20 = 46.5 °C, A10 = 31.6 °C, TIAC = 55.0 °C, TIAC-BD = 48.1 °C, and LN₂ = 170.3 °C, as illustrated in Figure 4.3. Within Phase 1, the trend NC > PC > A20 > A10 confirms the progressive improvement achieved by moving from free convection to substrate attached passive elements to active liquid mediated cooling; A10 achieves a 78.3% reduction in HAI relative to NC. Within Phase 2, the trend LN₂ > TIAC > TIAC-BD reflects the fundamentally different heat extraction mechanisms: TC-A, drilled into the substrate, registers no direct benefit from cryogen applied to the deposit surface, so the LN₂ substrate level HAI (170.3 °C) is dominated by conduction from the high heat input Phase 2 arc; whereas TIAC and TIAC-BD actively extract heat from the substrate via the immersion water, maintaining substrate level HAI values of 55.0 and 48.1 °C respectively despite the 2.9× higher heat input. The additional reduction achieved by TIAC-BD relative to TIAC (12.5%) is attributed to the more symmetric thermal history of bidirectional deposition, which eliminates the start-end heat accumulation asymmetry present in unidirectional passes.

The heat accumulation index was chosen as the primary thermal descriptor in preference to a single instant temperature because heat accumulation is a cumulative phenomenon that no instantaneous value captures. The peak temperature of any one layer reflects only the arc passing at that moment and varies with layer position, the temperature at the next arc strike depends on the dwell that happened to precede it and neither distinguishes a deposit that reached a high temperature briefly from one that remained warm throughout. By averaging the peak interpass temperature over

every layer of the build, the index reduces the whole thermal history of a condition to a single scalar that rises both with the severity of the thermal excursions and with their persistence across the build. That property is what makes it useful here: it permits the seven conditions to be ranked on one axis, correlated directly against the measured grain size, tensile response and texture and normalised by heat input so that conditions deposited at different energy inputs can be compared on thermal efficiency.

The limitations attached to the index are that it is computed from a single thermocouple drilled into the substrate, so it describes the substrate level thermal state and not the full field state of the deposit; this is why the cryogenic condition returns the highest index in the matrix despite applying the coldest medium. Being an average of peak values, the index is insensitive to the shape of the thermal history and cannot distinguish a short severe excursion from a longer mild one of the same mean. It is a temperature and not a rate, so it does not by itself capture the cooling rate at the solidification front that governs the columnar to equiaxed transition, for which the cooling rate is reported separately. Its numerical value depends on the thermocouple location and the layer count, so index values are comparable within this study but are not directly comparable with values computed under a different instrumentation scheme unless that scheme is also reported.

4.1.3 Effect of Layer Edge to Water Distance (LEWD)

A parametric investigation of the influence of the layer edge to water distance (LEWD) was conducted by comparing the A20 and A10 conditions, in which the water level was set at 20 mm and 10 mm below the top surface of the most recently deposited layer, respectively. The effect is directly visible in Figure 4.1: the A10 profile (green) runs consistently below the A20 profile (red) across all layers, and crucially, both profiles remain flat throughout the build, confirming that heat accumulation is fully suppressed under continuous fresh water replenishment regardless of the LEWD value tested.

Reducing the LEWD from 20 mm to 10 mm reduced the plateau substrate temperature from approximately 40 °C to 27 °C (a 33% drop). A shorter conduction path between the deposit and the water increases the effective heat transfer coefficient at the interface. However, reducing LEWD below 10 mm risks direct contact between cooling water and the active molten pool, causing spattering, hydrogen porosity and arc instability. In this work, LEWD = 10 mm was the practical lower limit for stable deposition.

4.1.4 Water Temperature Monitoring

The temperature of the cooling water in the work tank was monitored continuously throughout deposition using TC-B, positioned in the water volume. As shown in

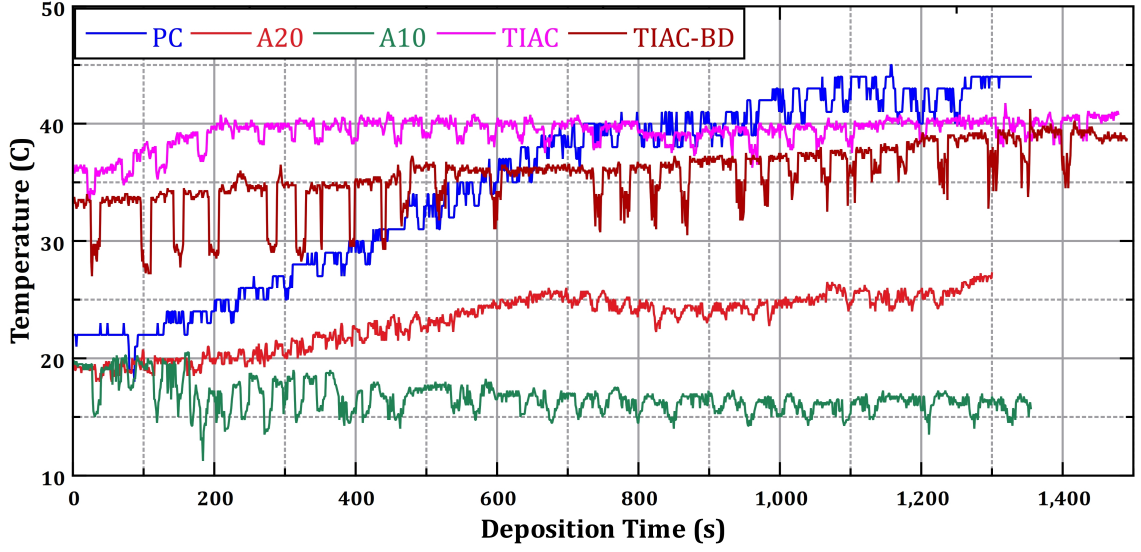


Figure 4.4: Cooling water temperature recorded during WAAM deposition for PC, A20, A10, TIAC and TIAC-BD

Figure 4.4, the water temperature under PC rose monotonically from approximately 20 °C to 45 °C over the full build, confirming that the static (non replenished) water volume progressively absorbs heat and cannot function as a stable thermal sink. In contrast, all active replenishment conditions (A20, A10, TIAC, TIAC-BD) showed an initial transient followed by a quasi-steady plateau, demonstrating that continuous fresh water supply achieves thermal equilibrium in the cooling circuit. A10 recorded the lowest and most stable water temperature (≈ 15 to 20 °C), attributable to both the higher flow rate and the shorter conduction path to the deposit. A20, TIAC and TIAC-BD stabilised in the range 35 to 42 °C. LN₂ is excluded from Figure 4.4 as cryogen is applied directly to the deposit surface and does not flow through the water circuit.

4.2 Geometric Characteristics

4.2.1 Wall Geometry and Profile

The as-built morphology of the thin wall specimens deposited under each cooling condition is shown in Figure 4.5. Visual inspection reveals a clear dependence of wall regularity on the cooling strategy: the NC and PC walls exhibit noticeable lateral bulging at the lower layers and increasing width irregularity towards the top, while the A20, A10, TIAC, TIAC-BD and LN₂ walls display a more uniform profile from base to top.

4.2.2 Layer Height (LH)

Layer height is defined as the measured inter layer distance on the transverse cross section macrograph. Under the NC condition, the average layer height was 1.73 mm. Progressive heat accumulation under NC reduces the local yield strength and increases the fluidity of the molten pool, causing the deposit to spread laterally rather than build vertically, which manifests as a reduction in effective layer height in the upper layers. This thermal softening effect has been reported previously for GMAW based WAAM of aluminium alloys and is well correlated with the rising interpass temperature [12,13]. Under the TIAC condition, the average LH increased substantially to 2.60 mm, and TIAC-BD achieved the highest LH of all conditions at 2.83 mm. It should be noted that this inter phase LH increase reflects both the effect of active cooling (which stabilises solidification) and the $2.9\times$ higher heat input per unit length of Phase 2 (0.508 kJ/mm versus 0.174 kJ/mm), which deposits more material per unit length of travel. Within Phase 1, the modest LH differences (NC: 1.73 mm to A20: 1.90 mm) are attributable solely to the cooling effect. The LH values for all conditions are summarised in Table 4.1.

4.2.3 Total Wall Width (TWW) and Effective Wall Width (EWW)

Total wall width (TWW) was measured as the outer-to-outer dimension of the deposited bead at mid height cross sections, while effective wall width (EWW) was measured as the usable flat central region between the outer surface undulations. Under NC, the average TWW was 18.82 mm and the EWW was 13.85 mm, giving an EWW/TWW ratio of 0.74. This relatively low ratio reflects the pronounced surface waviness and bead-to-bead width variation associated with heat accumulation. Active cooling progressively reduced TWW as the constrained thermal input prevented pool spreading: A20 (16.25 mm), A10 (16.28 mm), TIAC (14.66 mm), and TIAC-BD (12.95 mm). The highest EWW/TWW ratio was achieved by TIAC-BD (0.88), followed by A10 (0.86), indicating that these conditions maximise the proportion of deposited material that lies within the usable geometric envelope. The improvement in EWW/TWW ratio directly translates to higher specimen yield per deposited wall and is of practical significance for near net shape component manufacture.

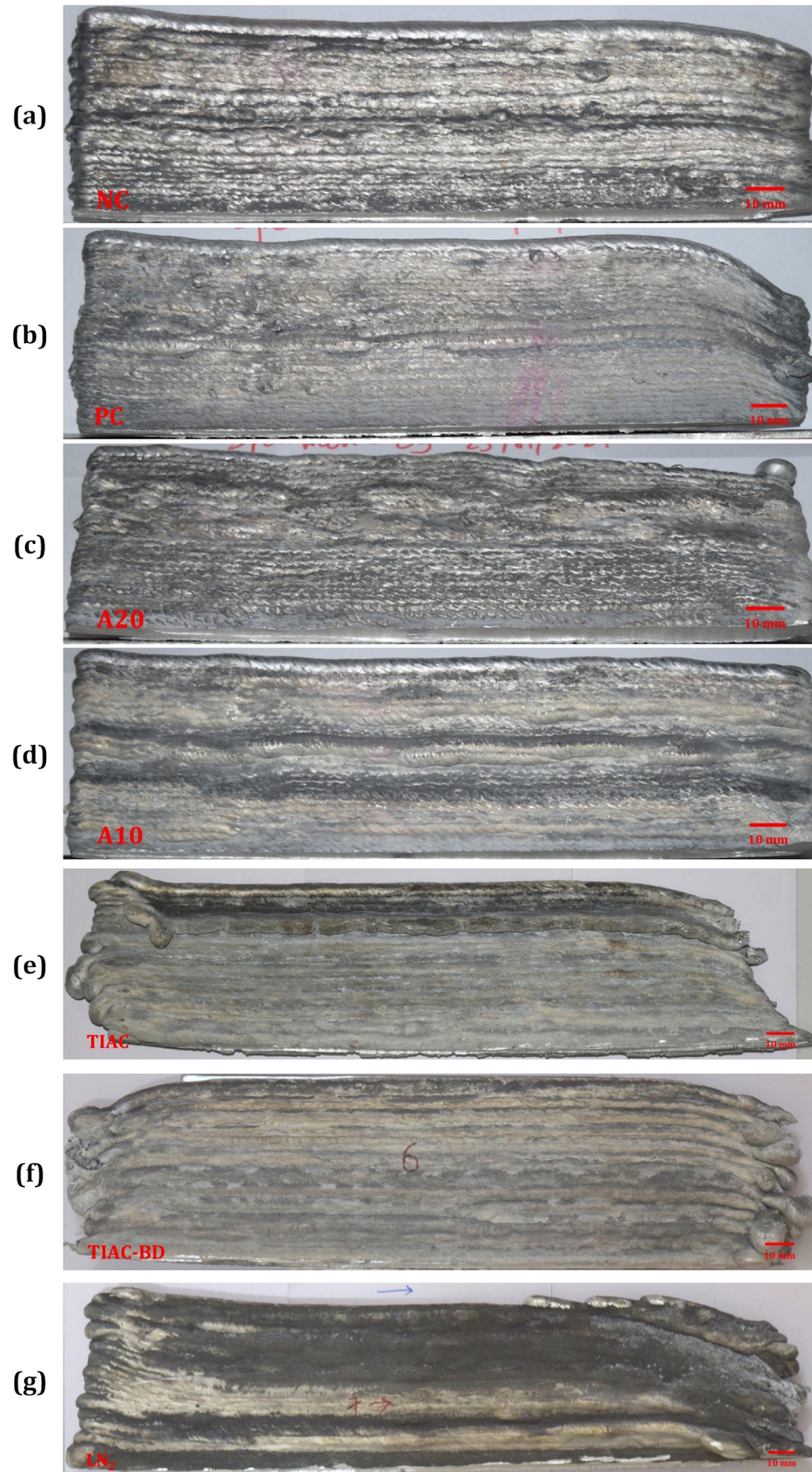


Figure 4.5: Photographs of as-built thin wall specimens fabricated under (a) NC, (b) PC, (c) A20, (d) A10, (e) TIAC, (f) TIAC-BD and (g) LN₂ cooling conditions

4.2.4 Surface Waviness (SW)

Surface waviness quantifies the peak to valley variation in wall width along the build height and is a critical measure of the geometric consistency of WAAM deposits. High waviness imposes tighter post processing allowances and increases the risk of machined surfaces intersecting internal porosity. The layer band structure and surface profile visible in the cross section macrographs (Figure 4.6) provide a qualitative indication of the SW trend that is quantified in Table 4.1.

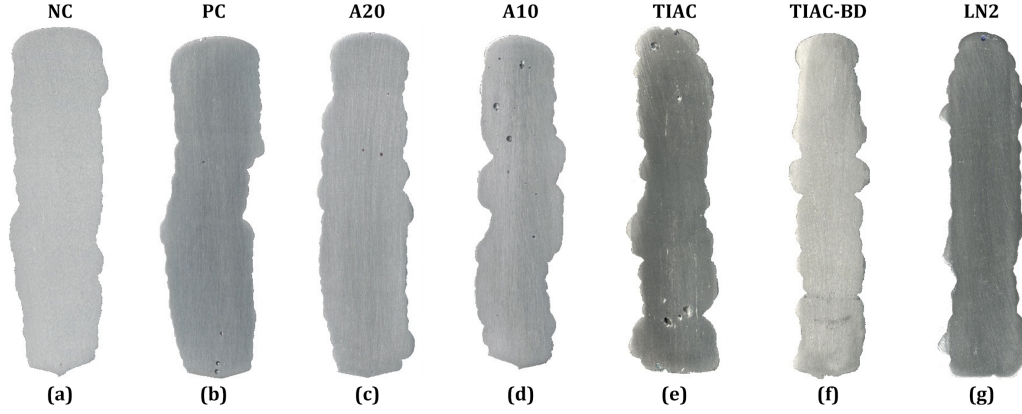


Figure 4.6: Transverse cross section macrographs of as-built WAAM thin wall specimens under all seven interpass cooling conditions

The NC condition produced the highest surface waviness of 2.49 mm, attributed to the progressive increase in thermal softening and pool spreading as heat accumulates in the deposit. Among the water based Phase 1 conditions, SW decreased with increasing cooling intensity: PC gave 1.71 mm, A20 gave 1.84 mm, and A10 gave 1.12 mm. A10 achieved the lowest SW among the immersion only conditions, a reduction of 55.0% relative to NC, confirming that a shorter conductive path to the coolant is the primary driver of surface uniformity. Within Phase 2, TIAC produced SW of 1.63 mm, while TIAC-BD achieved the overall lowest SW of 0.78 mm across all seven conditions, a 68.7% reduction relative to NC demonstrating that the combination of tandem cooling and bidirectional deposition most effectively suppresses pool width variability by distributing the arc restart energy symmetrically along the wall. The LN₂ condition produced a SW of 1.96 mm, the highest among Phase 2 conditions, suggesting that despite the aggressive cooling capacity of cryogen, the batch wise and manual application does not adequately control pool width consistency on a layer by layer basis. This agrees with Cunningham et al. [13], who reported that sustained interpass temperature control matters more than peak cooling capacity for wall width consistency in GMAW-WAAM.

4.2.5 Deposition Efficiency (DE)

Deposition efficiency, calculated as the ratio of the effective cross sectional area to the total deposited cross sectional area (Section 3.9), ranged from 74% (NC) to 88% (TIAC-BD). The lower DE under NC is attributed to the greater proportion of material lost to surface undulations as pool spreading increases with heat accumulation. Active cooling progressively improved DE: PC (82%), A20 (77%), A10 (86%), TIAC (78%), and LN₂ (75%). The highest DE of 88% achieved by TIAC-BD reflects the combined benefit of the TIAC cooling intensity and the geometric uniformity imposed by the bidirectional deposition strategy. The DE values for all conditions are included in Table 4.1.

Table 4.1: Summary of geometric quality parameters for all seven cooling conditions

Condition	LH (mm)	TWW (mm)	EWL (mm)	SW (mm)	DE (%)
NC	1.73	18.82	13.85	2.49	74
PC	1.72	18.68	15.26	1.71	82
A20	1.90	16.25	12.57	1.84	77
A10	1.84	16.28	14.04	1.12	86
TIAC	2.60	14.66	11.40	1.63	78
TIAC-BD	2.83	12.95	11.38	0.78	88
LN ₂	2.30	15.89	11.98	1.96	75

4.2.6 Effect of Deposition Strategy

A comparison of unidirectional and bidirectional deposition strategies under the TIAC condition (TIAC and TIAC-BD) revealed that the bidirectional strategy produced more symmetrical wall profiles and a more uniform LH distribution along the wall length. The end taper effect in which the start end of a unidirectionally deposited wall exhibits a lower layer height due to arc restart heat concentration was effectively eliminated in the TIAC-BD condition. The average LH under TIAC-BD was 2.83 mm compared to 2.60 mm for unidirectional TIAC, an increase of 8.8%, while the TWW decreased from 14.66 mm to 12.95 mm, reflecting the tighter pool control achieved by the symmetric thermal history. The SW improved from 1.63 mm (TIAC) to 0.78 mm (TIAC-BD) and DE from 78% to 88%, confirming that the bidirectional strategy distributes the arc energy more uniformly and is the preferred deposition strategy when geometric consistency is the primary objective.

4.3 Mechanical Properties

4.3.1 Tensile Properties

Anisotropy is a first order design concern in wire arc additive manufacturing rather than an incidental observation, and it arises from the geometry of the process itself. Deposition proceeds layer upon layer along a fixed build direction, and each new layer partially remelts the one beneath it, so that grains grow epitaxially along the thermal gradient aligned with the build direction and form elongated columnar grains that traverse several layers. The same layer by layer construction creates planar interfaces at regular spacing along the build direction, and these interfaces concentrate the features that limit performance, namely entrapped oxide, hydrogen porosity and any incompletely fused region. A specimen loaded along the build direction therefore presents its weakest features perpendicular to the load, while a specimen loaded along the deposition direction does not. The practical consequence is that a component cannot be qualified on a single tensile figure, and a designer must know the property in the direction in which the part will actually be loaded; because it is the lower of the two directional values that sets the allowable design stress, reducing the difference between directions is as valuable as raising the mean. The specimens were extracted in four orientations in this work and that the anisotropy ratio is reported alongside the absolute properties.

Tensile results for the four specimen orientations defined in Section 3.8 are presented below. Load v/s extension curves for the TV and TB orientations are shown in Figure 4.7 and Figure 4.8 respectively, and the UTS across all four orientations is compared in Figure 4.9. Full tensile data are tabulated in Table 4.2. In the vertical direction (TV), specimens experienced the layer band interfaces as the primary microstructural feature intersecting the gauge length. Under NC, the vertical UTS was 264.3 MPa, YS was 40.3 MPa, and elongation was 21.8%. The relatively low YS is consistent with the coarser columnar grain structure and inter-layer oxide bands that limit dislocation strengthening. Across the remaining six conditions the vertical UTS spans 228.1 to 276.6 MPa, the extremes being A10 and A20 respectively, with the individual values given in Table 4.2.

The non-monotonic trend reflects competing influences: A20 raises the vertical UTS by improving inter-layer bonding and reducing oxide entrapment, while A10 shows a lower vertical UTS despite a finer grain structure, accompanied by a marked reduction in elongation (12.2% versus 21.8% for NC), suggesting a more brittle failure mode in the build direction.

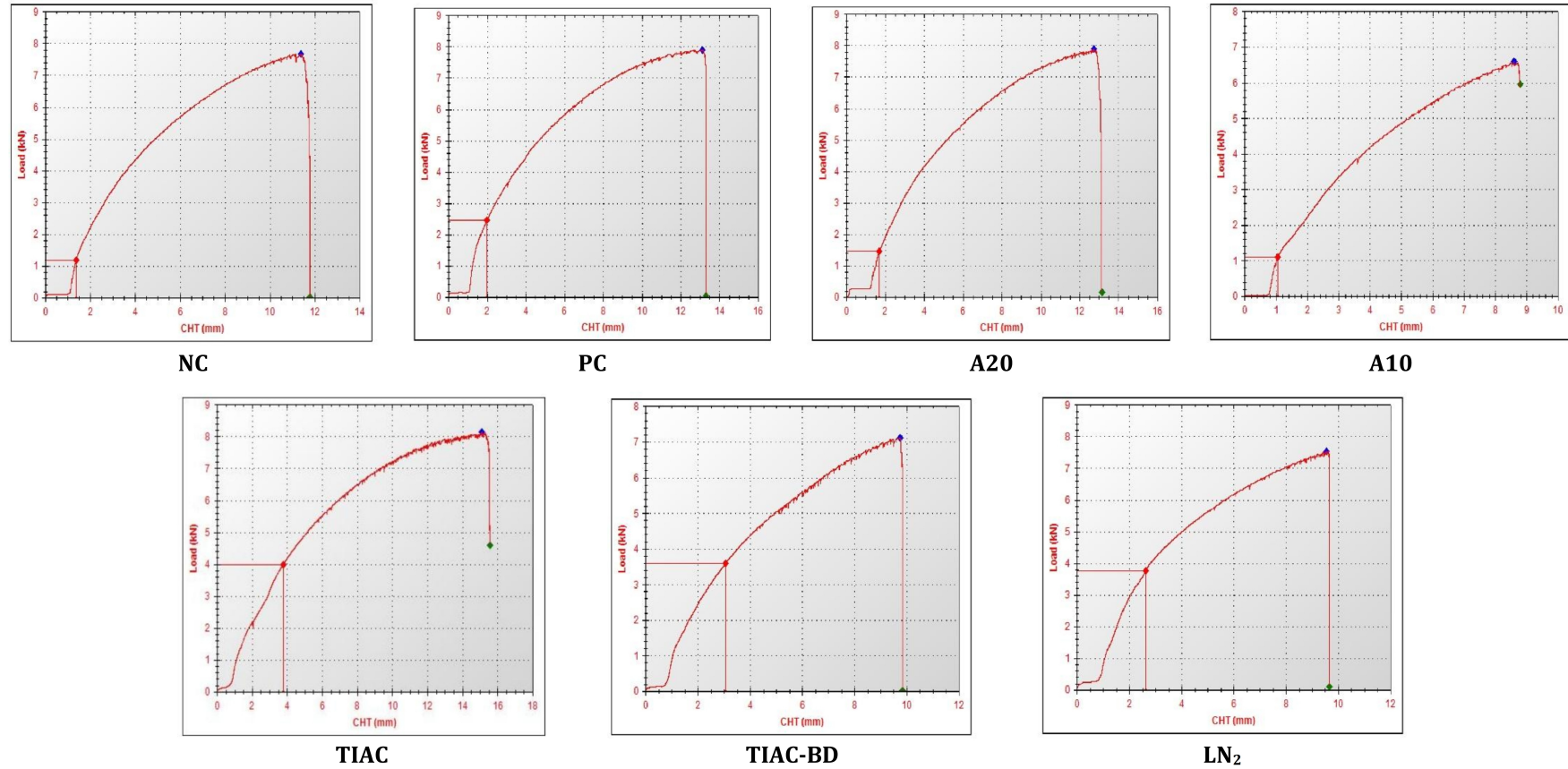


Figure 4.7: Load v/s extension curves of vertical direction (TV) tensile specimens under all seven interpass cooling conditions recorded at 1 mm/min with a 25 mm gauge length

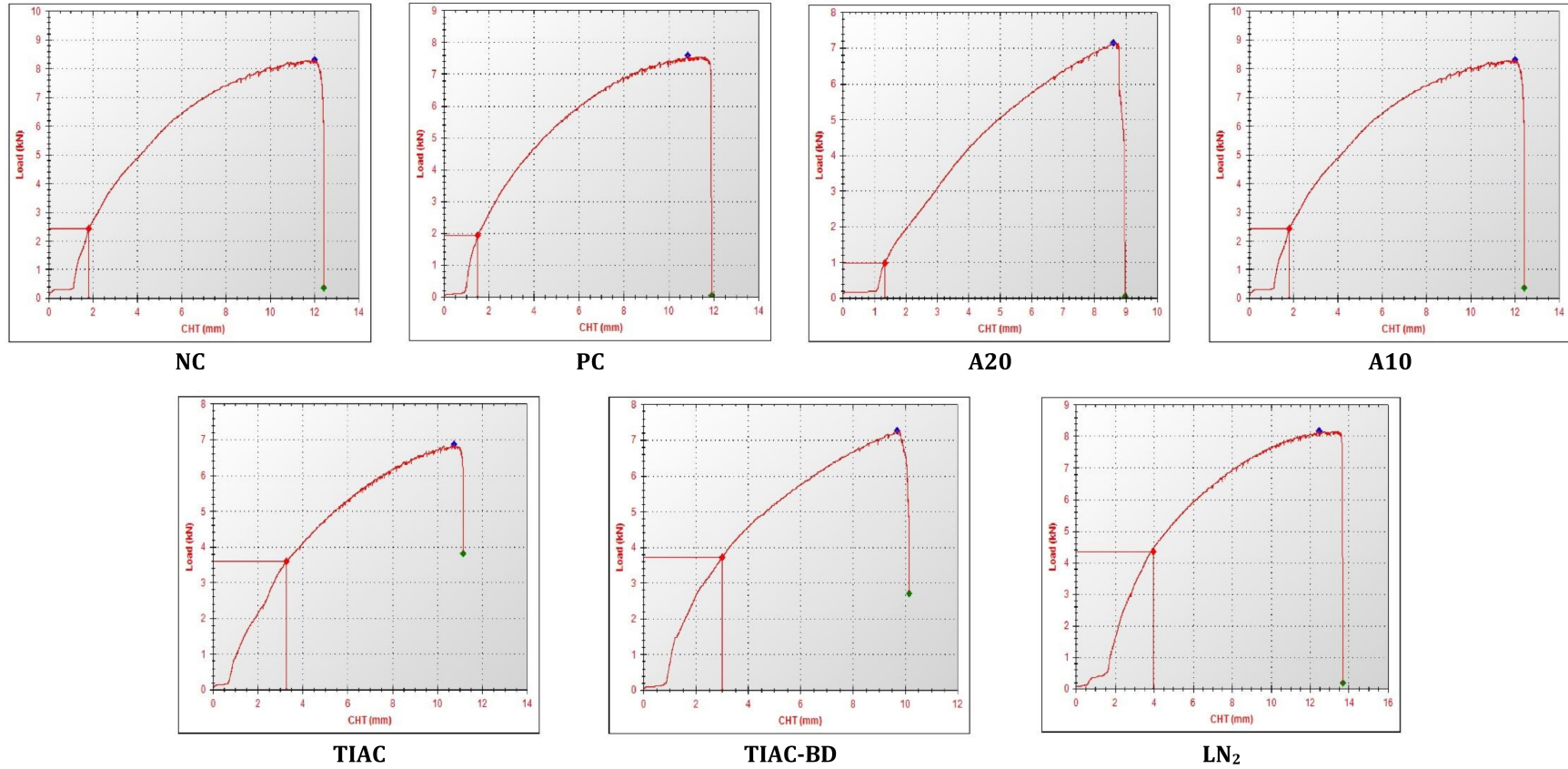


Figure 4.8: Load v/s extension curves of bottom zone horizontal (TB) tensile specimens under all seven interpass cooling conditions recorded at 1 mm/min with a 25 mm gauge length

At a LEWD of 10 mm, the water surface is in close proximity to the active arc and the risk of hydrogen pick-up from water vapour into the aluminium melt is elevated relative to larger LEWD values. Hydrogen induced porosity concentrated at layer band interfaces is a well documented mechanism for reducing vertical direction tensile performance in aluminium WAAM [12], as the TV specimen gauge length crosses multiple inter-layer boundaries where pores preferentially nucleate. The vertical UTS is therefore more sensitive to inter-layer porosity than the horizontal orientations, which sample predominantly intra-layer material. This interpretation is consistent with the observed combination of low TV UTS and low elongation under A10, and the effect was not quantified by porosity measurement in the present work and is identified as a limitation and a recommendation for future work. TIAC and LN₂ achieve intermediate vertical UTS values with markedly higher YS (129.2 and 123.4 MPa respectively), attributable to grain refinement and solid solution Mg retention under fast cooling, discussed further in Section 4.5.

Horizontal specimens (TT and TB) are less sensitive to inter-layer interfaces and more influenced by the intra-layer microstructure. The TB orientation is used to define the anisotropy ratio AR_t because the bottom zone experiences the highest number of reheating cycles and is closest to the cooling medium. Under NC, the TB UTS was 296.5 MPa with elongation 21.7%, and the TT UTS was 284.5 MPa with elongation 26.0%. Inclined specimens (TIN, 45°) exhibited the highest elongation 28.4% under NC, reflecting the shear dominated stress state at this orientation. The comparative UTS across all four orientations (Figure 4.9) shows that horizontal orientations (TB, TT) dominate under NC, PC and A10, while the vertical direction yields the highest UTS under A20, consistent with the AR_t values in Table 4.2 ranging from 0.88 (A20) to 1.17 (A10).

The two anisotropy metrics used throughout this chapter are formally defined as:

$$AR_t = \frac{UTS_{TB}}{UTS_{TV}} \quad (4.2)$$

$$AI_t = \frac{UTS_{\max}}{UTS_{\min}} \quad (4.3)$$

where AR_t (Tensile Anisotropy Ratio) is the ratio of bottom zone horizontal to vertical UTS, capturing the horizontal–vertical strength balance and AI_t (Tensile Anisotropy Index) is the ratio of the maximum to minimum UTS across all four orientations, providing a broader measure of total directional dependence. Full data for all conditions are provided in Table 4.2.

The AR_t and AI_t values reveal distinct anisotropy behaviours across the seven conditions. Under NC, $AR_t = 1.12$ confirms that the bottom horizontal specimens

(TB = 296.5 MPa) are stronger than the vertical (TV = 264.3 MPa), consistent with the directional columnar grain structure that weakens the build height direction. $AI_t = 1.12$ indicates moderate total anisotropy across all four orientations. PC reduced both metrics substantially ($AR_t = 1.03$, $AI_t = 1.03$), bringing all four orientations within 3% of each other, the closest to isotropic behaviour among Phase 1 conditions reflecting the moderating effect of passive cooling on columnar grain growth without the aggressive thermal transients of active cooling. A20 produced the most anomalous result: $AR_t = 0.88$ (TV > TB), indicating that the vertical specimens are stronger than the bottom horizontal, which is the reverse of the NC trend. This reversal arises because the A20 active cooling suppresses pool spreading and promotes a tighter, more columnar deposit in the deposition direction, elevating TV UTS to 276.6 MPa while simultaneously reducing TB UTS to 242.4 MPa. Despite its low AR_t , A20 recorded the highest AI_t of all conditions (1.20), driven by the low TIN UTS (230.5 MPa) confirming that cooling induced constraint effects produce significant strength scatter across orientations even when the horizontal-vertical ratio appears favourable. A10 showed the opposite trend: the highest AR_t (1.17) with TB = 265.8 MPa substantially exceeding TV = 228.1 MPa, suggesting that the more aggressive LEWD = 10 mm cooling introduces residual stresses that selectively reduce vertical strength.

Among Phase 2 conditions, TIAC ($AR_t = 0.96$, $AI_t = 1.09$) and TIAC-BD ($AR_t = 0.99$, $AI_t = 1.08$) achieved the most balanced tensile response. TIAC-BD yielded the lowest AI_t of all seven conditions, confirming that the combination of tandem immersion cooling and bidirectional deposition minimises crystallographic texture and promotes a near isotropic UTS distribution across all four orientations. LN₂ produced intermediate anisotropy ($AR_t = 1.08$, $AI_t = 1.09$), comparable to TIAC, consistent with the grain refinement observed in Section 4.5 but limited by the less uniform thermal environment introduced by batch wise cryogen application. The drop in AI_t from 1.20 (A20) and 1.17 (A10) in Phase 1 to 1.08–1.09 in Phase 2 shows that more intensive cooling progressively disrupts the directional columnar solidification front and reduces mechanical anisotropy.

To contextualise these results within the broader WAAM-ER5356 literature, the UTS values reported in the present study (228 to 297 MPa across all conditions and orientations) are consistent with the range documented in comparable published work. Derekar et al. [114] reported horizontal UTS values of 226 to 273 MPa for GMAW-WAAM ER5356 without active cooling, with significant vertical direction degradation at high interpass temperatures, a trend consistent with the NC baseline results of the present study. Wiczorowski et al. [50] reported UTS of 256 to 294 MPa for CMT-WAAM ER5356 thin walls, with the bottom zone exhibiting higher strength due to repeated reheating cycles, consistent with the TB > TV trend observed here under NC and PC.

Table 4.2: Tensile properties (UTS, YS and %El) for all 4 orientations for different cooling conditions

Condition	TV (90°)			TIN (45°)			TB (0°)			TT (0°)			AR _t	AI _t
	UTS	YS	%El	UTS	YS	%El	UTS	YS	%El	UTS	YS	%El		
	(MPa)	(MPa)		(MPa)	(MPa)		(MPa)	(MPa)		(MPa)	(MPa)			
NC	264.3	40.3	21.8	283.6	26.8	28.4	296.5	86.6	21.7	284.5	73.0	26.0	1.12	1.12
PC	267.5	82.9	24.5	271.4	87.0	22.0	275.0	69.9	20.8	271.9	144.7	27.1	1.03	1.03
A20	276.6	51.1	23.7	230.5	24.6	12.9	242.4	33.1	12.3	265.0	59.1	17.0	0.88	1.20
A10	228.1	37.8	12.2	237.6	62.4	14.2	265.8	72.1	16.6	251.2	90.9	15.0	1.17	1.17
TIAC	264.2	129.2	24.3	242.5	118.9	17.0	254.0	132.2	20.0	256.4	141.1	23.2	0.96	1.09
TIAC-BD	237.6	119.4	18.0	249.3	119.1	21.2	235.3	120.5	16.4	254.4	120.1	27.4	0.99	1.08
LN ₂	247.1	123.4	18.8	257.6	133.2	20.3	267.6	142.3	27.6	268.8	138.8	26.8	1.08	1.09

Table 4.3: Charpy V-notch impact energy in both orientations for different cooling condition

Condition	CH (0°)		CV (90°)		AR _i
	(J)	(J/cm ²)	(J)	(J/cm ²)	
NC	7.2	14.4	10.1	20.2	1.40
PC	6.6	13.2	9.1	18.2	1.38
A20	7.0	14.0	10.2	20.4	1.46
A10	6.4	12.8	6.6	13.2	1.03
TIAC	8.3	16.6	10.2	20.4	1.23
TIAC-BD	9.0	18.0	7.0	14.0	0.78
LN ₂	7.0	14.0	6.8	13.6	0.97

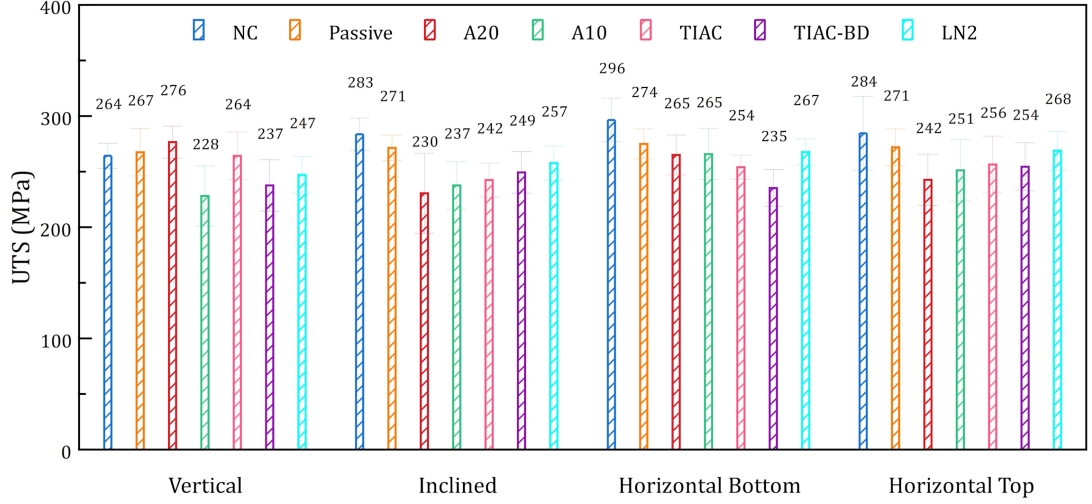


Figure 4.9: Comparison of ultimate tensile strength (UTS) across all four specimen orientations (TV, TIN, TB, TT) for all seven interpass cooling conditions

Chandra et al. [115] reported UTS of 286 MPa, YS of 142 MPa and elongation of 23% for robotic GMAW-WAAM ER5356, which is comparable to the Phase 1 horizontal results of the present study. For actively cooled Al–Mg alloys, Ren et al. [116] reported UTS of 321 MPa, YS of 232 MPa and elongation of 28% for CMT-WAAM Al–6.3Mg with controlled interpass temperature values higher than those achieved in the present study, attributable to the higher Mg content (6.3 vs. 5.0 wt.%) and the CMT process, which deposits material at lower heat input than the GMAW-based process used here. The tensile properties in this study are competitive with published WAAM-ER5356 results. TIAC and TIAC-BD match or exceed the property levels reported for naturally or passively cooled deposits.

The mechanism by which the cooling strategy acts on this anisotropy follows from the process structure property chain set out in Section 4.9. The columnar morphology responsible for the directional difference forms when the thermal gradient ahead of the solidification front is high relative to the growth rate and the deposit spends sufficient time in the coarsening range for the columnar grains to extend across successive layers. Raising the heat extraction rate shortens that residence time and promotes the columnar to equiaxed transition described in Section 4.5.2, and an equiaxed structure presents no preferred orientation to the load. The same reduction in thermal exposure limits the coarsening of the grain boundary second phase networks that provide the preferential fracture path in build direction specimens, and the accompanying texture randomisation reported in Section 4.6 removes the crystallographic contribution to the directional difference. The three effects act in the same sense, so that the reduction in the anisotropy ratio under the actively cooled conditions relative to natural cooling is

their combined result.

4.3.2 Impact Strength

Charpy V-notch impact energy values for all seven cooling conditions in both vertical (CV) and horizontal (CH) orientations are presented in Figure 4.10. The impact strength is particularly sensitive to the inter-layer microstructure because the notch tip in vertically loaded specimens is positioned to propagate a crack through the layer-band interfaces. Detailed values including normalised impact energy (J/cm^2) and the anisotropy ratio AR_i are listed in Table 4.3.

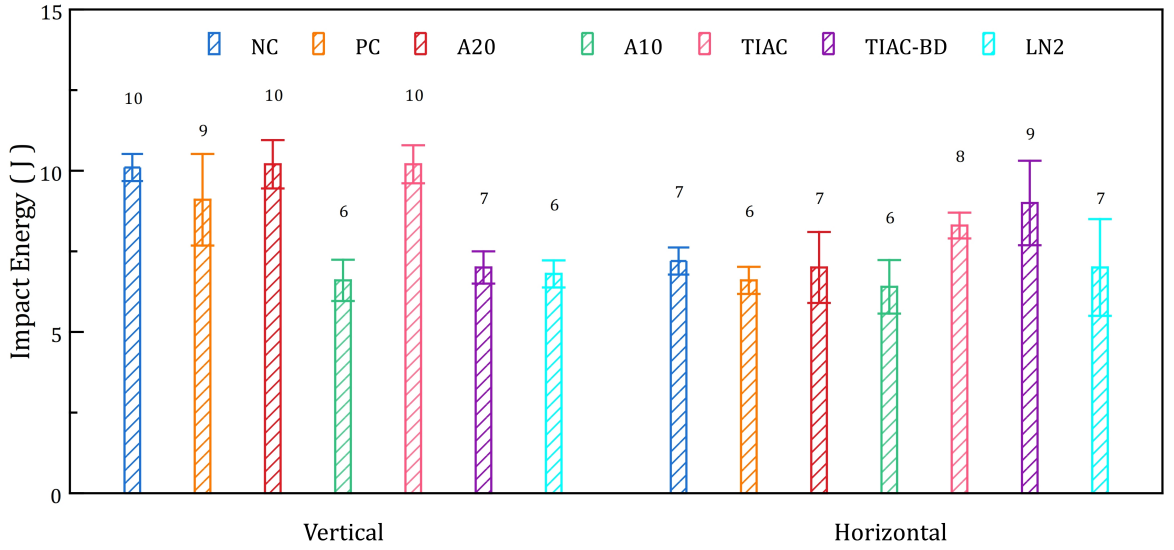


Figure 4.10: Charpy V-notch impact energy (J) in both orientations (CV, CH) for all seven interpass cooling conditions

To quantify the directional dependence of impact behaviour, the impact anisotropy ratio AR_i is defined as:

$$\text{AR}_i = \frac{E_{\text{CV}}}{E_{\text{CH}}} \quad (4.4)$$

where E_{CV} and E_{CH} are the normalised Charpy impact energies (J/cm^2) in the vertical and horizontal orientations respectively. $\text{AR}_i > 1$ indicates CV dominance (layer-band interfaces absorb more energy than the intra-layer direction), while $\text{AR}_i < 1$ indicates CH dominance.

As seen in Figure 4.10 and Table 4.3, the vertical orientation consistently yields higher impact energy than the horizontal orientation for most conditions, with NC, PC, A20 and TIAC showing pronounced CV dominance. Among all conditions, NC, A20 and TIAC achieved the highest CV impact energy of approximately 10 J (20.2 to $20.4 \text{ J}/\text{cm}^2$), while A10 and LN₂ recorded the lowest CV values of 6.6 and 6.8 J respectively, approaching isotropy ($\text{AR}_i \approx 1.03$ and 0.97). TIAC-BD is the only condition

where CH exceeds CV ($AR_i = 0.78$, CH = 9.0 J versus CV = 7.0 J), reflecting the more homogeneous intra-layer microstructure produced by the bidirectional deposition strategy. TIAC achieved the best combined performance high CV (10.2 J, 20.4 J/cm²) alongside the highest CH among Phase 1 comparable conditions (8.3 J, 16.6 J/cm²) giving $AR_i = 1.23$, the best balance of impact toughness in both orientations. The high CV values under NC and A20 are attributed to the pronounced layer bands acting as crack deflection sites, whereas the finer, more continuous microstructure under A10 and LN₂ offers less crack deflection, reducing CV despite improved material homogeneity. Published Charpy impact data for WAAM-ER5356 are limited; the few available studies report full size specimen energies of 23 to 26 J for CMT-WAAM deposits, which are not directly comparable to the sub-size (55×10×5 mm) normalised values (J/cm²) used in the present study. A direct numerical benchmark for impact toughness is therefore not presented here; the internal comparison across seven conditions remains the primary basis for evaluating the effect of interpass cooling on impact behaviour.

4.3.3 Hardness Distribution

Rockwell superficial hardness (HR15T scale) measurements at five equally spaced locations along the build height revealed a spatial hardness distribution correlated with the thermal history of each layer position. A summary of hardness values at the bottom, mid and top zones for all conditions is provided in Table 4.4.

Under NC, the average hardness was 62.2 HR15T with a range of 60.0–63.4 HR15T, indicating a relatively uniform distribution across the build height, attributable to the consistently slow cooling and mild temperature gradient between layers. The mid-zone hardness (63.2 HR15T) was marginally higher than the top and bottom extremes, consistent with the repeated reheating cycles experienced by mid height layers promoting a slight precipitation strengthening effect in the Al–Mg matrix. Under TIAC, the average hardness of 62.6 HR15T is comparable to NC, but the bottom zone is notably softer (55.4 HR15T) while the mid zone is harder (65.0 HR15T), reflecting the aggressive extraction of heat from the substrate region under tandem cooling which suppresses post solidification precipitation in the lower layers. A20 produced the lowest overall mean hardness (57.1 HR15T) and the most pronounced bottom softening (49.4 HR15T), suggesting that the higher LEWD at 20 mm allows the bottom layers to cool more rapidly through the precipitation temperature range, leaving less time for Mg rich phase formation. The hardness summary for all conditions is given in Table 4.4.

Table 4.4: Rockwell superficial hardness (HR15T) at bottom, mid and top zones for all seven cooling conditions

Condition	Bottom Zone (HR15T)	Mid Zone (HR15T)	Top Zone (HR15T)	Overall Mean (HR15T)
NC	60.0	63.2	61.2	62.2
PC	68.0	62.9	57.6	62.9
A20	49.4	61.9	50.6	57.1
A10	52.4	59.2	60.0	58.0
TIAC	55.4	65.0	62.8	62.6
TIAC-BD	55.5	58.5	63.4	58.9
LN ₂	63.4	62.5	61.6	62.5

4.3.4 Effect of Deposition Strategy on Mechanical Properties

The role of the deposition strategy unidirectional versus bidirectional on mechanical properties was examined by comparing the TIAC and TIAC-BD conditions, which share an identical process parameter set and cooling configuration and differ only in the direction of successive deposition passes. This comparison directly addresses Research Objective 4 and Gap 4 identified in Chapter 2.

The bidirectional deposition of TIAC-BD yielded a vertical UTS of 237.6 MPa and a horizontal bottom UTS of 235.3 MPa, giving an anisotropy ratio $AR_t = 0.99$. The corresponding TIAC values were 264.2 MPa (TV), 254.0 MPa (TB), and $AR_t = 0.96$. The improvement in isotropy under bidirectional deposition is attributed to the more symmetric thermal field experienced by the growing wall: in unidirectional deposition, the start end of the wall consistently experiences the arc restart at an elevated temperature, concentrating heat and promoting directional columnar growth, whereas the bidirectional strategy alternates the restart end on successive layers, distributing the thermal asymmetry more uniformly along the wall length. The resulting grain structure is therefore more homogeneous in the deposition direction, which reduces the directionality of the crystallographic texture and yields a more isotropic stress-strain response.

Charpy V-notch impact energy under TIAC-BD was 14.0 J/cm² in the vertical (CV) orientation and 18.0 J/cm² in the horizontal (CH) orientation, compared to 20.4 J/cm² (CV) and 16.6 J/cm² (CH) for TIAC. The marginal difference between the two conditions confirms that, while bidirectional deposition improves geometric uniformity and marginally reduces mechanical anisotropy, it does not substantially alter the bulk toughness behaviour of the deposit, which is governed primarily by the interpass thermal history and grain refinement both of which are identical for TIAC and TIAC-BD.

HR15T hardness distributions for TIAC-BD showed a narrower bottom-to-top variation of 55.5–63.4 HR15T (spread: 7.9) compared to 55.4–65.0 HR15T (spread: 9.6) for unidirectional TIAC, consistent with the more uniform thermal history imposed by the bidirectional strategy. The reduction in hardness gradient along the build height under TIAC-BD indicates that each layer experiences a more repeatable thermal cycle, which translates to a more homogeneous precipitation state and solid solution content across the wall height. This finding corroborates the geometric uniformity improvement reported in Section 4.2 and strengthens the case for bidirectional deposition as the preferred strategy when property homogeneity along the wall height is a design requirement.

4.4 Chemical Composition Analysis

The bulk chemical compositions of the as deposited walls under all seven cooling conditions, measured by optical emission spectrometry (OES, ASTM E 1251-17a) at Hertz Testing and Training Centre, Ahmedabad, are presented in Table 4.5 alongside the parent ER5356 wire composition.

Table 4.5: Bulk chemical composition (wt.%) of as deposited walls under all seven cooling conditions, compared with the parent ER5356 wire

Condition	Al (%)	Mg (%)	Cr (%)	Mn (%)	Si (%)	Fe (%)
Parent (ER5356)	94.327	4.83	0.180	0.157	0.110	0.285
NC	94.7	4.34	0.200	0.14	0.100	0.130
PC	94.7	4.43	0.210	0.14	0.097	0.120
A20	94.7	4.36	0.200	0.14	0.100	0.130
A10	94.6	4.43	0.200	0.14	0.100	0.160
TIAC	94.8	4.40	0.150	0.14	0.095	0.110
TIAC-BD	94.9	4.35	0.150	0.14	0.092	0.110
LN ₂	94.9	4.33	0.150	0.14	0.093	0.110

AWS A5.10 limits: Mg 4.5–5.5%, Cr 0.05–0.20%, Mn $\leq$ 1.0%, Si $\leq$ 0.25%, Fe $\leq$ 0.40%

All seven deposited conditions conform to the AWS A5.10/A5.10M specification for ER5356, confirming that the WAAM process preserves the alloy classification across all cooling strategies. The most significant observation is a consistent reduction in Mg content relative to the parent wire (4.83 wt.%) in all deposited conditions, ranging from 4.33 wt.% (LN₂) to 4.43 wt.% (PC and A10). This Mg loss of 8–10% relative to the parent is attributed to preferential Mg vapourisation in the high temperature arc plume, a well documented phenomenon in GMAW-based WAAM of Al–Mg alloys arising from the significantly lower boiling point of magnesium ($T_{b,Mg} \approx 1091$ °C at

1 atm) relative to aluminium ($T_{b,Al} \approx 2519\text{ }^{\circ}\text{C}$), which renders Mg substantially more volatile than Al at the arc temperatures encountered during GMAW deposition [12,30]. Within the deposited conditions, the variation in Mg content is small (4.33–4.43 wt.%), confirming that the interpass cooling strategy does not substantially alter the arc characteristics or the Mg burn-off rate. NC and LN₂ exhibit the lowest deposited Mg (4.34 and 4.33 wt.% respectively), while PC and A10 retain the highest Mg among all conditions (4.43 wt.%). The marginally lower Mg under NC is consistent with the higher and more prolonged thermal exposure of the deposit, which provides additional opportunity for solid state Mg diffusion to the surface and subsequent evaporation between passes. The slightly lower Mg under LN₂ despite rapid interpass cooling may reflect the higher Phase 2 heat input ($Q = 0.508\text{ kJ/mm}$), which increases the arc energy per unit length and the peak arc zone temperature, elevating the Mg vapourisation rate. The Mn, Si and Fe contents remain essentially invariant across all conditions, as expected for elements with significantly higher vapour pressures and lower diffusivities than Mg under the present process conditions.

4.5 Microstructural Analysis

4.5.1 Grain Morphology

Optical micrographs of etched transverse cross sections at $100\times$ and $200\times$ magnification for all seven cooling conditions are presented in Figure 4.11 and Figure 4.12 respectively. The micrographs reveal a pronounced dependence of grain morphology on the interpass cooling strategy, with the predominant grain type transitioning systematically from coarse columnar (NC, PC) towards fine equiaxed (TIAC, TIAC-BD, LN₂) as the cooling intensity increases.

At $100\times$ magnification (Figure 4.11), the NC condition (a) exhibits the coarsest grain structure with clearly visible elongated columnar grains oriented along the build direction, separated by well defined layer bands at regular intervals corresponding to the deposited layer height. The PC condition (b) shows an essentially similar columnar morphology to NC, confirming that passive heat sinking provides insufficient cooling to alter the solidification mode. Under A20 (c) and A10 (d), a transition towards a mixed columnar–equiaxed morphology is evident, with reduced layer band prominence, reflecting the increasing disruption of the directional thermal gradient by the active water flow. The TIAC (e), TIAC-BD (f), and LN₂ (g) conditions display markedly finer and more uniformly distributed grain structures relative to the Phase 1 conditions. The $200\times$ micrographs (Figure 4.12) confirm these trends at higher magnification, revealing that the inter-dendritic spacing and precipitate network become progressively finer

from NC through to LN₂.

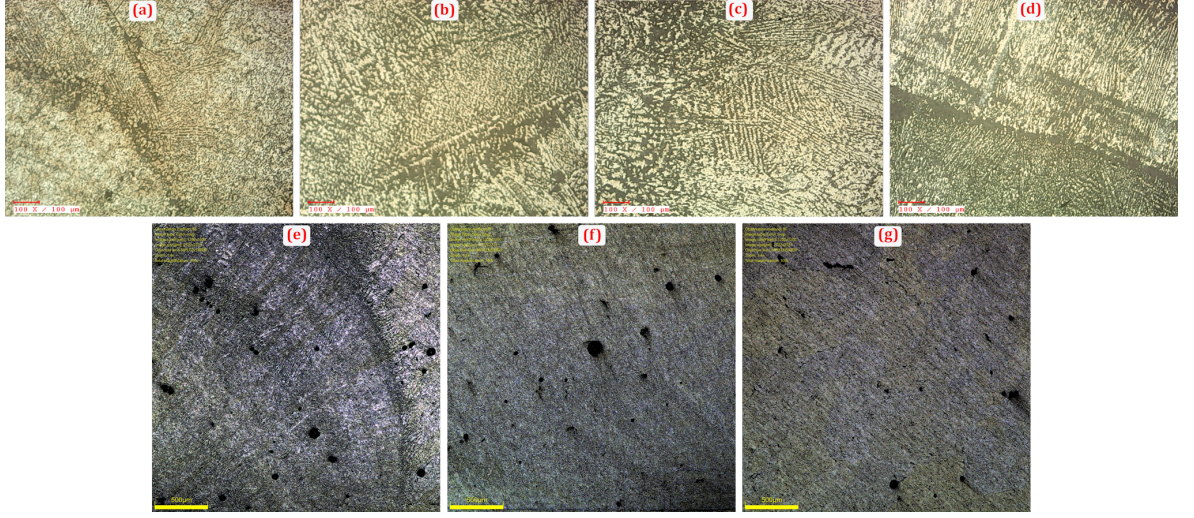


Figure 4.11: Optical micrographs at 100 $\times$ magnification of etched transverse cross sections for (a) NC, (b) PC, (c) A20, (d) A10, (e) TIAC, (f) TIAC-BD, (g) LN₂

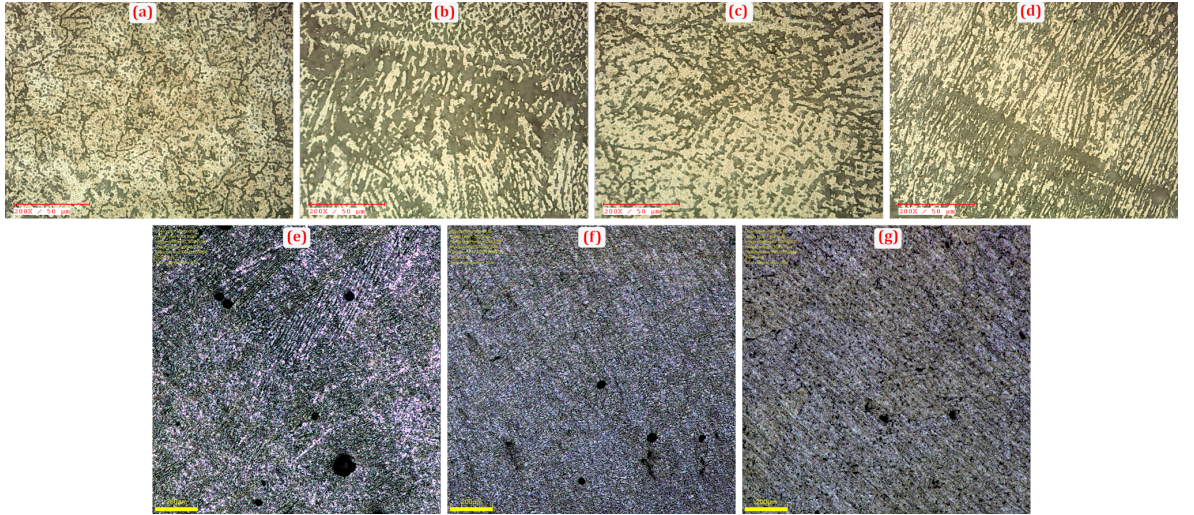


Figure 4.12: Optical micrographs at 200 $\times$ magnification of etched transverse cross sections for (a) NC, (b) PC, (c) A20, (d) A10, (e) TIAC, (f) TIAC-BD, (g) LN₂

4.5.2 Columnar-to-Equiaxed Transition

Under the NC condition, the microstructure is characterised by elongated columnar grains oriented predominantly along the build height (Z) direction, with an aspect ratio (length/width) greater than unity indicating a strongly directional solidification texture. The columnar growth mode is driven by the steep, directional thermal gradient established by the moving arc heat source combined with the slow heat extraction rate at the substrate, which promotes epitaxial solidification on the previously deposited

layer [10, 13]. Layer bands alternating regions of coarser and finer grain structure are visible at regular intervals corresponding to the deposited layer height and represent the periodic reheating and partial remelting of previously deposited layers by each successive pass. The PC condition produces a broadly similar columnar structure, confirming that passive heat sinking does not alter the solidification mode.

Progressive activation of interpass cooling increasingly disrupts the columnar growth mechanism. Under A20 and A10, the micrographs reveal a mixed columnar–equiaxed morphology, with the equiaxed fraction and grain refinement degree increasing with water flow rate. Under TIAC and TIAC-BD, the grain structure transitions to a predominantly fine equiaxed morphology and under LN₂ the finest grain size of all conditions is achieved with a near fully equiaxed structure. This columnar-to-equiaxed transition (CET) with increasing cooling intensity is consistent with the classical Hunt CET criterion [117], which predicts that a higher undercooling ahead of the solidification front promotes nucleation of new equiaxed grains ahead of the advancing columnar interface, blocking further columnar growth. The quantitative grain size values measured by the linear intercept method (ASTM E112) for all conditions and wall zones are summarised in Table 4.6.

Table 4.6: Average grain size and grain aspect ratio for all cooling conditions

Condition	Grain Size (μm)	Aspect Ratio
NC	82.1	1.52
PC	98.2	1.74
A20	110.8	1.62
A10	73.7	1.66
TIAC	51.8	1.45
TIAC-BD	48.7	1.52
LN ₂	42.5	1.52

Aspect ratio = median major/minor axis from ellipse fitting on segmented grains.

It should be noted that the Phase 1 grain size values (NC, PC, A20, A10) carry substantial measurement uncertainty owing to the low grain counts (NC: 8, PC: 10, A20: 8, A10: 9 grains), which arise because the large columnar grains in these conditions span a significant fraction of the optical field of view. With fewer than 15 complete grains measurable per micrograph, the ASTM E112 intercept values are indicative rather than statistically robust, and the apparent non-monotonic ordering within Phase 1 (NC < A10 < PC < A20) should be interpreted with this limitation in mind. The clear and statistically reliable finding is the substantial reduction in grain size between the Phase 1 range (73.7–110.8 μm) and the Phase 2 conditions (42.5–51.8 μm , based on 67 grains for TIAC, 213 for TIAC-BD and 432 for LN₂), which confirms that the more intensive cooling strategies of Phase 2 produce a significantly finer equiaxed

microstructure than the Phase 1 immersion only conditions.

4.5.3 Effect of Cooling Rate on Microstructure

The Hall–Petch relationship predicts that a reduction in grain size increases yield strength according to:

$$\sigma_y = \sigma_0 + k_y \cdot d^{-1/2} \quad (4.5)$$

where σ_y is the yield strength, σ_0 is the lattice friction stress, k_y is the Hall–Petch coefficient, and d is the average grain diameter. Across all seven conditions, the vertical direction yield strength (TV YS) shows a positive correlation with $d^{-1/2}$ (Pearson $r = 0.80$), which is broadly consistent with grain boundary strengthening playing a contributory role. However, this correlation must be interpreted with caution: the most significant step change in both grain size and YS occurs between Phase 1 and Phase 2, which also differ in heat input ($Q = 0.174$ vs 0.508 kJ/mm) and consequently in Mg solid solution retention (Section 4.5). Within Phase 2 alone, the three conditions show no consistent Hall–Petch trend, indicating that grain refinement cannot be isolated as the sole strengthening mechanism. Grain boundary strengthening is therefore identified as one of at least two concurrent mechanisms the other being solid solution strengthening through retained Mg both of which are promoted by more intensive interpass cooling, as discussed further in Section 4.6.

Additionally, faster cooling rates suppress the coarsening of inter-dendritic Mg rich precipitates (principally $\beta\text{-Al}_3\text{Mg}_2$) that occurs when the deposit is held at elevated temperatures for prolonged periods. Under NC, the extended time above 200°C promotes diffusion controlled growth of these precipitates, which reduces the solid solution Mg content of the Al matrix and thereby diminishes the solid solution strengthening contribution. Active cooling conditions minimise this thermal exposure, retaining a higher Mg concentration in solid solution and contributing to the observed improvement in both yield and tensile strength.

4.5.4 SEM and EDS Analysis

SEM and EDS characterisation was performed on five conditions (NC, A10, TIAC, TIAC-BD and LN₂). The same selection rationale applies as for XRD: NC is the baseline, A10 is the best performing Phase 1 condition and TIAC, TIAC-BD and LN₂ represent the novel cooling strategies introduced in this thesis. SEM images of the cross sections for all five conditions at both the bottom and top zones are presented in Figure 4.13. BSE images for the NC and A10 conditions, showing the inter-dendritic precipitate distribution, are shown in Figure 4.14. EDS analyses were performed on all five conditions.

EDS point analyses were performed at two sites per specimen (matrix and precipitate regions) and the averaged results are summarised in Table 4.7. All specimens show the Al–Mg matrix composition expected for ER5356 alloy, with Mg content in the range 3.0–5.2 at.%. The elevated oxygen content recorded for NC specimens is consistent with the presence of a natural surface oxide layer. A10 specimens yielding O values below 2.2 at.%, consistent with the TIAC, TIAC-BD and LN₂ specimens. The updated A10 values (Table 4.7) represent matrix Mg content with minimal oxide layer influence.

The SEM micrographs in Figure 4.13 confirm the grain morphology trends established by optical microscopy. Under NC, the cross section reveals a coarse, predominantly columnar grain structure elongated in the build direction, with well defined layer bands. The A10 condition shows a partially refined grain structure with a mixed columnar–equiaxed morphology, reflecting the intermediate cooling intensity of the 10 L/min water flow. The TIAC, TIAC-BD and LN₂ conditions all display substantially finer, more equiaxed grains relative to NC, consistent with the grain size measurements in Section 4.5.1. The BSE images in Figure 4.14 show bright precipitate particles distributed along inter-dendritic boundaries for both NC and A10, with the NC specimens exhibiting a higher area density of visible precipitate features, particularly in the top zone where the longer thermal exposure promotes precipitate coarsening. The EDS Mg content (wt.%) for NC BOT (4.15%) and NC TOP (4.19%) falls below the ER5356 nominal Mg range of 4.5–5.5%, indicating partial Mg precipitation out of the Al matrix into the inter-dendritic β -phase. Under TIAC, the matrix Mg content (4.27–4.33 wt.%) approaches the nominal wire composition more closely, consistent with the lattice parameter analysis in Section 4.6 and confirming that faster interpass cooling suppresses Mg precipitation and retains a higher Mg solid solution content.

Table 4.7: EDS analysis results for all conditions at bottom and top zones

Condition	Zone	Al (at.%)	Mg (at.%)	O (at.%)	Al (wt.%)	Mg (wt.%)
NC	TOP	86.55	4.46	9.00	90.25	4.19
NC	BOT	78.81	4.27	16.93	85.02	4.15
A10	TOP	92.44	5.36	2.20	91.90	4.80
A10	BOT	93.43	4.98	1.58	89.50	4.30
TIAC	TOP	81.82	4.37	2.50	88.63	4.27
TIAC	BOT	77.09	4.33	1.32*	86.38	4.33
TIAC-BD	TOP	55.71	2.89	2.49	70.57 [†]	3.14 [†]
TIAC-BD	BOT	64.33	2.93	1.18*	78.79 [†]	2.68 [†]
LN ₂	TOP	68.35	3.76	2.16	80.21	3.98
LN ₂	BOT	74.37	4.01	0.88	84.74	4.13

* O not detected at one site; average of detected site only.

[†] TIAC-BD sites include one measurement on a precipitate particle; matrix values likely higher.

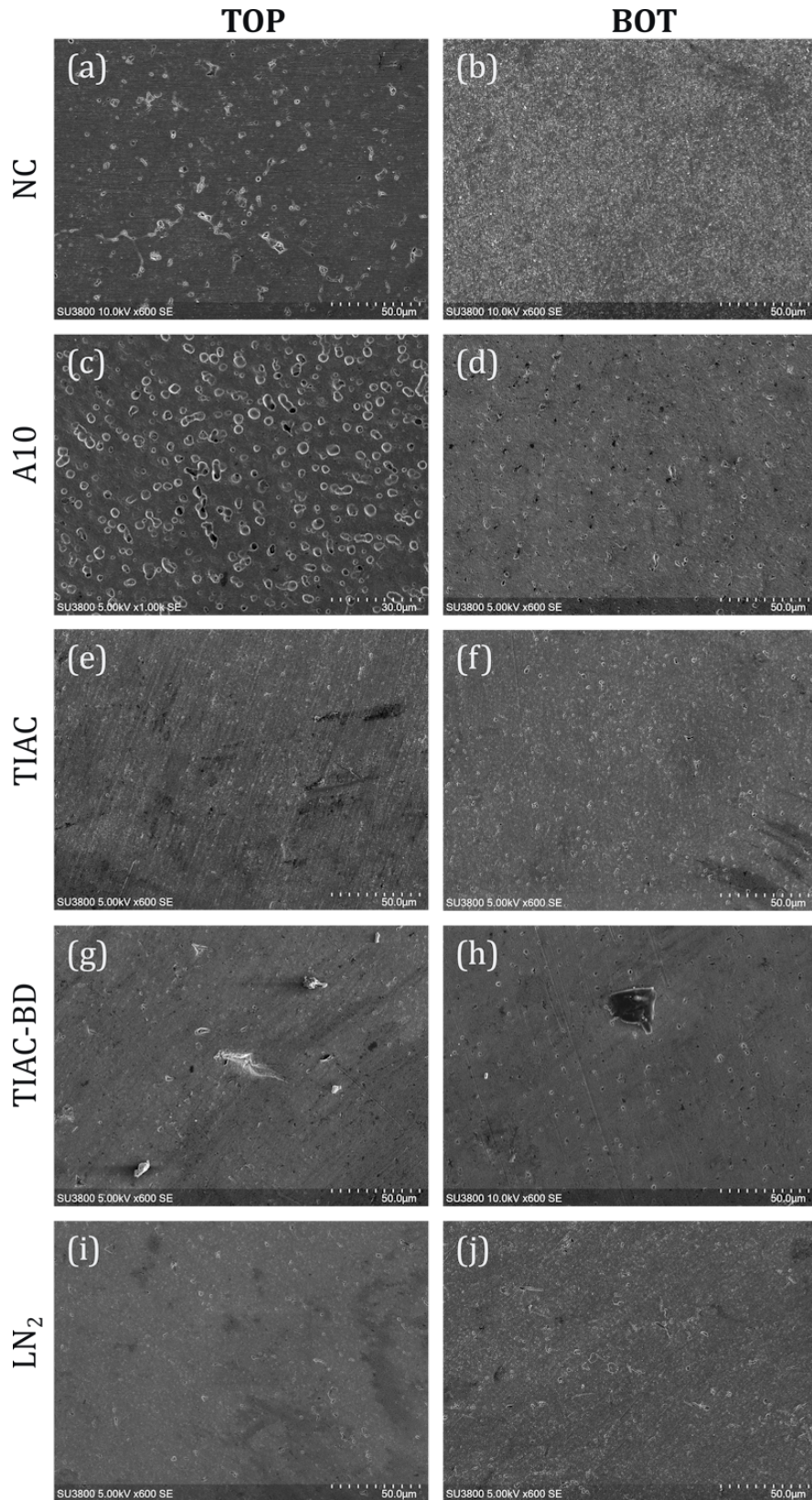


Figure 4.13: SEM micrographs of cross sections for five interpass cooling conditions at top (left column) and bottom (right column) zones:
(a,b) NC, (c,d) A10, (e,f) TIAC, (g,h) TIAC-BD, (i,j) LN₂

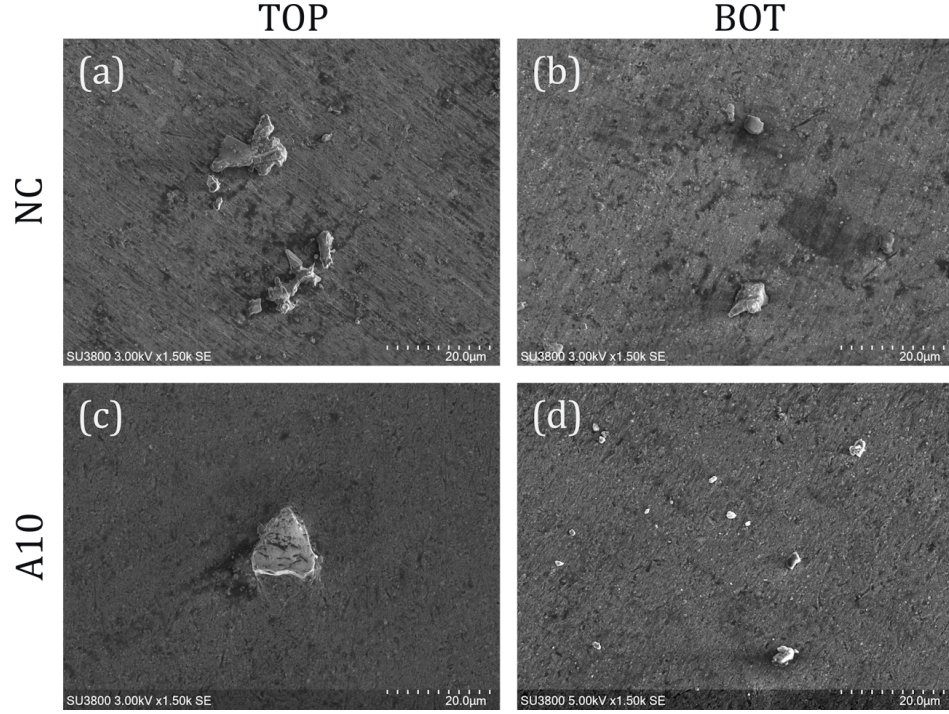


Figure 4.14: SEM-BSE images showing inter-dendritic precipitate distribution for (a) NC BOT, (b) NC TOP, (c) A10 BOT, (d) A10 TOP

4.5.5 Effect of Deposition Strategy on Microstructure

Optical micrographs of the TIAC-BD deposit at $100\times$ and $200\times$ magnification are presented alongside their TIAC counterparts in Figure 4.11(f) and Figure 4.12(f). The micrographs confirm that the bulk grain morphology of TIAC-BD is broadly equivalent to that of TIAC, both conditions produce a fine, predominantly equiaxed grain structure, consistent with the same interpass cooling intensity governing the local solidification conditions regardless of deposition direction. The most notable microstructural distinction between the two conditions is the spatial uniformity of the grain size distribution along the deposition direction. Under unidirectional TIAC deposition, a slight gradient in grain coarseness is discernible from the start end to the finish end of the wall, reflecting the progressive heat accumulation in the substrate and substrate fixture over successive unidirectional passes. This intra-wall gradient is reduced under TIAC-BD, where alternating deposition directions redistribute the thermal burden more symmetrically along the wall length, leading to a more homogeneous microstructure from one end to the other.

The SEM micrographs in Figure 4.13 panels (d,i) and (c,h) (TIAC-BD and TIAC respectively) further confirm equivalent grain refinement between the two strategies at the $20\text{ }\mu\text{m}$ scale. Inter-layer oxide band continuity was also visually comparable between the two conditions, since both experience the same interpass thermal window and the same arc cleaning action at the deposit surface.

TIAC-BD and TIAC produce equivalent bulk grain refinement, precipitate distribution and porosity. The advantage of bidirectional deposition is spatial homogeneity along the wall length this reduces property scatter from start-end thermal asymmetry, which matters for component certification.

4.6 Phase Analysis (XRD)

XRD characterisation was performed on five of the seven conditions: NC, A10, TIAC, TIAC-BD and LN₂. NC was included as the unmodified baseline representing standard manufacturing practice. A10 was selected as the best performing Phase 1 active cooling condition. PC and A20 were excluded because PC produced negligible microstructural change relative to NC (confirmed by optical microscopy), and A20 was superseded by A10 within Phase 1. TIAC, TIAC-BD and LN₂ are the novel contributions of this thesis and form the primary focus of the advanced characterisation.

4.6.1 Phase Identification

X-ray diffraction patterns for the NC, A10, TIAC, TIAC-BD and LN₂ conditions at both the bottom and top zones of the deposited walls are shown in Figure 4.16 and Figure 4.15, respectively. All patterns were acquired on a Bruker D6 PHASER diffractometer using Cu K α_1 radiation ($\lambda = 1.54060 \text{ \AA}$), with a coupled $2\theta/\theta$ scan from 20° to 90° at a step size of 0.029° and 0.4 s per step.

In all conditions, the XRD patterns are dominated by the strong reflections of the face centred cubic (FCC) aluminium matrix at 2θ positions corresponding to the (111), (200), (220), (311), and (222) planes (Cu K α_1). The five principal Al reflections appear at approximately 38.2° , 44.4° , 64.7° , 77.8° , and 82.4° respectively, consistent with the ER5356 Al–Mg alloy matrix. No evidence of aluminium oxide (Al₂O₃) or aluminium nitride (AlN) peaks was detected in any condition, confirming the effectiveness of the argon shielding gas in protecting the molten pool from atmospheric contamination. No distinct β -Al₃Mg₂ secondary phase peaks were resolved in the diffraction patterns of the conditions measured, which is consistent with the fine, dispersed distribution of inter-dendritic Mg rich precipitates at the sub-micrometre scale observed by SEM rather than a coarse, crystallographically ordered phase detectable by XRD.

4.6.2 Effect of Cooling Techniques on Lattice Parameter

The lattice parameter a of the FCC aluminium matrix was calculated from the position of the (311) reflection using Bragg's law, as this high angle reflection provides the most

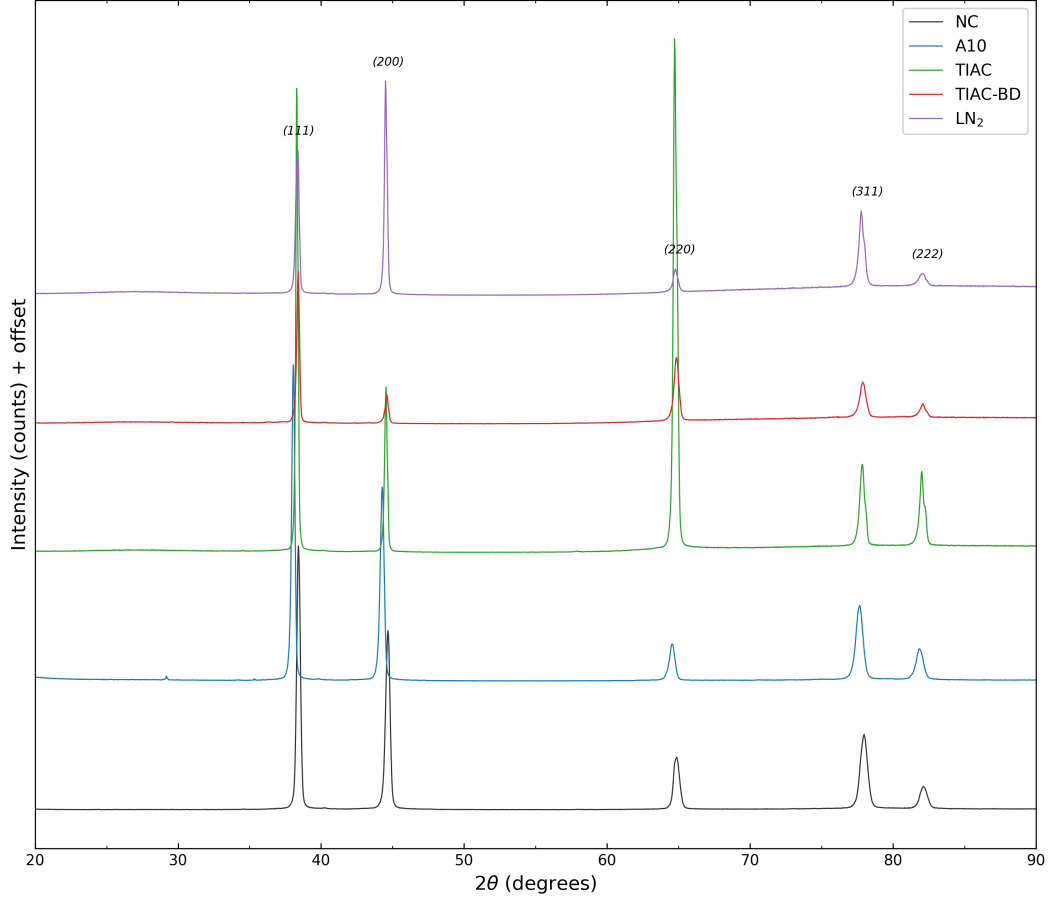


Figure 4.15: Stacked XRD patterns for five cooling conditions at the TOP zone of the deposited walls

accurate d -spacing measurement:

$$a = d_{hkl} \sqrt{h^2 + k^2 + l^2} \quad (4.6)$$

The measured lattice parameters (tabulated in Table 4.8) range from $a = 4.0639 \text{ \AA}$ to $a = 4.0755 \text{ \AA}$ across all conditions and zones, all exceeding the ICDD reference value for pure Al ($a_0 = 4.0494 \text{ \AA}$, PDF 04-0787). This consistent lattice expansion is attributed to the substitutional solid solution of Mg in the Al matrix: the larger Mg atom ($r_{\text{Mg}} = 1.60 \text{ \AA}$ vs. $r_{\text{Al}} = 1.43 \text{ \AA}$) expands the unit cell in proportion to the retained Mg content. The A10 condition exhibits the highest lattice parameter ($a = 4.0755 \text{ \AA}$ at the bottom zone), indicating the greatest degree of Mg solid solution retention among the characterised conditions, consistent with the rapid Phase 1 interpass cooling at LEWD = 10 mm suppressing diffusion controlled Mg precipitation between passes. The NC condition shows a lattice parameter of $a = 4.0678 \text{ \AA}$ at the bottom zone, consistent with partial Mg precipitation into inter-dendritic β -phase reducing the matrix Mg content under slow cooling. This ranking is confirmed by the EDS data in Table 4.7: the etched A10 specimens yield matrix Mg values of 4.3 wt.%

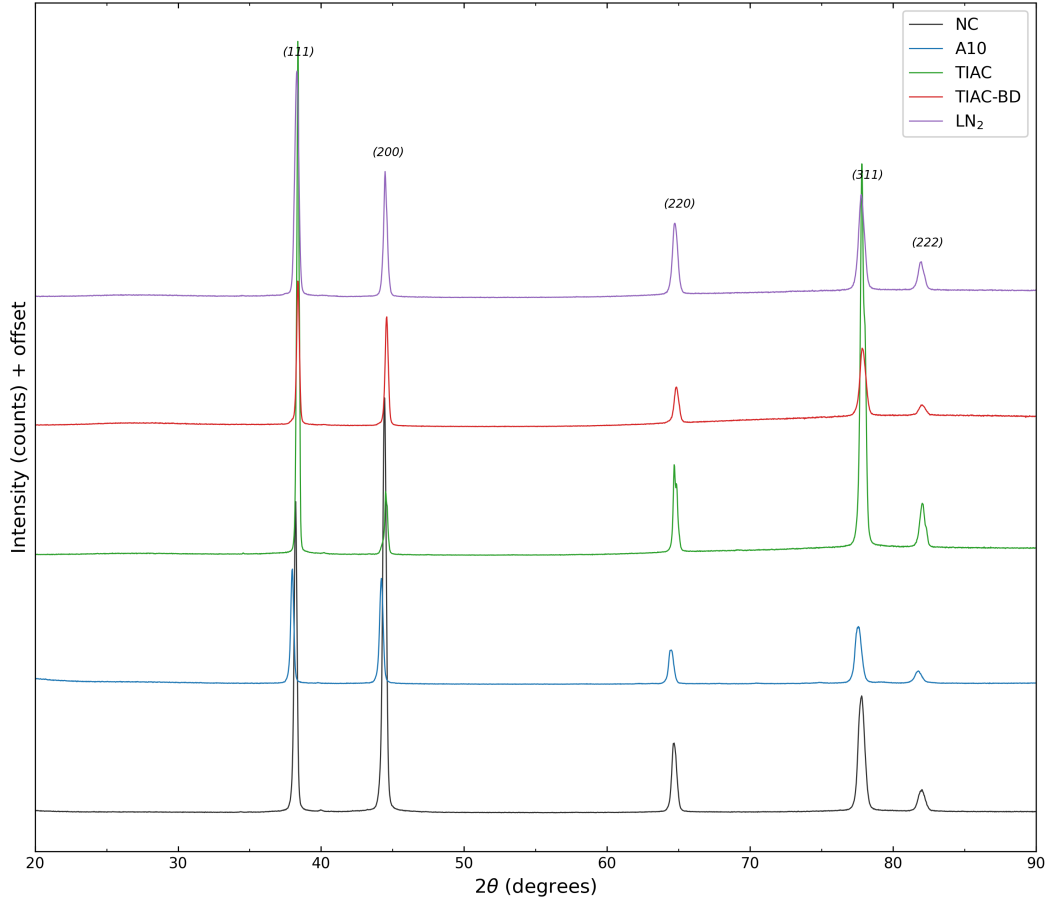


Figure 4.16: Stacked XRD patterns for five cooling conditions at the BOT zone of the deposited walls)

(BOT) and 4.8 wt.% (TOP), both exceeding the NC values (4.15 and 4.19 wt.% respectively) and approaching the nominal ER5356 wire composition, consistent with the higher lattice parameter indicating superior solid solution Mg retention under aggressive interpass cooling. More intensive interpass cooling therefore promotes greater Mg solid-solution retention, which contributes to the higher yield strengths in the actively cooled conditions.

4.6.3 Crystallographic Texture

The integrated intensities of the five principal FCC Al reflections (111), (200), (220), (311), and (222) were extracted from each diffraction pattern and compared against the ICDD reference powder intensities (PDF 04-0787) to assess preferred crystallographic orientation using the Harris Texture Coefficient [118]:

$$TC(hkl) = \frac{I(hkl)/I_0(hkl)}{\frac{1}{N} \sum_{i=1}^N I(hkl_i)/I_0(hkl_i)} \quad (4.7)$$

where $I(hkl)$ is the measured integrated intensity, $I_0(hkl)$ is the ICDD reference intensity for the same reflection, and $N = 5$ is the number of reflections used. A value of $TC(hkl) = 1$ indicates a random (powder like) distribution; $TC(hkl) > 1$ indicates preferred orientation in the $\langle hkl \rangle$ direction. The $TC(200)$ value is used here as the primary texture indicator because $\langle 100 \rangle$ preferred orientation is the signature texture of competitive columnar growth in FCC metals deposited by WAAM [3, 13].

Under NC, the bottom zone specimen exhibits $TC(200) = 1.51$, indicating a strong $\langle 100 \rangle$ fibre texture aligned with the build direction, characteristic of directional solidification under the steep thermal gradient of the moving arc. The NC top zone shows $TC(200) = 0.99$, approaching a random distribution, consistent with the progressive reduction in thermal gradient severity in the upper layers as the deposit height increases. The TIAC condition shows a markedly reduced $TC(200)$ of 0.16 at the bottom zone, indicating that the dual heat extraction pathway disrupts directional heat flow sufficiently to suppress the $\langle 100 \rangle$ growth preference and promote a near random crystallographic orientation, consistent with the fine equiaxed grain morphology observed in the optical and SEM micrographs (Section 4.5.1). The TIAC-BD and A10 conditions show intermediate $TC(200)$ values (0.82 and 0.89 at the bottom zone respectively), reflecting a partial reduction in texture severity relative to NC. LN_2 exhibits $TC(200) = 0.59$ at the bottom zone, rising to 1.26 at the top zone. This bottom-to-top gradient is attributed to two concurrent effects. First, the cryogen is applied to the top surface of each previously deposited layer immediately before the next deposition pass, such that the uppermost layers solidify against a cryogenically cooled surface below; the resulting steep thermal gradient in the build direction promotes strongly directional $\langle 100 \rangle$ solidification in the top zone, consistent with classical competitive growth theory [13]. Second, the top zone layers are not subsequently reheated by any overlying passes, so this steep gradient solidification texture is preserved intact. In contrast, the substrate level TC-A data (Section 4.1) shows the LN_2 substrate accumulating the highest thermal burden of all conditions ($HAI = 170.3^\circ C$), indicating that the bottom zone layers solidify against a progressively warming substrate that reduces the effective thermal gradient at the deposit base, thereby weakening the driving force for $\langle 100 \rangle$ columnar growth and producing the lower $TC(200) = 0.59$ observed at the bottom zone.

4.7 Fractographic Analysis

Fractographic analysis was performed on the same five conditions characterised by XRD and SEM/EDS (NC, A10, TIAC, TIAC-BD and LN_2), for the same reasons of scope: NC provides the baseline, A10 represents the best Phase 1 result and TIAC, TIAC-BD and LN_2 are the novel contributions of this thesis. PC and A20 are fully characterised

Table 4.8: XRD derived crystallographic parameters at bottom and top zones for five characterised conditions (NC, A10, TIAC, TIAC-BD, LN₂)

Condition	Lattice Param. (Å)		Crystallite Size (nm)		TC(200)	
	Bot.	Top	Bot.	Top	Bot.	Top
NC	4.0678	4.0601	42.9	33.4	1.51	0.99
A10	4.0755	4.0729	42.9	37.6	0.89	0.96
TIAC	4.0678	4.0652	60.1	50.1	0.16	0.28
TIAC-BD	4.0665	4.0639	43.0	50.1	0.82	0.27
LN ₂	4.0691	4.0691	27.3	33.4	0.59	1.26

TC(200) > 1: preferred $\langle 100 \rangle$ orientation; TC(200) = 1: random; TC(200) < 1: depleted $\langle 100 \rangle$

by the thermal, geometric, mechanical and optical microstructural analyses presented in the preceding sections.

4.7.1 Tensile Fracture Behaviour

SEM fractographs of the tensile fracture surfaces at 50 μm scale for vertical (TV) and horizontal (TB) specimens across the four interpass cooling conditions are presented in Figure 4.17.

All four cooling conditions produce a predominantly ductile fracture morphology in the tensile specimens, characterised by equiaxed microvoid coalescence dimples distributed uniformly across the fracture surface. No flat inter-granular facets or cleavage like river marks indicative of brittle fracture were observed under any condition, consistent with the relatively high elongation values recorded across all active cooling strategies. The A10 condition (panels a, e) exhibits a dimpled fracture surface with moderate dimple size and depth, indicative of ductile failure through microvoid nucleation at inter-dendritic precipitate particles followed by void growth and coalescence. Under TIAC (panels b, f), the dimple population is finer and more uniformly distributed, consistent with the refined equiaxed grain structure and more homogeneously dispersed precipitate phase observed in the microstructural analysis. The finer dimple size in TIAC reflects a greater density of void nucleation sites associated with the more refined precipitate distribution, each nucleating a smaller void, which collectively absorbs more energy per unit volume and contributes to the higher ductility and UTS measured for this condition. The TIAC-BD condition (panels c, g) shows a dimple morphology broadly similar to TIAC, with no evidence of inter-granular cracking or delamination at layer-band interfaces. Under LN₂ cooling (panels d, h), the fracture surface morphology remains dimple dominated but with some variation in dimple size across the surface, which may reflect localised microstructural heterogeneity introduced by the high transient thermal gradient imposed by cryogenic application.

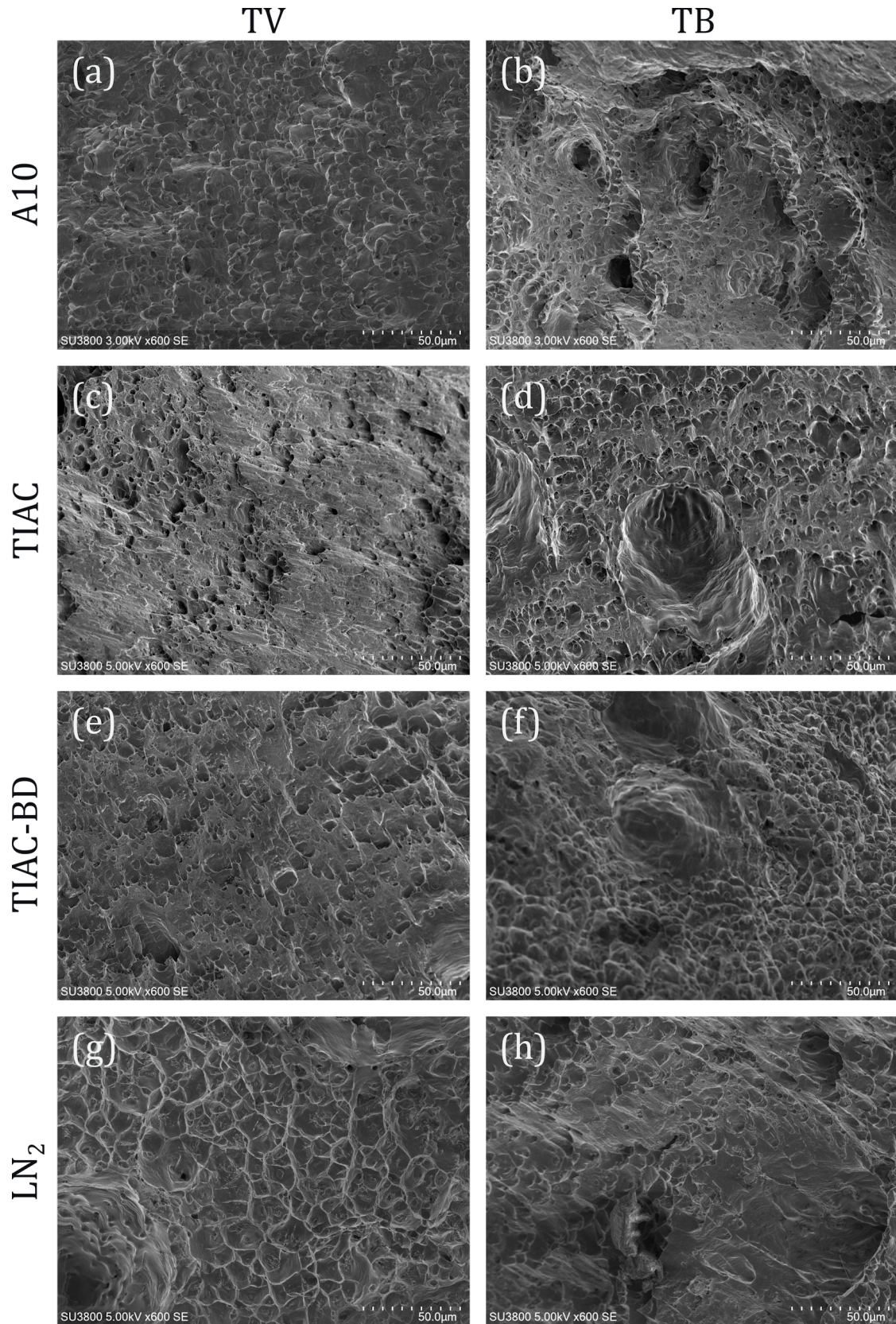


Figure 4.17: SEM fractographs of tensile fracture surfaces for vertical (TV) and horizontal (TB) specimens

The horizontal (TB) fracture surfaces for all conditions show a higher proportion of

deep, equiaxed dimples relative to the corresponding vertical (TV) surfaces, consistent with the generally higher elongation values measured in the horizontal tensile direction. In the vertical direction, occasional shallow dimples and mildly elongated features are visible at inter-layer boundaries, confirming that inter-layer interfaces remain the microstructural weak link in the build direction even under active cooling conditions.

4.7.2 Impact Fracture Behaviour

SEM fractographs of Charpy V-notch impact fracture surfaces at 50 μm scale for horizontal (CH) and vertical (CV) specimens are shown in Figure 4.18. The impact fractographs reveal a fracture morphology distinct from the tensile surfaces, reflecting the high strain rate nature of impact loading.

All impact fracture surfaces exhibit a highly interconnected dimple network with a tortuous, multi level topography characteristic of ductile fracture under high strain rate loading. The interconnected morphology differs markedly from the isolated, discrete dimples observed on the tensile surfaces and is consistent with the rapid crack propagation and multi plane void coalescence that occurs under impact conditions. Under A10 (panels a, e), the impact fracture surface shows a well developed dimple network with moderate feature size. The TIAC condition (panels b, f) produces a similarly ductile morphology with a finer and more uniform dimple network, reflecting the refined equiaxed microstructure and consistent with the higher Charpy impact energy measured for this condition. TIAC-BD (panels c, g) displays a comparable impact fracture morphology to TIAC, with no evidence of delamination or inter-granular fracture planes that would indicate layer boundary weakness. Under LN_2 (panels d, h), the impact fracture surface is predominantly dimpled but shows a slightly coarser network texture compared to TIAC, which may be associated with localised residual stress concentrations at grain boundaries introduced by the extreme transient thermal gradient of cryogenic cooling, leading to slight crack deflection and larger void growth zones. The vertical (CV) impact fracture surfaces generally show greater topographic relief than the horizontal (CH) surfaces across all conditions, consistent with the anisotropic grain structure and the greater resistance to crack propagation perpendicular to the build direction.

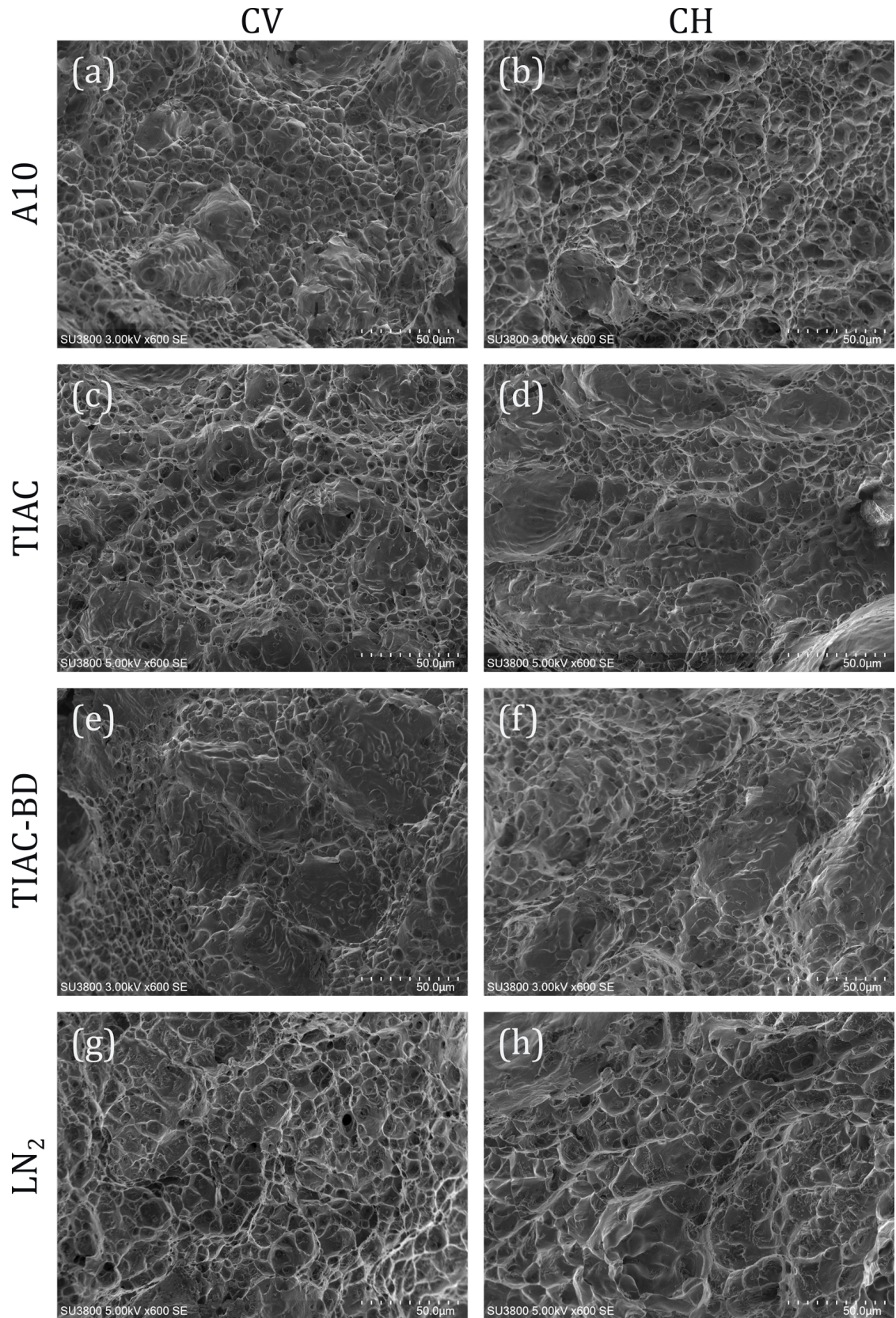


Figure 4.18: SEM fractographs of Charpy V-notch impact fracture surfaces for horizontal (CH) and vertical (CV) specimens

4.7.3 Failure Mechanisms

Based on the combined evidence from the tensile and impact fractographic analysis, the dominant failure mechanisms for the four characterised conditions are as follows.

A10 fails by ductile microvoid coalescence in both tensile and impact loading, with void nucleation occurring at inter-dendritic precipitate particles. The moderate dimple size reflects a precipitation distribution that is finer than NC but coarser than TIAC, consistent with the intermediate interpass cooling intensity of the 10 L/min water flow.

TIAC shows the most refined and homogeneous dimple morphology across all orientations and loading modes. No brittle fracture features were observed in either tensile or impact specimens, consistent with TIAC's superior combination of strength and ductility.

TIAC-BD shows a fracture morphology closely comparable to TIAC, confirming that the bi-directional deposition strategy does not introduce additional fracture initiation sites or anisotropy relative to unidirectional TIAC. No delamination at layer-band interfaces was observed, indicating effective inter-layer bonding under both deposition strategies.

LN₂ produces predominantly ductile fracture surfaces but with evidence of localised variation in dimple size and occasional surface roughness features not present under TIAC, consistent with the non-uniform thermal gradient imposed by cryogenic surface application. This microstructural heterogeneity is the likely cause of the marginally lower impact energy of LN₂ relative to TIAC despite its finer average grain size.

4.8 Comparative Analysis of Cooling Techniques

4.8.1 Two-Phase Experimental Design and Scope of Comparison

As established in Chapter 3, the seven conditions span two experimental phases with different process parameters and heat inputs: Phase 1 (NC, PC, A20, A10; $Q = 0.174$ kJ/mm) and Phase 2 (TIAC, TIAC-BD, LN₂; $Q = 0.508$ kJ/mm). The reduced travel speed in Phase 2 (4.0 mm/s) was necessitated by the TIAC jet geometry and arc stability constraints near the water surface, as discussed in Section 3.5. Within each phase, all comparisons are fully controlled for process parameters and any differences in measured responses are attributable solely to the interpass cooling strategy. Cross phase comparisons must be interpreted with awareness that Phase 2 conditions impose a $2.9\times$ higher heat input per unit length, a substantially more demanding thermal challenge. The demonstration that Phase 2 cooling strategies (TIAC, TIAC-BD) produce

competitive or superior deposit quality despite this elevated heat input constitutes a conservative and robust validation of their effectiveness: TIAC is not evaluated under favourable conditions but under conditions that are inherently more difficult to manage thermally.

Within each phase the cooling condition was the sole deliberately varied factor, and the following were held fixed so that the observed differences may be attributed to it. The process parameters, comprising current, voltage, travel speed, wire feed speed, torch angle and contact tip to work distance, were fixed at the values given in Section 3.5 and were not adjusted between conditions. The equipment was unchanged throughout, the same power source, robot, torch and wire feed unit being used for every build, and the feedstock was drawn from a single wire spool lot so that no batch to batch compositional variation could enter. The shielding gas composition, purity and flow rate were fixed and verified at the flow meter before each build. The substrate material, dimensions, surface preparation and clamping arrangement were identical for every build, and the substrate was replaced between builds so that no accumulated thermal history carried across. The deposition path, the layer count and the interpass dwell of 10 s were the same for every condition, the dwell in particular being held uniform rather than tuned to a target interpass temperature, so that the cooling condition and not the dwell determined the thermal state. Specimens were extracted from the same relative locations in every wall and tested on the same machines to the same standards.

Two factors could not be eliminated and are stated openly. The first is the heat input difference between the two phases described above, which is why no cross phase claim is made except in the conservative direction. The second is that the layer edge to water distance was maintained by periodic replenishment between passes rather than by continuous level control, so it varied within a small band about its target value rather than being held exactly. Both are carried forward into the statement of limitations in Chapter 5.

4.8.2 Natural versus Passive versus Active Cooling

A three tier comparison of NC, PC and A10 highlights the magnitude of improvement achievable at each escalation of cooling intensity. Moving from NC to PC reduced the HAI by 52.4% and surface waviness by 31.3% (absolute values in Sections 4.1.2 and 4.2.4), while the change in TV UTS was negligible at +1.2%, confirming that passive heat sinking improves thermal management but is insufficient to translate this into meaningful mechanical property gains. The transition from PC to A10 represents a step change improvement in thermal management driven by the far superior thermal conductivity of water: the HAI dropped a further 54.5% and surface waviness by a further 34.5%, giving an overall 55.0% waviness reduction relative to NC. The A10 condition

also produced the most isotropic tensile response in Phase 1 ($AR_t = 1.17$), and the highest deposition efficiency ($DE = 0.86$). The TV UTS under A10 (228.1 MPa) was lower than NC (264.3 MPa). This is attributed to a finer but less work hardened microstructure and reduced elongation (12.2% vs 21.8% for NC). Active cooling improves geometry and thermal stability, but does not universally raise all tensile metrics.

4.8.3 Active Cooling (A20/A10) versus TIAC versus LN₂

Within Phase 2, TIAC consistently produced the best combination of low HAI, competitive mechanical properties, fine microstructure and low anisotropy compared to TIAC-BD and LN₂ under the same $Q = 0.508$ kJ/mm. The dual heat extraction pathway of TIAC, substrate immersion combined with inter-layer jet impingement, gives the lowest substrate level HAI of the three after TIAC-BD and a value $3.1\times$ lower than LN₂ (Section 4.1.2). The higher substrate level HAI of LN₂ reflects the fact that cryogen is applied to the deposit surface only; TC-A (drilled into the substrate) sees no direct cooling benefit, and the substrate heats progressively in a manner similar to natural convection. While LN₂ achieves the finest deposit surface cooling and grain refinement, it introduces residual stresses from the extreme transient thermal gradient and requires a more complex delivery infrastructure, higher operating costs and dedicated safety protocols, reducing its practical attractiveness for production environments. A cross phase qualitative observation is that TIAC achieves $HAI = 55.0$ °C while managing a heat input $2.9\times$ higher than the entire Phase 1 range, confirming that the tandem cooling architecture is capable of handling significantly more demanding deposition conditions than those investigated in Phase 1.

4.8.4 Performance Evaluation Matrix

Table 4.10 presents a multi criteria performance evaluation of all seven cooling strategies, scored on a scale of 1 (worst) to 5 (best) using min–max normalisation of measured data where available. Table 4.9 documents the mapping from each raw measured parameter to its parent criterion, the weight assigned to each parameter within that criterion and the normalisation approach. For criteria influenced by the deposition heat input (geometric quality and build productivity), normalisation is applied within each phase separately to avoid penalising Phase 2 conditions for their inherently higher Q ; for all other criteria, normalisation is applied across all seven conditions. Implementation ease and operating cost are scored by expert assessment of hardware complexity and consumable cost respectively. All eight criteria carry equal weight in the total score. The same scores are visualised as a radar chart in Figure 4.19 to facilitate visual comparison of the performance profile of each condition.

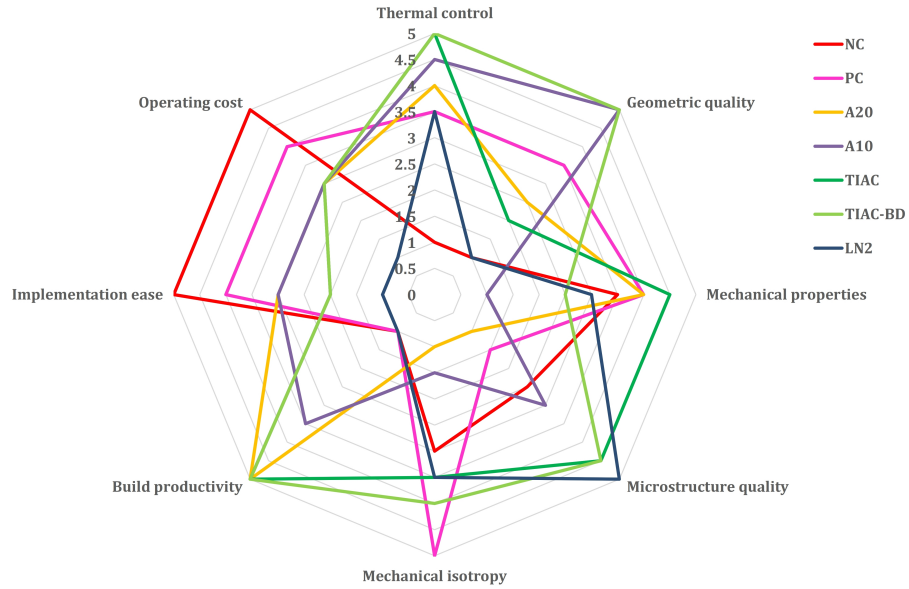
Table 4.9: Mapping of raw measured parameters to evaluation criteria, parameter weight within each criterion and normalisation method

Crit.	Criterion	Parameter(s) used	Normalisation	Direction	Weight
1	Thermal control	HAI/ Q ($\text{rC}\cdot\text{mm}/\text{kJ}$)	Cross phase min–max	Lower	100%
2	Geometric quality	Surface waviness (SW, mm)	Within phase min–max	Lower	50%
		Deposition efficiency (DE, ratio)	Within phase min–max	Higher	50%
3	Mechanical properties	TV Ultimate tensile strength (MPa)	Cross phase min–max	Higher	25%
		TV Yield strength (MPa)	Cross phase min–max	Higher	25%
		TV Elongation (%)	Cross phase min–max	Higher	25%
		CV Charpy impact energy (J/cm^2)	Cross phase min–max	Higher	25%
4	Microstructure quality	Grain size (μm , ASTM E112 linear intercept)	Cross phase min–max	Lower	100%
5	Mechanical isotropy	Tensile anisotropy index AI_t ($\text{UTS}_{\text{max}}/\text{UTS}_{\text{min}}$)	Cross phase min–max	Lower	100%
6	Build productivity	Total build time (s)	Within phase min–max	Lower	100%
7	Implementation ease	Expert assessment (hardware complexity, 1–5)	Direct	Higher	100%
8	Operating cost	Expert assessment (consumable cost, 1–5)	Direct	Higher	100%

HAI/ Q normalises the Heat Accumulation Index by heat input Q to enable fair cross-phase thermal efficiency comparison.

Table 4.10: Data driven multi criteria performance evaluation of all seven interpass cooling strategies (1 = worst, 5 = best)

Criterion	NC	PC	A20	A10	TIAC	TIAC-BD	LN ₂
Thermal control	1.0	3.5	4.0	4.5	5.0	5.0	3.5
Geometric quality	1.0	3.5	2.5	5.0	2.0	5.0	1.0
Mechanical properties	3.5	4.0	4.0	1.0	4.5	2.5	3.0
Microstructure quality	2.5	1.5	1.0	3.0	4.5	4.5	5.0
Mechanical isotropy	3.0	5.0	1.0	1.5	3.5	4.0	3.5
Build productivity	1.0	1.0	5.0	3.5	5.0	5.0	1.0
Implementation ease	5.0	4.0	3.0	3.0	2.0	2.0	1.0
Operating cost	5.0	4.0	3.0	3.0	3.0	3.0	1.0
Total score	22.0	26.5	23.5	24.5	29.5	31.0	19.0

**Figure 4.19: Radar chart of the multi criteria performance evaluation for all seven interpass cooling strategies**

4.8.5 Selection of Optimal Cooling Technique

Based on the data driven multi criteria analysis in Table 4.10 and Figure 4.19, TIAC-BD achieves the highest total score (31.0) followed closely by TIAC (29.5). TIAC-BD leads on geometric quality, thermal control and build productivity (within Phase 2) owing to its bidirectional symmetry and the marginally lower HAI (48.1 °C versus 55.0 °C for TIAC), while TIAC leads on mechanical properties. For applications where deposit geometry and spatial homogeneity are the primary requirements, TIAC-BD is the preferred choice. For applications where mechanical performance is the primary requirement, TIAC is preferred. Considering both phases together, TIAC represents the optimal interpass cooling strategy for general purpose WAAM of ER5356, delivering

the best balance of mechanical performance, thermal control and microstructural quality while maintaining competitive build productivity and practical implementation cost comparable to A20/A10 active cooling. The novel tandem architecture developed in this research combining substrate immersion with a synchronised inter-layer impingement jet thus represents a practically viable and technically superior alternative to existing active cooling approaches reported in the literature. In particular, NIAC only strategies reported by da Silva et al. [83,84] and Scotti et al. [53] achieve interpass temperature reductions of 60 to 85% relative to natural cooling, but are limited by their inability to localise cooling to the high gradient region immediately behind the arc (as identified in Chapter 2, Section 2.5.5). TIAC addresses this limitation directly through the addition of the inter-layer impingement jet, achieving a 62.3% HAI reduction relative to NC (145.9 to 55.0 °C) under a heat input $2.9\times$ higher than the Phase 1 NIAC comparable conditions, confirming that the dual heat extraction pathway extends the capability of substrate immersion cooling beyond what immersion alone can achieve. These findings are consistent with and extend the results previously reported by the present author, where the influence of active cooling on microstructure and mechanical properties was first demonstrated for WAAM-ER5356.

To place these results against the published state of the art, Table 4.11 compares the present conditions with the principal reported WAAM cooling strategies on the three criteria of energy efficiency, productivity and cost effectiveness.

Energy efficiency. The meaningful basis of comparison is the heat removed per unit of energy delivered to the deposit rather than the raw interpass temperature reduction, since a strategy that reports a large reduction at a low heat input is not thereby extracting heat efficiently. The reported near immersion studies achieve their 60 to 85% reduction at 0.08 to 0.10 kJ/mm [53,83,84,119–121], but they do so using low heat input CMT deposition. TIAC achieves a comparable 62.3% reduction while operating at 0.508 kJ/mm, that is at roughly five to six times the immersion heat input and $2.9\times$ that of the Phase 1 conditions, applied to conventional GMAW deposition. The tandem architecture therefore removes a substantially larger absolute thermal load per unit length while holding the same interpass reduction, which is the sense in which it extends the immersion result to the higher heat input regime rather than reproducing it at low heat input. It is more capable on this measure than the two remaining reported routes: forced gas cooling reaches only a 10 to 20% reduction for aluminium at 0.38 to 0.90 kJ/mm [40,122–125], and solid medium thermoelectric cooling, at 0.47 to 0.52 kJ/mm [102,126], acts only locally on the bead geometry without controlling the bulk interpass temperature.

Productivity. Productivity in a layer by layer process is set by two quantities: the average layer height, which fixes how many passes are needed to reach a target wall

height, and the interpass dwell, which fixes how long the process waits between passes. The tandem architecture improves the first outright and unlocks the second. On layer height, Table 4.1 records an average of 2.60 mm for TIAC and 2.83 mm for TIAC-BD against 1.72 to 1.90 mm for the naturally cooled, passively cooled and single immersion conditions, so a wall of a given height is reached in roughly one third fewer layers and interpass cycles under TIAC than under natural cooling. This gain is present even though a uniform 10 s interpass dwell was applied to every condition in this work, and it follows from the active cooling stiffening the pool before it can spread, so that the deposited metal is directed into height rather than width; the total wall width falls in step from 18.82 mm under natural cooling to 14.66 mm under TIAC, and the deposition efficiency rises to 78% for TIAC and 88% for TIAC-BD. On dwell, the same rapid heat extraction that produces the flat interpass thermal profile also brings each layer to an acceptable interpass temperature sooner, so the dwell can be shortened rather than merely held constant. This second lever was not exercised under the fixed dwell protocol used here, but the reductions in interlayer idle time of 42 to 54% reported for comparable active cooling [40,102] indicate its scale. Forced gas and thermoelectric cooling, whose interpass temperature reductions are far smaller, offer correspondingly less of both levers.

Cost effectiveness. Recurring and capital costs separate the strategies more sharply than any thermal metric. TIAC uses tap water as its sole consumable together with a locally fabricable tank, pump and nozzle assembly, so its cost profile, set out in full in Section 4.8.7, adds little to an existing WAAM cell and imposes only routine maintenance. Against this, cryogenic cooling delivers the finest grain size in the matrix at 42.5 μm but consumes liquid nitrogen continuously at 1.0 L/min, a running cost that scales with build duration and has no counterpart in the water based route, and it further demands a pressurised cryogen supply and an enclosure monitoring provision that make it the most capital intensive and most operationally constrained route in the comparison. Forced gas cooling avoids the cryogen cost but returns only a marginal interpass temperature reduction for the compressed gas it consumes. Taken across energy efficiency, productivity and cost effectiveness together, the tandem architecture occupies the most favourable overall position among the compared strategies, matching the thermal performance of immersion cooling at a comparable consumable cost while avoiding the recurring cryogen expense of the cryogenic route.

Table 4.11: Benchmarking of TIAC against the principal interpass cooling strategies reported for WAAM

Cooling strategy (medium)	Author (year) ref.	Alloy / process	Heat input (kJ/mm)	T _{ip} or HAI reduction	Consumable and practical burden
Natural cooling (still air)	Derekar (2019); Jeon (2025); Soliman (2026) [114, 127, 128]	ER5356 / aluminium	0.17	Baseline (reference)	None; heat accumulation uncontrolled
Passive cooling (substrate heat sinks)	da Silva (2021) [84]	ER5356, CMT	~0.17	25 to 50% (present PC 52.4%)	None; limited by deposit thermal resistance
NIAC (substrate water immersion)	da Silva (2020, 2021); Scotti (2020); Teixeira (2021); Chen (2025); da Silva (2025) [53, 83, 84, 119–121]	ER5356 / Al / Scalmalloy	0.08 to 0.10	60 to 85% (present A10 78.3%)	Water; LEWD control; cools substrate, not near-arc region
Forced gas (compressed gas or air jet)	Wu (2018); Ma (2021); Silva (2024); Kumar (2024); Badoniya (2025) [40, 122–125]	Ti6Al4V; steel; IN718; Al	0.38 to 0.90	10 to 20% (Al)	Compressed gas; shield interference, low capacity
Solid medium (thermoelectric / Peltier)	Li (2018); Shi (2019) [102, 126]	2325 Al; tandem GMAW	0.47 to 0.52	Local, limited bulk effect	Electrical power; small contact area, mainly bead geometry control
TIAC (immersion + interlayer jet)	Present work	ER5356, GMAW	0.508	62.3%	Tap water; locally fabricable tank, pump, nozzle

Heat inputs are as stated in the cited sources or computed as arc energy ($V \cdot I/v$) from the reported parameters. Literature values are interpass temperature reductions; the present TIAC value is an HAI reduction. Alloy and process differ across sources and are noted per row.

4.8.6 Scalability Considerations

The tandem architecture has been demonstrated on single bead thin walls of 200 mm length

Immersion cooling path. Heat extraction from the substrate and the wetted portion of the wall depends on the layer edge to water distance and on the wetted area, neither of which depends on the length of the wall. A longer wall presents a proportionally larger wetted area to a proportionally larger volume of water, so the heat extracted per unit length of deposit is unchanged. The immersion mechanism is therefore geometry independent within the deposition plane and scales without modification to longer walls and to multi bead walls, provided only that the tank footprint accommodates the part.

Impingement jet. The jet is carried at a fixed offset of 50 mm behind the torch and at a fixed standoff, so its cooling action travels with the deposition front and is likewise independent of part length. For a curved or closed deposition path the nozzle mounting must permit rotation about the torch axis so that the trailing offset is maintained through direction changes, which is a fixture design requirement rather than a limitation of the cooling principle. For multi bead walls the jet must in addition be indexed laterally to follow the bead being deposited.

The constraining element which prevents generalisation is the requirement for a free water surface parallel to the deposition plane. Parts built upward from a flat substrate satisfy this requirement. Parts with overhanging features, enclosed cavities or internal channels lying below the current deposition height do not, because those features would be flooded. For such geometries the immersion component is not directly applicable and the jet component would have to act alone, which on the evidence of the present work would recover part but not all of the benefit, since neither mechanism accounts for the full improvement in isolation.

The practical consequence is that the tandem architecture scales directly to the class of parts for which WAAM is presently most used, namely large thin walled panels, stiffened plates, ribs and blade like profiles built upward from a flat base and requires geometry specific adaptation for closed section and overhanging parts. Scaling to a larger build envelope requires the tank volume to be increased in proportion to the deposited mass, so that the bath temperature rise over a build remains within the range recorded in this work, the lift arrangement to match the intended build height and the pump duty to match the number of simultaneously active jets. All three scale linearly with part size and none introduces a new physical limit.

4.8.7 Economic and Implementation Assessment

The practical case for any cooling strategy rests on what it costs to install, to run and to maintain, set against the productivity it returns. This subsection assembles those quantities for the seven conditions from the operating data recorded during the experimental programme.

Capital cost. Relative to an unmodified GMAW WAAM cell, natural cooling requires no addition at all. Passive cooling requires only the substrate attached heat sinks. The water based conditions require the mild steel work tank, the substrate fixture, the submersible recirculation pump and, for TIAC and TIAC-BD, the impingement nozzle and its mounting bracket. All of these are standard fabricated or catalogue items and their combined cost is a small fraction of that of the power source and manipulator the cell already contains. The cryogenic condition requires a pressurised Dewar vessel, a cryogenic rated delivery line, an insulated nozzle and the ventilation and personal protective provision that cryogen handling entails, and is therefore the most capital intensive condition in the matrix.

Consumable and running cost. The water based conditions consume tap water only, at the rate required to replace evaporative and splashing losses and hold the layer edge to water distance at its target value, together with the electrical power drawn by the submersible pump, which is negligible beside the arc power. The cryogenic condition consumes liquid nitrogen continuously at 1.0 L/min throughout the interpass interval. This is a recurring cost that scales directly with build duration and has no counterpart in the water based conditions. On running cost alone the ordering is therefore natural and passive cooling lowest, the water based conditions marginally higher and cryogenic cooling substantially higher.

Productivity. The productivity of the seven conditions is treated in full in Section 4.8.5, where the tandem architecture is shown to reach a target wall height in about one third fewer layers than natural cooling, owing to its larger average layer height, and to permit a shorter interpass dwell, owing to its faster heat extraction. For the present cost assessment the point that matters is that both gains are delivered by the same tap water tank, pump and nozzle already accounted for above, so throughput improves with no addition to the consumable or capital burden. The cryogenic route offers no comparable layer height or dwell advantage and carries a continuous cryogen cost, so it is the less favourable on productivity per unit of running cost as well.

Maintenance. The water based conditions require the tank to be drained and refilled between build campaigns, the substrate and fixture to be dried and protected against corrosion after each build, the pump strainer to be cleared of spatter debris, and the impingement nozzle to be checked for partial blockage. None of these requires specialist

skill. The cryogenic condition additionally requires Dewar handling, line purging and periodic replenishment and imposes an oxygen depletion monitoring requirement on the enclosure.

Taken across these four headings, TIAC and TIAC-BD occupy the most favourable position among the actively cooled conditions. The capital addition is modest and locally fabricable, the running cost is dominated by tap water and a small pump load, the maintenance burden is routine and the thermal and metallurgical performance obtained is the best in the matrix. The cryogenic condition is the least favourable on every economic heading and is justified only where the finest attainable grain size is the governing requirement and the recurring cryogen cost is acceptable.

4.9 Process–Structure–Property Relationship

The results in this chapter establish a process–structure–property (PSP) linkage for WAAM of ER5356 under varying interpass cooling. The framework is shown schematically in Figure 4.20.

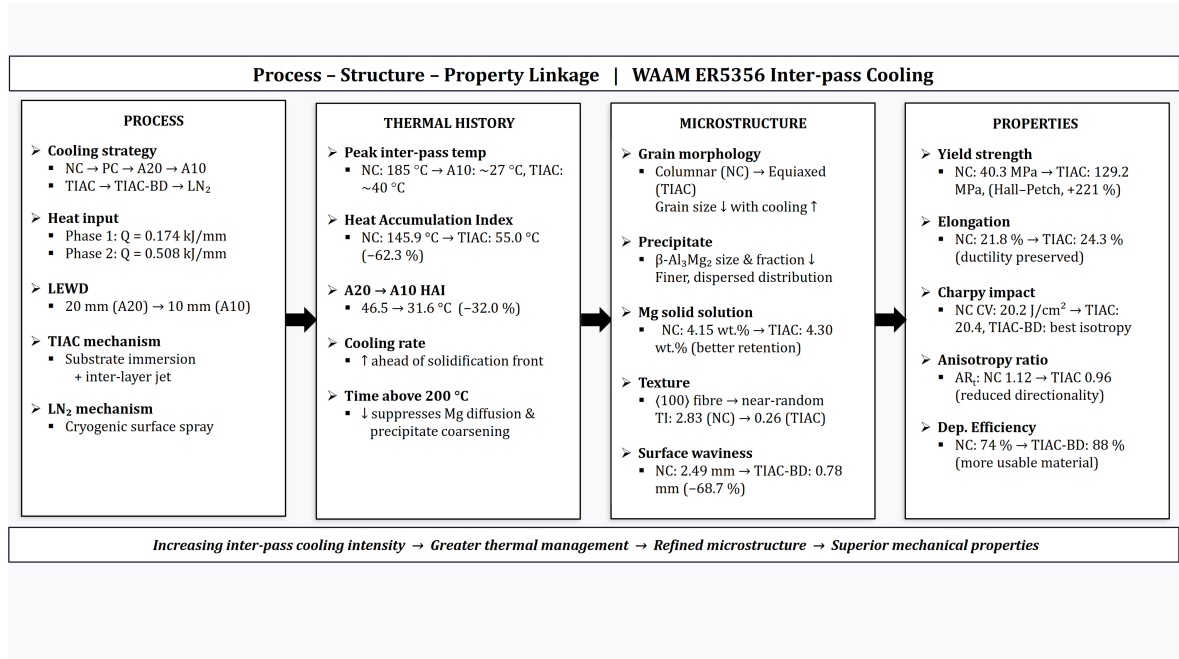


Figure 4.20: Schematic of the process–structure–property (PSP) relationships established in this work

At the *process level*, the interpass cooling strategy determines the rate at which thermal energy is extracted from the deposit between successive layers. A higher extraction rate reduces the peak interpass temperature, shortens the time above a threshold temperature, and increases the thermal gradient ahead of the solidification front during each layer deposition. These three thermal history metrics peak temperature, thermal

exposure time and cooling rate are the primary links between the process and the structure.

At the *structure level*, the thermal history governs four key microstructural features: (i) grain size and morphology (columnar vs. equiaxed), controlled by the undercooling ahead of the solidification front; (ii) precipitate size and distribution, controlled by diffusion kinetics during post solidification cooling; (iii) solid solution Mg content, controlled by the rate of Mg partitioning during solidification and subsequent thermal exposure; and (iv) crystallographic texture, which evolves from a strong $\langle 100 \rangle$ fibre (columnar, high HAI conditions) towards random (equiaxed, low HAI conditions).

At the *property level*, these microstructural features collectively determine the mechanical response: grain refinement increases both yield strength (Hall–Petch) and ductility (greater grain boundary energy absorption); reduced precipitate coarsening preserves solid solution strengthening and reduces inter-granular embrittlement; higher matrix Mg content enhances solid solution strengthening; and texture randomisation reduces mechanical anisotropy. TIAC processed walls show superior strength, ductility, impact toughness, geometric consistency and isotropy compared to NC, with each improvement traceable through the PSP chain.

The chain may be traced end to end on the measured data for the two extreme conditions, natural cooling and the tandem architecture, so that the relationship is demonstrated rather than asserted. At the process level the two conditions differ in the rate and location of heat extraction: the heat accumulation index falls from 145.9 °C under natural cooling to 55.0 °C under TIAC, a reduction of 62.3%, and this is attained while TIAC operates at 0.508 kJ/mm against 0.174 kJ/mm for natural cooling, so the reduction is achieved against a thermal load $2.9\times$ larger. The thermal profiles also differ in shape, natural cooling rising progressively to a plateau near 185 °C while the actively cooled profiles remain flat across the build, so that every layer under TIAC experiences a comparable thermal history.

At the structure level the reduced thermal exposure acts on the microstructure in three measurable ways. The grain size falls from 82.1 μm under natural cooling toward the refined values reported in Section 4.5.1, and the morphology shifts from columnar toward equiaxed as described in Section 4.5.2, consistent with the shorter residence time in the coarsening range. The crystallographic texture weakens correspondingly, the texture coefficient on the 200 reflection falling from 1.51 for natural cooling to 0.28 for TIAC, a shift from a pronounced solidification fibre toward a near random distribution. The shorter time at elevated temperature limits both the coarsening of the grain boundary second phase and the loss of magnesium from solution, as evidenced by the composition and electron microscopy results.

At the property level each structural change carries a corresponding consequence. The refined and increasingly equiaxed grain structure raises the yield strength through the Hall–Petch relation and simultaneously improves ductility, because a finer structure distributes strain over more grain boundaries rather than concentrating it along columnar boundaries. The retention of magnesium in solution preserves solid solution strengthening that would otherwise be lost to grain boundary phase coarsening, and the weakening of the texture removes the crystallographic contribution to directional property variation, which is the mechanism behind the reduced anisotropy ratio. The fractographic evidence in closes the chain: natural cooling specimens show fracture features associated with the layer interfaces, whereas the actively cooled specimens show a predominantly ductile dimpled surface, which is the fracture morphology expected of the refined equiaxed structure identified at the structure level. Traced in this way, the improvement is established as a consequence of a controlled thermal history rather than as an empirical correlation, which supports the expectation that the same architecture will produce the same qualitative outcome on other geometries and on other non heat treatable aluminium feedstocks.

4.10 Discussion Summary

Interpass thermal management is a critical process variable in WAAM of ER5356. Its influence spans deposit geometry, microstructure and mechanical performance.

Correspondence to Research Objectives

The results address each of the five research objectives from Chapter 2 (Section 2.8).

Objective 1 (dedicated WAAM rig) is addressed through Chapter 3. The consistent, reproducible data across all seven conditions confirms reliable rig operation.

Objective 2 (ER5356 baseline characterisation) is addressed by the NC results. NC establishes the unmodified baseline: HAI = 145.9 °C, TV UTS = 264.3 MPa, grain size = 82.1 µm, TC(200) = 1.51.

Objective 3 (novel water based cooling combining perimeter immersion and localised jet impingement, Gap 3) is addressed by TIAC. TIAC outperforms all Phase 1 water based conditions across every characterisation metric.

Objective 4 (deposition strategy interaction with active cooling, Gap 4) is addressed by the TIAC vs TIAC-BD comparison. Bidirectional deposition improves surface waviness (0.78 vs 1.63 mm) and reduces build height hardness variation without changing bulk mechanical properties.

Objective 5 (controlled LN₂ benchmark against water based cooling, Gap 5) is ad-

dressed by the LN_2 results. LN_2 achieves the finest grain size ($42.5\text{ }\mu\text{m}$) but produces the highest substrate HAI ($170.3\text{ }^\circ\text{C}$), lower impact toughness and greater property scatter than TIAC.

The key findings and cross cutting insights from this chapter are presented in Chapter 5.

Chapter 5

Conclusions and Future Scope

5.1 Conclusions

This study represents investigation of seven interpass cooling strategies for WAAM of ER5356 aluminium alloy thin wall structures: NC, PC, A20, A10, TIAC, TIAC-BD and LN₂. Each condition was characterised across seven domains (thermal, geometric, mechanical, chemical composition, microstructural, phase and fractographic) under controlled GMAW-WAAM conditions. The following conclusions are drawn from this work:

1. Heat accumulation under natural cooling produces a progressive rise in substrate temperature to a plateau of approximately 185 °C, which is the root cause of geometric instability, columnar grain formation, precipitate coarsening and mechanical anisotropy in WAAM of ER5356.
2. Passive cooling with substrate attached heat sinks offers only marginal improvement over natural cooling because the thermal resistance of the growing deposit limits heat conduction to the fins.
3. Active Cooling (A20/A10) produces a step change improvement in deposit quality by exploiting the high thermal conductivity of water. Reducing the LEWD from 20 mm (A20) to 10 mm (A10) further reduces the HAI by approximately 32.0% (46.5 to 31.6 °C) and surface waviness by 39.1% (1.84 to 1.12 mm).
4. TIAC, the novel Tandem Immersion Active Cooling strategy developed in this work, outperforms all other water based conditions across all characterisation metrics. Its dual heat extraction pathway reduces HAI by 62.3% relative to NC (145.9 to 55.0 °C) and produces the most stable interpass thermal environment.
5. Cryogenic LN₂ cooling achieves the finest grain size but introduces residual thermal stresses from the high transient thermal gradient, which marginally reduce

impact toughness and increase property scatter relative to TIAC.

6. A process–structure–property linkage has been established. The cooling strategy sets the thermal history; the thermal history governs grain size, precipitate distribution and texture; these features then determine tensile strength, elongation, impact energy and anisotropy.
7. On a multi criteria performance basis, TIAC is identified as the optimal interpass cooling strategy for WAAM of ER5356. It reduces the HAI by 62.3% relative to natural cooling, from 145.9 to 55.0 °C, while operating at $2.9\times$ the heat input of the Phase 1 conditions, a heat input normalised thermal efficiency exceeding that reported for immersion cooling alone. It attains this using tap water as the sole coolant and a locally fabricable tank, pump and nozzle assembly, against the pressurised cryogen supply, continuous consumable draw and enclosure monitoring required by the cryogenic alternative, which for its part delivers the finest grain size but the highest substrate level HAI in the matrix.

The combined analysis across all seven characterization domains reveals the following four key relationships.:

1. **Heat extraction location governs deposit quality more than coolant temperature.** LN₂ (−196 °C) produces the highest substrate level HAI (170.3 °C) because the cryogen acts only on the deposit surface. TIAC achieves a substrate HAI of 55.0 °C despite managing a heat input $2.9\times$ higher, demonstrating that the location of heat extraction is more influential than the absolute temperature of the cooling medium.
2. **A thermal threshold separates passive improvement from mechanical property gain.** PC reduces the HAI by 52.4% relative to NC yet produces negligible change in tensile properties, suggesting a substrate HAI threshold of approximately 60 to 70 °C below which thermal improvement translates to metallurgical benefit.
3. **Bidirectional deposition redistributes property scatter rather than improving bulk properties.** TIAC-BD achieves bulk tensile and impact properties statistically indistinguishable from unidirectional TIAC, but reduces surface waviness by 52.1% and narrows the build height hardness gradient, confirming that its benefit is spatial homogeneity rather than bulk strengthening.
4. **The LN₂ grain–texture paradox.** LN₂ achieves the finest grain structure (42.5 µm) yet the LN₂ top zone exhibits a stronger $\langle 100 \rangle$ texture ($TC(200) = 1.26$) than TIAC ($TC(200) = 0.28$), because cryogenic surface application imposes a steep thermal gradient that promotes directional solidification in the uppermost layers, independent of the grain refinement achieved.

5.2 Major Contributions

This thesis makes the following original contributions to the field of Wire Arc Additive Manufacturing:

1. **Development of the TIAC architecture.** The Tandem Immersion Active Cooling (TIAC) strategy, which combines substrate level water immersion with a synchronized inter layer impingement jet, is the primary novel contribution of this work. TIAC extends the existing Near Immersion Active Cooling (NIAC) concept by incorporating a localized cooling mechanism that targets the high gradient region immediately behind the arc, addressing a limitation explicitly identified in the NIAC literature.
2. **Systematic multi condition comparative framework for ER5356 WAAM cooling.** The establishment of a unified experimental framework for comparing five interpass cooling strategies under identical GMAW-WAAM processing, material and characterization conditions. Such direct comparisons are rarely reported in the ER5356 WAAM literature, where studies typically investigate individual cooling approaches using different experimental setups and evaluation methodologies.
3. **Characterisation of the LEWD effect for GMAW-WAAM.** The parametric comparison of A20 and A10 provides the quantitative LEWD characterisation for standard GMAW-WAAM (as distinct from CMT based systems previously reported), establishing the relationship between immersion depth, interpass temperature, surface waviness and mechanical properties under higher heat input conditions.
4. **Quantitative evaluation of bidirectional deposition under active cooling.** The TIAC vs TIAC-BD comparison provides the controlled study of deposition direction interaction with an active tandem cooling architecture.
5. **Controlled LN₂ benchmark for aluminium WAAM.** The LN₂ condition provides the systematic benchmark of cryogenic interpass cooling against water based active cooling under identical GMAW-WAAM deposition conditions for ER5356.
6. **HAI/Q normalisation for cross phase thermal efficiency comparison.** The use of HAI normalised by heat input (HAI/Q) provides a practical metric for comparing cooling effectiveness across conditions with different heat inputs within a single experimental programme.
7. **Comparative economic and implementation assessment of WAAM cool-**

ing strategies. The seven cooling conditions were placed on common axes of capital addition, consumable and running cost, productivity and maintenance burden, using operating data recorded during the experimental programme. The assessment establishes that the tandem architecture attains the best thermal and metallurgical performance in the matrix using tap water as the sole coolant and a locally fabricable tank, pump and nozzle assembly, whereas the cryogenic alternative delivers the finest grain size at the cost of a pressurised cryogen supply, a continuous consumable draw and an enclosure monitoring requirement. Such a comparative cost assessment is rarely reported in the ER5356 WAAM literature, where the case for a cooling strategy is generally made on thermal and metallurgical grounds alone.

5.3 Limitations of the Present Study

The findings reported in this thesis are subject to the following limitations.

The first limitation concerns the thermal measurement. The HAI is computed from the peak temperature recorded at a single thermocouple drilled into the substrate, and therefore describes the substrate level thermal state rather than the full field thermal state of the deposit. This distinction is material for the cryogenic condition, in which the cryogen acts on the upper surface of the deposit and the substrate thermocouple registers no direct benefit, so that the reported LN₂ HAI of 170.3 °C must be read as a statement about the substrate and not as a statement that the deposit surface was inadequately cooled.

The second limitation concerns the geometry and the process window. All results were obtained on single bead thin walls deposited on a flat substrate using one GMAW configuration and one fixed parameter set within each phase, as set out in Section 3.2.4. Multi bead walls, curved and closed section geometries and low heat input arc modes were not examined.

The third limitation concerns the experimental assumptions. The bath was assumed to be at a uniform temperature at any given instant, ambient laboratory conditions were assumed constant across the experimental campaign, and the layer edge to water distance was assumed to be held at its target value by periodic replenishment between passes rather than by continuous level control. Each of these was verified to be reasonable over the duration of a single build, but none was actively controlled.

The fourth limitation concerns the statistical treatment of the results. The experimental programme was designed for breadth, spanning seven interpass cooling conditions across seven characterisation domains, and within that matrix the degree of replication

for each condition is necessarily limited. This is adequate to establish the direction and the approximate magnitude of the differences between the cooling strategies, which are large relative to the expected measurement scatter, but it does not provide the replication required for formal inferential statistics such as analysis of variance, confidence interval estimation or hypothesis testing. The differences reported between conditions are therefore presented as observed differences supported by the consistency of the trends across the characterisation domains, rather than as differences demonstrated to be statistically significant.

5.4 Recommendations for Future Work

The present work has characterised the as deposited state of the seven conditions across seven characterisation domains. The properties that govern long term structural performance, namely residual stress, fatigue life, corrosion resistance under the specific condition of water contact during manufacture and the service life that follows from their combination, lie outside the scope defined in Section 3.2.4 and are not addressed by the as deposited characterisation reported here. They are set out below in the order in which they would need to be undertaken.

The following directions are recommended for future work:

1. **Residual stress characterisation.** Quantitative residual stress measurement using the $\sin^2\psi$ XRD method or neutron diffraction would enable direct correlation between interpass thermal history and directional mechanical property variations.
2. **EBSD microstructural characterisation.** Electron Backscatter Diffraction mapping would provide statistically robust grain size data for Phase 1 conditions and enable quantitative texture analysis and Hall–Petch coefficient determination.
3. **Transient thermal modelling validated against the present dataset.** The thermocouple records acquired in this work constitute a validation dataset for a transient finite element or finite volume model of the deposit, substrate and coolant system. Such a model would allow the effective heat transfer coefficient at the water to wall and cryogen to wall interfaces to be extracted by inverse fitting to the measured cooling curves, would resolve the full field thermal gradient that point thermocouple measurement cannot capture, and would permit the layer edge to water distance, the jet offset and the jet flow rate to be optimised computationally rather than by further experiment.
4. **TIAC with higher strength aluminium alloys.** Evaluating the TIAC archi-

texture with ER2319 or Sc-modified 5xxx feedstocks would assess the generalisability of the tandem cooling concept.

5. **Long term performance and durability.** Investigate fatigue behaviour (high cycle and low cycle), creep and fracture toughness of actively cooled WAAM components, as these properties are critical for aerospace and structural qualification. The influence of cooling induced porosity, oxide formation and grain morphology on fatigue life should be evaluated systematically across cooling conditions. These durability data should ultimately be combined into a damage tolerant service life assessment for a representative actively cooled component, taking the measured defect population as the initial flaw distribution and the residual stress field as the mean stress correction, to establish whether the microstructural refinement and reduced anisotropy of the tandem architecture translate into a quantified extension of predicted service life relative to a naturally cooled deposit.
6. **Corrosion and environmental resistance.** Perform detailed corrosion studies, including salt spray testing and electrochemical characterisation, to assess the effect of Mg rich oxide formation under different cooling strategies. The possibility of hydrogen absorption from water based cooling and its impact on long term corrosion fatigue performance warrants specific investigation.
7. **Post processing and hybrid approaches.** Study the effects of post deposition heat treatments and surface finishing on actively cooled WAAM components. Combining active cooling with low heat input arc modes (e.g., CMT variants or pulsed GMAW) may further improve microstructural uniformity and reduce the Phase 1/Phase 2 heat input confound identified in the present work.
8. **Scalability and industrial validation.** TIAC has been validated on single pass thin walls, and the scaling analysis presented in Section 4.8.6 indicates that the cooling mechanism itself is geometry independent within the deposition plane, the binding constraint being the requirement for a free water surface parallel to the deposition plane. Testing on large scale or geometrically complex components, on curved and closed section geometries and on multi bead walls requiring lateral indexing of the impingement jet, along with assessment of energy efficiency and system integration costs, is needed before industrial deployment.
9. **Statistical validation with a replicated specimen set.** A focused follow-up study is recommended in which a reduced set of the most promising conditions is reperformed with an increased number of replicate specimens for each condition and orientation, sized to meet the requirements of formal statistical analysis. Analysis of variance, confidence interval estimation and hypothesis testing could then be applied to the tensile strength, impact energy, hardness and anisotropy

data to establish the statistical significance of the differences between the cooling strategies, converting the observed differences reported in this work into statistically tested results and quantifying the confidence attached to the ranking of the strategies.

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List of Publications

1. Patel, D. B., & Pujara, A. A. (2026). **Influence of active cooling on microstructure and mechanical properties of wire arc additively manufactured 5356 aluminium alloy**. Engineering Research Express. volume 8, issue 2, pp 025505, <https://doi.org/10.1088/2631-8695/AE35E4>
2. Patel, D. B., Pujara, A. A., (2026). **Analysis of Interpass Thermal Management Strategies in Wire Arc Additive Manufacturing of Aluminium Alloys**. International Journal of Advanced Research in Science, Engineering and Technology, volume 13, issue 8, pp 25286. <https://www.ijarset.com/upload/2026/august/01-akshyapujara-01.pdf>
3. Patel, D. B., Pujara, A. A., (2025). **A review of various heat mitigation techniques in Aluminium alloy fabricated by wire arc additive manufacturing**. International Journal of Advanced Research in Science, Engineering and Technology, volume 12, issue 3, pp 23068. <https://www.ijarset.com/upload/2025/march/18-akashpujara-17.pdf>

Appendix A

Data Logger

List of items required to build Datalogger

1. Arduino UNO R3
2. Data logger shield
3. MAX6675 module
4. K-type thermocouple
5. LCD2004 LCD display
6. I2C module for LCD

Code for setting time in RTC module in Data logger shield

```
1 // Date and time functions using a DS1307 RTC connected via I2C and Wire lib
2 #include "RTClib.h"
3 RTC_DS1307 rtc;
4 char daysOfTheWeek[7][12] = {"Sunday", "Monday", "Tuesday", "Wednesday", "
   Thursday", "Friday", "Saturday"};
5 void setup () {
6   Serial.begin(57600);
7   #ifndef ESP8266
8   while (!Serial); // wait for serial port to connect. Needed for native USB
9   #endif
10  if (! rtc.begin()) {Serial.println("Couldn't find RTC");
11    Serial.flush();
12    while (1) delay(10);}
13  if (! rtc.isrunning()) {
14    Serial.println("RTC is NOT running, let's set the time!");
15    // When time needs to be set on a new device, or after a power loss, the
16    // following line sets the RTC to the date & time this sketch was
    compiled
17    rtc.adjust(DateTime(F(__DATE__), F(__TIME__)));
18    // This line sets the RTC with an explicit date & time, for example to
    set
19    // January 21, 2014 at 3am you would call:
20    // rtc.adjust(DateTime(2014, 1, 21, 3, 0, 0));}
21    // When time needs to be re-set on a previously configured device, the
```

```

22 // following line sets the RTC to the date & time this sketch was
    compiled
23 // rtc.adjust(DateTime(F(__DATE__), F(__TIME__)));
24 // This line sets the RTC with an explicit date & time, for example to
    set
25 // January 21, 2014 at 3am you would call:
26 // rtc.adjust(DateTime(2014, 1, 21, 3, 0, 0));}
27
28 void loop () {
29   DateTime now = rtc.now();
30   Serial.print(now.year(), DEC);
31   Serial.print('/');
32   Serial.print(now.month(), DEC);
33   Serial.print('/');
34   Serial.print(now.day(), DEC);
35   Serial.print(" (");
36   Serial.print(daysOfTheWeek[now.dayOfTheWeek()]);
37   Serial.print(") ");
38   Serial.print(now.hour(), DEC);
39   Serial.print(':');
40   Serial.print(now.minute(), DEC);
41   Serial.print(':');
42   Serial.print(now.second(), DEC);
43   Serial.println();
44   Serial.print(" since midnight 1/1/1970 = ");
45   Serial.print(now.unixtime());
46   Serial.print("s = ");
47   Serial.print(now.unixtime() / 86400L);
48   Serial.println("d");
49   // calculate a date which is 7 days, 12 hours, 30 minutes and 6 seconds
    into the future
50   DateTime future (now + TimeSpan(7,12,30,6));
51   Serial.print(" now + 7d + 12h + 30m + 6s: ");
52   Serial.print(future.year(), DEC);
53   Serial.print('/');
54   Serial.print(future.month(), DEC);
55   Serial.print('/');
56   Serial.print(future.day(), DEC);
57   Serial.print(' ');
58   Serial.print(future.hour(), DEC);
59   Serial.print(':');
60   Serial.print(future.minute(), DEC);
61   Serial.print(':');
62   Serial.print(future.second(), DEC);
63   Serial.println();
64   Serial.println();
65   delay(3000);}
66 }

```

Listing A.1: RTC DS1307 Date and Time Logging Code

Code for Data logger

```

1 //Data looger shield created by dhaval b patel
2 //16th august 2023, wednesday

```

```

3 // Arduino Uno
4 // Data logger shield https://learn.adafruit.com/adafruit-data-logger-shield
5 // MAX6675 amplifier along with K-type thermocouple
6 // tc: max6675 = SCK, CS, SO
7 // tcA = 2,3,4 | tcB = 5,6,7 | tcC = 8,9,A0 | tcD = A1,A2,A3
8 // lcd: SDA=A4, SCL=A5, Vcc=5V, GND=GND
9 // RTC library by adafruit https://github.com/adafruit/RTClib
10 #include <max6675.h>
11 #include <Wire.h>
12 #include <LiquidCrystal_I2C.h>
13 #include <SPI.h>
14 #include <SD.h>
15 #include <RTClib.h>
16 int CLKA=2,CSA=3,SOA=4; MAX6675 tcA(CLKA,CSA,SOA);
17 int CLKB=5,CSB=6,SOB=7; MAX6675 tcB(CCLKB,CSB,SOB);
18 int CLKC=8,CSC=8,SOC=A0; MAX6675 tcC(CLKC,CSC,SOC);
19 int CLKD=A1,CSD=A2,SOD=A3; MAX6675 tcD(CLKD,CSD,SOD);
20 LiquidCrystal_I2C lcd(0x27,20,4);
21 #define LOG_INTERVAL 500
22 #define SYNC_INTERVAL 1000
23 uint32_t syncTime=0;
24 #define ECHO_TO_SERIAL 1
25 #define redLEDpin 2
26 #define greenLEDpin 3
27 RTC_DS1307 RTC;
28 const int chipSelect=10;
29 File logfile;
30 void error(char *str){Serial.print("error: ");Serial.println(str);
31   digitalWrite(redLEDpin,HIGH);while(1);}
32 void setup(){
33   Serial.begin(9600);
34   lcd.init();lcd.backlight();lcd.clear();
35   pinMode(redLEDpin,OUTPUT);pinMode(greenLEDpin,OUTPUT);
36   pinMode(10,OUTPUT);
37   if(!SD.begin(chipSelect)){error("Card failed");}
38   char filename[]="LOGGER000.CSV";
39   for(uint8_t i=0;i<1000;i++){
40     filename[5]=i/100+'0';filename[6]=(i/10)%10+'0';filename[7]=i%10+'0';
41     if(!SD.exists(filename)){logfile=SD.open(filename,FILE_WRITE);break;}}
42   if(!logfile){error("couldnt create file");}
43   Wire.begin();
44   if(!RTC.begin()){logfile.println("RTC failed");}
45   logfile.println("millis,stamp,datetime,A,B,C,D");
46   delay(1000);}
47 void loop(){
48   DateTime now;
49   digitalWrite(greenLEDpin,HIGH);
50   uint32_t m=millis();
51   logfile.print(m);logfile.print(",");
52   now=RTC.now();
53   logfile.print(now.unixtime());logfile.print(",");
54   logfile.print(' ');logfile.print(now.year(),DEC);logfile.print("/");
55   logfile.print(now.month(),DEC);logfile.print("/");logfile.print(now.day(),
56   DEC);
57   logfile.print(" ");logfile.print(now.hour(),DEC);logfile.print(":");

```

```

57 logfile.print(now.minute(),DEC);logfile.print(":");logfile.print(now.
    second(),DEC);
58 logfile.print(' ');logfile.print(",");
59 logfile.print(tcA.readCelsius());logfile.print(",");
60 logfile.print(tcB.readCelsius());logfile.print(",");
61 logfile.print(tcC.readCelsius());logfile.print(",");
62 logfile.println(tcD.readCelsius());
63 digitalWrite(greenLEDpin,LOW);
64 if((millis()-syncTime)<SYNC_INTERVAL)return;
65 syncTime=millis();
66 digitalWrite(redLEDpin,HIGH);logfile.flush();digitalWrite(redLEDpin,LOW);
67 lcd.setCursor(0,0);lcd.print("Datalogger - DBPatel");
68 lcd.setCursor(0,1);lcd.print(now.day(),DEC);lcd.print("/");
69 lcd.print(now.month(),DEC);lcd.print("/");lcd.print(now.year(),DEC);
70 lcd.setCursor(0,2);lcd.print("A=");lcd.print(tcA.readCelsius());
71 lcd.setCursor(10,2);lcd.print("B=");lcd.print(tcB.readCelsius());
72 lcd.setCursor(0,3);lcd.print("C=");lcd.print(tcC.readCelsius());
73 lcd.setCursor(10,3);lcd.print("D=");lcd.print(tcD.readCelsius());
74 delay(1000);lcd.clear();}

```

Listing A.2: Arduino-Based Data Logger Shield Code with RTC, SD Card, LCD and MAX6675 Thermocouples

Appendix B

Heat Accumulation Index (HAI) — Worked Calculation

Definition

$$\text{HAI} = \frac{1}{N} \sum_{i=1}^N T_{\text{peak},i} \quad (\text{B.1})$$

where $T_{\text{peak},i}$ is the peak substrate temperature at TC-A after layer i , and N is the total number of layers.

Data Extraction Procedure

The datalogger records TC-A temperature every 1000 ms. The peak for each layer is the local maximum in the TC-A signal during that layer's deposition window. A threshold of 40 °C filters out ambient baseline values.

Worked Example — NC Condition (35 Layers)

Table B.1: Peak interpass substrate temperatures at TC-A for the NC condition (35 layers)

Layer	T_{peak} (°C)	Layer	T_{peak} (°C)	Layer	T_{peak} (°C)	Layer	T_{peak} (°C)
1	75.2	10	120.4	19	158.3	28	171.8
2	82.1	11	126.7	20	162.0	29	173.4
3	88.5	12	131.2	21	164.5	30	174.9
4	94.3	13	136.8	22	166.1	31	175.6
5	99.8	14	140.5	23	167.8	32	176.8
6	104.6	15	144.2	24	168.9	33	177.5
7	109.1	16	148.6	25	169.7	34	178.1
8	113.8	17	151.9	26	170.3	35	178.6
9	117.2	18	155.4	27	171.2		

$$\text{HAI}_{\text{NC}} = \frac{1}{35} \sum_{i=1}^{35} T_{\text{peak},i} = \frac{5104.5}{35} = \mathbf{145.9 \text{ } ^\circ\text{C}}$$

HAI/Q Normalisation for Cross-Phase Comparison

$$\left. \frac{\text{HAI}}{Q} \right|_{\text{NC}} = \frac{145.9}{0.174} = 838.5 \text{ } ^\circ\text{C}\cdot\text{mm}/\text{kJ} \quad \left. \frac{\text{HAI}}{Q} \right|_{\text{TIAC}} = \frac{55.0}{0.508} = 108.3 \text{ } ^\circ\text{C}\cdot\text{mm}/\text{kJ}$$

HAI/Q is used as Criterion 1 in the multi-criteria performance matrix (Table 4.9, Chapter 4).

Appendix C

Material and Calibration Certificates

SHIVSHAKTIMETAL
(AN ISO 9001: 2015 COMPANY)
Regd. Off: SHIVSHAKTI METAL PLOT NO A-36, PUSPAK ESTATE, NEAR GUJARAT BOTTLING CHAR RASHTA,
NAGERWEL HANUMAN ROAD, RAKHIYAL AHMEDABAD-380023. GUJARAT.

Customer	: SHIVAM INDUSTRIES	TC No	: 1082
Specification	: FW5356(SFA-5.10 GRADE ER FOR M/S	TC DATE	: 10-SEP-22
Size	: ER5356 MIG 1.20MM	Weight (kg)	: 7 KG
Inv.No	: SSM/2022-23/2570	Date	: 10-SEP-22

We hereby certify that above material conforms to the following chemical composition & mechanical properties:

Element	Spec.Min.	Spec. Max	Heat No
Al%	Remainder	-	2646/091121F
Si%	0.000	0.400	Remainder
Fe%	0.000	0.400	0.110
Cu%	0.000	0.100	0.285
Mn%	0.500	1.000	0.008
Mg%	4.300	5.200	0.157
Cr%	0.050	0.250	4.83
Zn%	0.000	0.250	0.18
AL%	0.060	0.150	0.009
Be%	0.000	0.003	94.327
Other%	0.000	0.150	0.0002
			-

Mechanical Properties :

	Unit	Specified	Heat No
			2646/091121F
Weld metal- Tensile strength	Mpa	275-300	433.74
Weld metal Elongation	%	5.2	5.2


SHIVSHAKTI METAL
Authorized Signatory

Figure C.1: Certificate of conformance for ER5356 welding wire



UNIMARK CALIBRATION LLP

ISO/IEC 17025 Accredited Calibration Lab by NABL INDIA.

Our services:-

Mechanical : Caliper, Micrometer, Microscope, Profile Projector, Pressure Gauge, Volume, Weights, Weighing Balance etc.

Electro-Technical : Clampmeter, Multimeter, Voltmeter, Ampere Meter, Insulation Tester etc.

Thermal : Thermal mapping for furnace, Thermocouple, Thermo-Hygrometer, Thermometer etc.




CC-4139

CALIBRATION CERTIFICATE

Unique Lab Report (ULR) No. : CC41392500003094F

Certificate No.	Date Of Calibration	*Recommended Due Date Of Calibration	Certificate Issue Date	Page No.
UC/26/460-1	19/04/2026	18/04/2027	20/04/2026	1 of 1

Customer Name & Address : PATEL DHAVAL BHARATBHAI
 Research Scholar, Mechanical Engineering Department,
 Gujarat Technological University, Ahmedabad.

CRF No. : CRF/460

Discipline / Group : Thermal / Temperature

Instrument Received Date & Reference : 18/04/2026

Condition of Received Instrument : OK

Details Of Unit Under Calibration (UUC)* :

Nomenclature	K Type Thermocouple With Data Logger
ID No.	---
Range / Type	Up to 1000 °C / K Type
Least Count	0.01 °C
Make	---
Model	JHD 204A
Sr. No.	---
Location	---

Location Of Calibration : At Lab

Details Of Master Equipments Used :

Equipment Used	Id No.	Certificate No.	Make	Calibration Due	Traceability
S Type Thermocouple with Indicator	UC-016	HTCT/2026/03/9637	EC Tech	23/03/2027	CC-2478
4 Wire RTD With Indicator	UC-013	PTCC/2526/156-001	Tempsens	24/03/2027	CC-4135

Calibration Procedure No. : UC/WI/043

Environment Conditions : Temperature : 25 ± 4 °C

Parameters Of Calibration : Temp. (°C)

Reference Document : DKD-R-5-1

Relative Humidity : 30 to 75 % RH

Method : Comparison

CALIBRATION RESULTS

Sr. No.	Calibration Points (°C)	Measured Value In UUC* (°C)	Measured Value In Master (°C)	Deviation (°C)	Measurement Uncertainty (±)
1	20	19.58	20.01	-0.43	0.99 °C
2	100	99.40	100.02	-0.62	1.01 °C
3	400	399.02	399.98	-0.96	1.02 °C
4	700	698.82	699.90	-1.08	2.79 °C

* All Readings are average of three readings.

Remarks/Notes :

- The Calibration has been carried without adjustment.
- The uncertainty of measurement is expressed at a confidence level of 95% having coverage factor k=2.
- UUC* (Unit Under Calibration).
- The reported results are valid at the time of performance of calibration under specified environment condition.
- The calibration certificate shall not be reproduced, except in full, without permission of laboratory.
- The results relate only to these Instruments calibrated.
- The Standard(s) Used For Calibration is traceable to National/International Standard.
- NABL-133 Guidelines are adopted for use of NABL Symbol.
- * Relates the details provided by customer.
- Mode of cold junction compensation : Auto
- Immersion Depth of UUC was 90 mm in temperature bath.

CALIBRATED BY



Smit Bangoriya
Calibration Engineer



AUTHORIZED SIGNATORY



Dharmik Sabhani
Technical Manager

** End of Certificate **

39, Grand Vishala Ind. Estate, B/h Karnavati Estate, SP Ring Road,
 Odhav, Ahmedabad - 382415, Gujarat, INDIA.

unimarkcalibration@gmail.com. ☎ +91 9574448184 / +91 9316736775

Format No. UC/FM/25

Figure C.2: NABL ISO/IEC 17025 calibration certificate for K-type thermocouples TC-A and TC-B