Microstructure and mechanical properties of wire arc additively manufactured Hastelloy C276 alloy

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Qiu, Zhijun; Wu, Bintao; Zhu, Hanliang; Wang, Zhiyang; Hellier, Alan K.; Ma, Yan; Li, Hui Jun; Muransky, Ondrej; and Wexler, David, "Microstructure and mechanical properties of wire arc additively manufactured Hastelloy C276 alloy" (2020). Faculty of Engineering and Information Sciences - Papers: Part B. 4355. https://ro.uow.edu.au/eispapers1/4355

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Abstract
© 2020 The Authors In this study, a Hastelloy C276 thin-wall structure was fabricated by wire arc additive manufacturing (WAAM). The microstructure and mechanical properties of the as-deposited structure were evaluated in detail using specimens extracted from different orientations and locations along the deposition direction. The results show that the primary dendrite arm spacing, dislocation density, and second-phase precipitates change along the deposition direction of the material, and they are responsible for the occurrence of anisotropy and heterogeneity in the observed mechanical properties. However, the well-distributed hardness values generated in the as-deposited sample are considered to be due to the uniform directional dendrites. This study enables a better understanding of the WAAM processing-microstructure-properties relationship for the nickel-based alloy, providing useful information for process optimization and improvement in the resultant mechanical properties.

Disciplines
Engineering | Science and Technology Studies

Publication Details

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This journal article is available at Research Online: https://ro.uow.edu.au/eispapers1/4355
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HIGHLIGHTS

• Defect-free Hastelloy C276 alloy wall structure was produced using WAAM.
• Microstructure of component was characterized in various directions and locations.
• Anisotropic mechanical properties existed in as-built alloy were discussed.

ABSTRACT

In this study, a Hastelloy C276 thin-wall structure was fabricated by wire arc additive manufacturing (WAAM). The microstructure and mechanical properties of the as-deposited structure were evaluated in detail using specimens extracted from different orientations and locations along the deposition direction. The results show that the primary dendrite arm spacing, dislocation density, and second-phase precipitates change along the deposition direction of the material, and they are responsible for the occurrence of anisotropy and heterogeneity in the observed mechanical properties. However, the well-distributed hardness values generated in the as-deposited sample are considered to be due to the uniform directional dendrites. This study enables a better understanding of the WAAM processing-microstructure-properties relationship for the nickel-based alloy, providing useful information for process optimization and improvement in the resultant mechanical properties.

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1. Introduction

There has been an increased focus from the academic community and manufacturing industry on the area of additive manufacturing (AM), owing to its ascendancy in component fabrication over the past few decades. AM technologies may be classified into the powder-based process and the wire-based process based on the supplement of additive manufacturing materials. Wire arc additive manufacturing (WAAM) is one type of wire-based process in which an electrical arc is adopted as a heat source to melt welding wire for fabrication of 3D components in a layer-by-layer deposition strategy [1]. The WAAM approach is advantageous for the production of large metallic components of complex geometry, due to its high deposition rate in comparison to...
the other AM approaches using laser or electron beams as heating sources, plus it requires relatively low apparatus investment. These characteristics enable a significant reduction in capital cost and efficient usage of the raw materials [2]. Currently, various metallic materials, such as titanium alloys, aluminum, nickel-based superalloys, and titanium aluminate intermetallics have been successfully utilized in the WAAM process [3–5]. Nickel-based superalloys offer superior corrosion resistance and excellent mechanical properties at ambient and elevated temperatures, thereby enabling widespread applications in the aerospace, nuclear, and chemical industries, and remain to be constantly improved due to accelerated advancement of these fields [6–8]. Hastelloy alloys are a typical family of Ni-based alloys used in these fields; research on Hastelloy series alloys, therefore, has both scientific and industrial significance. Among this family, Hastelloy C276 is a Ni-Cr-Mo based superalloy, which has been widely applied in extreme corrosion environments, such as flue gas desulfurization systems, hot contaminated media. Due to its high corrosion resistance and excellent mechanical properties [9]. Numerous investigations have been conducted on the microstructure evolution, mechanical properties, machinability, and corrosion resistance of Hastelloy C276 [10–12]. Cieslak et al. [13] have proved the occurrence of p and μ phase by studying the TIG welding of Hastelloy C276, which are topologically close-packed (TCP) phases, generally being of concern as brittle intermetallic phases. The μ phase is transformed from the p phase after a long term exposure, which might be 8 h or more [14]. Ahmad et al. [15] explored the Mo and W rich micro-eutectoid (no separation of p and μ phases) in the molten zone (MZ) of welded Hastelloy C276 by the electron beam (EB) method, and found the hardness in the MZ to be higher than that for the as-received alloy due to the fine lamellar microstructure and micro-eutectic phase hardening in this region. The authors further reported the growth of the μ phase after heat treatment. Dhananchezian [10] studied the machining of Hastelloy C276 using a cryogenic cooling method which can improve the cutting tool performance. So far, the component fabrication techniques are still mainly concentrated on welding and traditional subtractive technologies, and only limited research [16] has been reported on additive manufacturing of Hastelloy C276 alloy; this makes an attempt to apply the AM process timely. It has been widely accepted that various additive manufactured components would possibly exhibit anisotropy with respect to both microstructure and mechanical properties from the bottom to the top part of the fabricated components, resulting from the line-by-line and layer-by-layer manufacturing strategy [17,18].

Hastelloy is normally supplied in forged, rolled, and extruded states, and welding has been widely used for fabricating Hastelloy components. Hastelloy is readily weldable, and it can be welded with gas tungsten arc welding (GTAW) [19], electron beam welding (EBW) [20], and laser welding (LW) [11]. However, a welded joint is generally treated as the weakest part of the manufactured component due to microstructure diversity, which is an unavoidable aspect leading to the occurrence of defects or variations in mechanical properties [7,21]. In recent years, additive manufacturing (AM) techniques have been used to fabricate the components by one grade of Hastelloy alloy, named Hastelloy X, by laser powder-bed fusion (LPBF) [22,23] and selective laser melting (SLM) [24]. LPBF and SLM, as metal powder-based AM technologies, are often associated with the concerns that the metal powder raw material does not show the same competitive efficiency with regard to cost and availability as the metal wire, in addition to the challenges concerning the properties of manufactured components caused by the potential defect of porosity [2,24]. In contrast, WAAM, as a wire-based AM technology, is a potentially cost-effective alternative to fabricate defect-free components, albeit that the post-machining of the WAAM processed components is often required due to their low geometric accuracies [25].

In this study, the gas tungsten arc welding based WAAM (GT-WAAM) process was utilized to produce Hastelloy C267 samples. The microstructure (columnar dendrites and precipitates) and mechanical performance (tensile properties and hardness) were investigated through comprehensive comparison and evaluation in different regions and orientations of the component after deposition. These results show that Hastelloy C276 components with high-quality performance can be efficiently fabricated using the GT-WAAM process.

2. Experimental procedures

2.1. Experiment setup

The WAAM system used in this study consists of a GTAW welding system, shielding gas unit, a "cold" wire feeder, and a traveling device [26]. Hastelloy C276 wire with a diameter of 1.2 mm was applied as feedstock, and its chemical composition is listed in Table 1. A thin-wall structure with dimensions of $10 \times 10 \times 45$ mm (width $\times$ length $\times$ height) was deposited on a plain carbon steel substrate, and the process parameters for the experiment are listed in Table 2. Fifty layers were deposited on a single thin-wall structure, and no adjacent layer was built. High purity argon was utilized for both the GTAW torch and the trailing shielding gas to prevent oxidation of the deposited part during the manufacturing process. Following the extinguishment of the arc for each layer, additional argon shielding gas continued to flow for about 30 s, and complemented friction by a stainless steel brush was utilized to remove potential spattering or particles from welding fume deposition. Hence, the dwell time of interpass is about 60 s without an interpass temperature control in this research.

2.2. Material characterization

Mechanical testing and metallographic specimens were extracted from the middle of an as-deposited Hastelloy C276 wall sample, as shown in Fig. 1, which also shows the traveling direction and deposition direction, and defines the selection of reference planes, y-z, x-y, and x-z. To avoid element dilution from the substrate, the test samples were extracted from the regions in the as-deposited materials of at least four layers away from the substrate. The metallographic samples were ground and polished following standard procedures, and subsequently electro-etched in a mixed solution containing 5 mg oxalic acid and 15 mL hydrochloric acid with 6 V direct current for 2 s at room temperature. Microstructures were observed using a Leica Optiphot optical microscope (OM). ImageJ software was applied to calculate the size of the dendrites [27]. The Thermo-Calc® software (2019 version) was used for the thermodynamic calculation of the Hastelloy C276 system. The characteristic fast solidification and cooling rate of the WAAM process relative to the conventional casting process were taken into account in the simulation. The classic Scheil simulation based on the well-known Scheil-Gulliver model [28] which assumes zero diffusion in the solidified material and the TCNi9 thermodynamic database for Ni-based superalloys were used to predict the phases transformation during the solidification process of the investigated Hastelloy C276 with the chemical composition listed in Table 1. Detailed observation of the matrix grain structure, precipitation, fracture mode, and chemical composition was performed using a JEOL JSM-6490LA Scanning Electron Microscope (SEM) equipped with Oxford Instruments X-Max Energy Dispersive X-ray Spectroscopy (EDS) detector. Transmission Electron Microscopy (TEM) analyses were carried out using a JEM 2200FS TEM device operating at 200 kV. The identification of phase constituents was performed using X-ray diffraction (XRD) with a Cu Kα radiation source ($\lambda = 1.5418 \text{Å}$). The data were collected over a scattering angle (2θ) of 30° to 100° on the samples taken from the longitudinal region in the y-z plane. Vickers microhardness samples taken from the y-z plane for vertical (z) and horizontal (y) direction tests were measured by applying a Matsuzawa Via-F automatic Vickers Hardness Tester at a load of 1.96 N with a spacing of 0.5 mm between each step, and indentation duration of 15 s for each test. The mechanical samples for tensile...
testing were extracted from both the vertical direction (y-z plane, deposition direction) and horizontal direction (x-y plane, traveling direction). Tensile tests were performed at a constant crosshead displacement rate of 1 mm/min at room temperature using an Instron universal testing machine equipped with a digital camera to assure precise displacement measurement.

3. Results and discussion

3.1. Macrostructure

Fig. 2 illustrates the overall macro morphology of the as-deposited sample of Hastelloy C276 using the WAAM process. Slight deformation of the substrate is observed, with a corrugated morphology surface on the sample as a result of layer-by-layer deposition. Also, the component is characterized by free of visual defect or any crack, the following microscopic observation from numerous sections proved this. It appeared to be a uniform surface finish from the visual appearance after the whole manufacturing process. These results indicate that the applied WAAM process yields the defect-free Hastelloy C276 material in this work.

3.2. Microstructural evolution

3.2.1. Typical morphology of as-deposited microstructure

The microstructures and microstructural anisotropies of the as-deposited component were thoroughly investigated from the bottom to top regions. Samples were extracted in the traveling direction and deposition direction from the previously defined y-z, x-y and x-z planes, as shown in Figs. 3–5, respectively. Typical columnar dendrites are clearly observed in the samples in different regions from top to bottom, with additional evidence for the occurrence of precipitates (dark regions in Fig. 3–f).

The longitudinal cross-section (y-z plane), shown in Fig. 3, displays a typical dendritic microstructure. The directional arrangement of these dendrites is expected to lead to a strong texture, which is the topic of future research. A large number of elongated dendrites have grown across several deposited layers and are oriented upwards, almost perpendicular to the substrate in the z-direction. These appear homogenous and aligned in the middle region (Fig. 3d), but take the form of mixed dendrites, with a cellular-like pattern in local areas of the side regions (Fig. 3b) and the first several layers in the bottom region (Fig. 3e).

Layer boundaries (fusion interfaces between deposited layers) can be readily observed, and are roughly parallel to the traveling direction (x), as shown in Fig. 3d. Formation of the epitaxial columnar dendrites along the deposition direction is attributed to heat transfer from subsequent layers, and growth which follows the highest thermal gradient. During the WAAM process, the dendrites of preceding deposited layers are partially remelted by the thermal cycling, becoming nucleation regions for the subsequent layer [29]; simultaneously, layer bands are formed (highlighted in Fig. 3d) [30]. In contrast, at the commencement of the deposition, due to rapid heat transfer into the cold substrate, the dendrite growth is facilitated during the solidification process. The more rapidly growing dendrites can encroach on space in front of the slower-growing ones, and finally suppress their growth [31]. Dendrites which experience heat accumulation from the deposition of previous layers experience relatively lower thermal dissipation. However, despite this complex thermal history, it is possible that all the dendrites survive [31,32]. In accordance with expected heat transfer, the bottom region and side regions display more cellular form without secondary dendrites, whereas the main section displays a well-aligned elongated dendritic structure. The top region in the last deposited layer, as shown in Fig. 3b and Fig. 3f, exhibits a transitional microstructure, from

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Table 1
Chemical composition of Hastelloy C276 wire used in this study.

<table>
<thead>
<tr>
<th>Element</th>
<th>Ni</th>
<th>Mo</th>
<th>Cr</th>
<th>Fe</th>
<th>W</th>
<th>C</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Co</th>
<th>V</th>
<th>Si</th>
</tr>
</thead>
<tbody>
<tr>
<td>wt%</td>
<td>Bal.</td>
<td>16.5</td>
<td>16.0</td>
<td>5.8</td>
<td>4.0</td>
<td>0.01</td>
<td>0.19</td>
<td>0.015</td>
<td>0.01</td>
<td>0.20</td>
<td>0.06</td>
<td>0.08</td>
</tr>
</tbody>
</table>

Table 2
Process parameters for WAAM deposition.

<table>
<thead>
<tr>
<th>Process parameters</th>
<th>Details</th>
</tr>
</thead>
<tbody>
<tr>
<td>Power</td>
<td>Current 140 A</td>
</tr>
<tr>
<td>Arc voltage</td>
<td>13.0 V</td>
</tr>
<tr>
<td>Speed</td>
<td>Torch travel speed 100 mm/min</td>
</tr>
<tr>
<td>Wire feed speed</td>
<td>1000 mm/min</td>
</tr>
<tr>
<td>Shielding gas</td>
<td>Welding torch 10 L/min</td>
</tr>
<tr>
<td>Shielding gas cover</td>
<td>15 L/min</td>
</tr>
<tr>
<td>Time</td>
<td>Interpass dwell time 60 s</td>
</tr>
<tr>
<td>Distance</td>
<td>Electrode to workpiece 3 mm</td>
</tr>
<tr>
<td>Angle</td>
<td>Wire feeder and substrate 30°</td>
</tr>
</tbody>
</table>

---

Fig. 1. Schematics of sample extraction for metallurgical and mechanical tests.

Fig. 2. Macro morphology of the fabricated component.
directional elongated dendrites to fine equiaxed dendrites. It is attributed to last layer experiencing an accelerated cooling induced by the atmosphere surrounding the sample and the lack of remelting as it occurs during the WAAM process [33]. Similar fine top layer microstructures have been observed in other AM processed materials including, the Ni-based alloy Inconel 625 [17], and NiTi intermetallic [34].

Similarly to the y-z plane, the x-z plane also exhibits a columnar microstructure with epitaxial dendrites growing upwards (z), as is shown in Fig. 4. However, there is a slight deviation, compared to the upward dendrite growth orientation in the y-z plane. For the y-z plane, most of the upwardly elongated dendrites occur at an angle of about 80° to the substrate, as shown in Fig. 4a. A similar phenomenon has been observed in previous WAAM investigations of Inconel 625 [17]. During the GT-WAAM process, the heat conduction of the molten pool predominantly comes from the substrate for the first few layers and previously deposited layers with buildup (z), and partially from the adjacent solidification along the traveling direction (x). Due to the temperature gradient from these two orientations, the angular grain growth appears in

Fig. 3. Microstructure morphology in cross-section (y-z plane) of as-fabricated component: (a) whole profile of the part in cross-section (y-z plane), (b-h) are representative microstructures in different areas; (b) the side region and the last layer; (c) the top region; (d) the middle region; (e) the bottom region; (f) higher magnification of the last layer in Fig. 3a; (g) higher magnification of the middle part in Fig. 3d; (h) higher magnification of the bottom part in Fig. 3e.

Fig. 4. Microstructure morphology in cross-section (x-z plane) of as-fabricated part: (a) the bottom region; (b) the middle region; (c) the top region.
Fig. 5. Microstructure morphology in cross-section (x-y plane) of as-fabricated part: (a) the bottom region; (b) the middle region; (c) the top region.

Fig. 6. Illustration of primary dendrite arm spacing (PDAS): (a) illustration of dendritic structure; (b) PDAS distribution frequency in different regions.

Fig. 7. High magnification OM image showing the microstructure of precipitates in different regions: (a) distribution of precipitates in the bottom region; (b) distribution of precipitates in the middle region; (c) distribution of precipitates in the top region.
higher layers along the deposition direction instead of being perpendicular to the substrate, which follows the maximum temperature gradient. However, in this research, the angle is slightly different from observation in the previous study, which was about 70° [17]. This is attributed to differences in wire traveling speed and interlayer temperature control for the different studies. Taking into account the close link between the dendrite microstructure and the development of crystallographic texture, this result indicates that the manipulation of texture is potentially viable during the WAAM of Hastelloy C276, via careful control of processing conditions including the temperature gradients.

The microstructure in the x-y plane is displayed in Fig. 5. A typical equiaxed microstructure is observed in this region and is attributed to sectioning of the epitaxial dendrites, which grow upwards along the deposition direction. This cross-section reveals both the primary dendrites, short secondary dendrite arms perpendicular to each other, and tertiary dendrite arms which are visible on a few of the secondary dendrites. Slight differences observed in the sizes of the primary dendrites can be attributed to different growth velocities, especially at the bottom area, as is shown in Fig. 5.

Fig. 8. (a) Backscattered electron image (BEI) of the deposited sample from the bottom area in the y-z plane, with the spectrum of the matrix (Location A) and precipitates (Location B); (b) the respective chemical composition distribution maps for Ni, Mo, Cr, Fe and W in the bottom, middle and top.

Table 3

<table>
<thead>
<tr>
<th>Area</th>
<th>Ni</th>
<th>Mo</th>
<th>Cr</th>
<th>Fe</th>
<th>W</th>
</tr>
</thead>
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<tr>
<td>Matrix</td>
<td>56.6</td>
<td>17.8</td>
<td>15.8</td>
<td>5.9</td>
<td>4.5</td>
</tr>
<tr>
<td>Top</td>
<td>57.0</td>
<td>17.0</td>
<td>16.3</td>
<td>5.7</td>
<td>4.1</td>
</tr>
<tr>
<td>Middle</td>
<td>58.5</td>
<td>14.0</td>
<td>15.9</td>
<td>7.2</td>
<td>4.5</td>
</tr>
</tbody>
</table>

Table 4

<table>
<thead>
<tr>
<th></th>
<th>Ni</th>
<th>Mo</th>
<th>Cr</th>
<th>Fe</th>
<th>W</th>
</tr>
</thead>
<tbody>
<tr>
<td>Precipitation (Average)</td>
<td>32.6</td>
<td>42.3</td>
<td>14.7</td>
<td>4.0</td>
<td>6.3</td>
</tr>
<tr>
<td>p phase [36]</td>
<td>34</td>
<td>40</td>
<td>16</td>
<td>4</td>
<td>7</td>
</tr>
</tbody>
</table>
3.2.2. Primary dendritic arm spacing

Fig. 6 shows a statistical analysis of primary dendrite arm spacing (PDAS) from different regions of the as-deposited sample. The PDASs are primarily distributed between 20 μm to 40 μm, except for several large size spacings associated with the top and bottom regions. As previously explained, the cooling rate in the first few layers above the cold substrate is quite high, and during the deposition of subsequent layers, the cooling rate is lower, commensurate with the lowered temperature gradient. However, the cooling rate is accelerated again in the top region. The similar thermal behavior is also studied via the heat transfer numerical simulation by Zhao et al. [35]. Complexities associated with the thermal cycling contribute to slight variations in WAAM Hastelloy C276 microstructures.

3.2.3. Precipitation

Fig. 7 shows precipitations found in the as-deposited samples at different locations from bottom to top taken in the y-z plane along the deposition direction. A considerable number of precipitates, of differing sizes, are dispersed at the interdendritic area, many of these aligned with the growth orientation of the primary dendrites. Based on optical microscopy (Fig. 7), it appears that a secondary phase predominantly precipitates at the interdendritic area in the vicinity of secondary dendrites, regardless of the sample location.

In order to study the microstructure at high magnification and identify the element distribution in dendrites, SEM with EDS analysis was performed. As shown in Fig. 8, Mo segregation is evident, indicating Mo is a positive segregation element that tends to be rejected into the interdendritic residual liquid. Elemental analysis in the matrix and precipitations is listed in Tables 3 and 4, respectively. It was found that the precipitates are rich in Mo and W. Due to the insensitivity to the C element as well as its extremely low content, the carbon content cannot be measured properly by EDS [37], and thus EDS results in Tables 3 and 4 do not include the C element. Fig. 9 presents the process of microsegregation during the solidification process predicted by the Scheil-Gulliver model based on the composition content of Hastelloy C276 in this research. In this model, carbon is treated as a fast solute element. Solidification starts from liquid to γ phase, and the α phase and ρ phase are formed last during the solidification process, so it can be seen that the ρ phase is formed during solidification by the WAAM process. Obviously, the high deposition efficiency of WAAM will not expose the heat input for so long period. DuPont et al. also presented that the lower content of Cr in Hastelloy C 276 pushes the solidification further away from the ternary point with γ, α, and ρ phases [38]. Hence, based on the composition of the ρ phase cited by Ma et al. (Table 4) [36], the precipitates with similar elemental composition in this research were induced as ρ phase. The slight variation in the content of the composition might result from some of the tested ρ phases in a relatively small size. The interaction volume of EDS analysis is normally 1–2 μm, so the accuracy of EDS spot analysis on small particles will be compromised by the surrounding matrix.

Fig. 9. Phase transformation illustrated as mole fraction by Scheil-Gulliver model.

![Figure 9](image9.png)

Fig. 10. XRD analysis result for the y-z plane of the deposited sample.

![Figure 10](image10.png)
Note that in Fig. 8 the constitutive elements are not distributed uniformly. Specifically, Mo is more concentrated in the interdendritic area than in the dendritic cores, and a reverse situation is observed for Ni. In addition, compared with the precipitate, the matrix is rich in Fe and deficient in W, from the results given in Tables 3 and 4. This solute segregation in the γ-matrix is considered to be attributed to the significant partition of Mo element between the solid and liquid phases during WAAM as well as the low diffusivity of heavier Mo element than other lighter alloying elements such as Ni [13,38]. During the WAAM process of Hastelloy C267 alloy, primary dendrites first develop in the initial stage of solidification, and then secondary and tertiary dendrites are formed. In the present case, the partition coefficient ($K$) for the Mo element in the solid and liquid γ-matrix phases is believed to be lower than 1 [39], where $K = C_s / C_l$ ($C_s$ and $C_l$ are the Mo solute

![Typical TEM images for the three sections: (a-b) top region; (c-d) middle region; and (e-f) bottom region.](image)

![TEM, SAD and EDS of carbides in the bottom area of the sample: (a-b) TEM images showing carbides at grain boundaries; (c) SAD patterns; and (d) EDS composition of carbides.](image)
concentrations in the solid and liquid phases, respectively). As a consequence, the interdendritic areas are rich in Mo compared with the dendrite cores (Fig. 8b). This also explains the observation that the Mo-rich phase tends to precipitate out in the interdendritic areas (Fig. 7), especially in the vicinity of secondary or tertiary dendrites.

The XRD analysis of different regions from top to bottom in a perpendicular cross-section (y-z plane) to the traveling direction is presented in Fig. 10. It can be confirmed that the matrix is a Ni-γ phase. Secondary phases in the matrix are not detected by XRD due to their low volume fraction. The very high peak reveals an intense columnar dendritic texture growth, and a variation in preferential growth orientation can be observed from bottom to top, which is consistent with the microstructural observations. The diffraction pattern at (311) in the bottom region demonstrates a peak-shift towards higher angles, which is from °90 in the top region to 90.1° in the bottom region. A possible explanation of the observed (311) peak shifts is the chemical changes among the bottom, middle and top regions, i.e., the depletion of Mo occurs together with the segregation of Ni and Fe in the matrix at the bottom, which is supported by the EDS results shown in Table 3. The atomic radii of Ni and Fe are smaller compared with Mo, hence this chemical difference leads to a decrease of the lattice constant and the resultant peak shift towards a higher angle for the bottom region. Furthermore, the top region sample shows a broader (200) peak with the full-width-at-half-maximum (FWHM) of 0.51° compared with that of the middle region sample (FWHM = 0.40°). This may be because of more dislocations existing in the top region. More discussions will be present below in conjunction with the TEM results.

The microstructure of the samples in different areas was further investigated by TEM, as shown in Fig. 11. It is evident that the matrix of the alloy is γ-Ni, and a larger number of dislocations exist in the top region (Fig. 11a and b) than in the other two regions (Fig. 11c-f). During manufacturing, the dislocations can move at high temperatures and

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**Fig. 13.** Hardness comparison of the specimen taken from the y-z plane: (a) Illustration of hardness testing position; (b) Comparison of average hardness in different regions and directions; (c) comparison of horizontal (y-direction) hardness from top to bottom.
form arrays of dislocations, i.e. subgrain boundaries with an example shown in Fig. 11d. This is in agreement with the XRD results. One set of selected area diffraction (SAD) patterns for the two areas in the middle region further confirms that the boundary between the two areas is a subgrain boundary (Fig. 11c), whereas different SAD patterns between the two areas in the bottom region indicate that the boundary is a grain boundary (Fig. 11e). Moreover, it can be observed that from the top to the bottom region, the length and density of dislocations within grains decrease, but the amount of subgrain boundaries seems to increase. In addition, arrays of particles are found at the grain boundaries in the bottom part, which is rich in Mo, as shown in Fig. 12a, whereas the carbide is not found in the middle and top regions in the examined TEM specimens. The particle in Fig. 12b is identified as Ni₂Mo₄C according to the SAD patterns in Fig. 12c. The chemical composition of the particles (Fig. 12d) further confirms that these particles are carbides. From the exploration of the thermal cycle in 3D rapid prototyping by Zhao et al. [35], the bottom region can experience longer heat immersion after solidification, which acts as heat treatment from the upper layers. This is considered due to the reasonably homogeneous dendritic structure. The consistent hardness of WAAM Hastelloy C276 is a positive aspect of the mechanical properties.

3.3. Mechanical properties

3.3.1. Microhardness

Fig. 13 shows the microhardness profile of the specimens in the cross-section part (y-z plane) along the lines at the top, middle, and bottom in the horizontal direction (y) and along the centerline in the vertical direction (z). Fig. 13a shows the test locations, among which the bottom test line is 5 layers away from the substrate to avoid the influence of dilution. The traverse hardness measurements were performed on both as-deposited layers and re-heated layers. Comparing the microhardness distribution (Fig. 13b-c), the values from horizontal (y) and vertical (z) tests display a uniform state and no significant variations of the microhardness values in different regions or directions, fluctuating from 200 to 215 HV₀.₁ in readings. This is considered due to the reasonably homogeneous dendritic structure. The consistent hardness of WAAM Hastelloy C276 is a positive aspect of the mechanical properties.

3.3.2. Tensile strength

The mechanical properties, including ultimate tensile strength (UTS), yield strength (YS), and elongation of as-fabricated samples in the horizontal direction (x) and vertical direction (z) are shown in Fig. 14. As can be seen in Fig. 14a, the average values of UTS and YS for the horizontal samples (469 ± 52 MPa and 287 ± 50 MPa) are larger than those (399 ± 12 MPa and 186 ± 27 MPa) for the vertical samples. Both directions have excellent ductility properties, but also display different values. A deterioration in elongation is present in the x-y plane, being 43 ± 6%, but higher ductility is exhibited in the y-z plane, being 55 ± 3%. This anisotropy in mechanical properties can be attributed to microstructure variation. As is seen in Fig. 3, the sample has intense upward elongated dendrites. Compared to the vertical samples, the horizontal samples have more dendritic boundaries or interdendritic areas in the loading direction and, therefore, more dislocations with the increase of loading can be accommodated within the crack nucleation in this area before a fracture occurs [17].
Fig. 14b-c presents the distribution of the sample tensile properties from top to bottom (Horizontal direction H1-H10) in the x-y plane and three vertical samples (Vertical direction V1-V3) in the y-z plane. Generally, the tensile strength and elongation are relatively uniform, but a slightly decreased tensile strength can be found at the top part, as shown in Fig. 14b. Conversely, ductility exhibits a roughly contrary tendency, as shown in Fig. 14c. The variability in the horizontal direction test is generally consistent with the finding of the PDAS, where the top part possesses more larger-sized PDAS while a refined microstructure exists at the lower part. A fine PDAS induces better mechanical properties to a large extent in this research. More dendritic boundaries possibly resist further dislocation movement, increasing resistance to some extent to plastic deformation, resulting in higher tensile strength. However, due to the significant difference in size and amount of p phase, there is no evidence to lead the connection between p phase and mechanical properties in this research.

3.3.3. Fracture behavior

Fig. 15 presents the fracture surfaces of the as-received tensile samples. It may be observed from Fig. 15a that the tensile specimens were not fully separated when the tensile tests stopped. A large number of dimples scattered throughout the fracture surfaces demonstrate a typical ductile morphology (Fig. 15b-d), evidencing favorable ductility properties for the as-deposited WAAM Hastelloy C276. Anisotropy in mechanical properties may be deduced from the fracture morphology. Comparing the horizontal and vertical samples, the fracture surface for the horizontal samples (Fig. 15b-c) is quite sharp, implying a lower ductility than the vertical ones where there are significant microvoids scattered uniformly throughout (Fig. 15d).

Fig. 15b-c presents the fracture morphology of the samples with the lowest (Fig. 15b, sample H3 at the top part), and the highest (Fig. 15c, sample H8 at lower part) strength values in the horizontal direction. From the images under high magnification, there is no significant difference. However, at lower magnification, it was found that the fracture at failure occurs along the dendrite boundaries, showing an interdendritic fracture morphology in Fig. 15b, while Fig. 15c displays a transgranular fracture. It may be concluded that interdendritic fracture leads to a lower tensile strength than transgranular fracture. The tensile properties are significantly affected by microstructure where the occurrence of large dimples on the fracture surface results in a low tensile strength [40,41]. In this research, a finer microstructure existed in lower regions than the top part, which is in agreement with previous findings, as is shown in Fig. 6.

4. Conclusions

Applying the GT-WAAM process to a Hastelloy C276 component is feasible, enabling accurate fabrication without any visible defects. In the current research, the microstructure and related mechanical properties of the WAAM Hastelloy C276 component are explored and evaluated. Anisotropy is found both in microstructure and mechanical properties. Based on the results obtained, the following conclusions may be drawn:

(1) The microstructure of GT-WAAMed Hastelloy C276 has Ni-γ matrix, dominated with directional elongated columnar dendrites and scattered with intermetallic p phases in the interdendritic area. Ni₂Mo₄C carbide is found only at the bottom regions, which might be due to a longer period of heat accumulation than the upper regions. There are a large number of dislocations which can form subgrain boundaries, and the amount of subgrain boundaries seems to increase from top to bottom.

(2) The variations in microstructure lead to anisotropy in mechanical properties. The horizontal samples (x-y plane) exhibit higher tensile strength but lower ductility than the vertical samples (y-z plane), which is considered due to the elongated dendrites in the matrix.
(3) The well-distributed microhardness in different locations of the as-deposited specimens can be explained by the strong directional dendrites that dominate with microstructural evolution.

Declaration of Competing Interest

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

Acknowledgments

The authors would express sincere acknowledgement to Australian Nuclear Science and Technology Organisation (ANSTO) and the Australian Institute for Innovative Materials (AIIM) center for the use of their electron microscopy.

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