

On the connection of the heat input to the forming quality in wire-and-arc additive manufacturing of stainless steels

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Abstract. WAAM (wire-and-arc additive manufacturing) is recently gaining much interest for the fabrication of parts with wide size. A crucial issue associated to this technology is how to control the process variables to achieve expected qualities of parts. The aim of this investigation is to explore the relationship of the heat input (Q) with the forming quality of 308L stainless steel. The outcomes demonstrate that the heat input has significant effects on the shape, microstructures, and mechanical properties of the walls. Both the height and width of single weld beads increase with an increment in Q . The walls fabricated with the low heat inputs feature better mechanical properties and lower surface roughness. The tensile strengths ($YS_{0.2\%}$ and UTS) and the surface roughness (S_a) of the walls built with the lower heat input (230 J/mm) are 292 ± 4.95 MPa, 523 ± 4.24 MPa, and $106 \mu\text{m}$, respectively, whilst $YS_{0.2\%} = 273 \pm 4.34$, UTS = 482 ± 5.66 MPa, and $S_a = 179 \mu\text{m}$ for the wall fabricated with the higher heat input (576 J/mm). This investigation contributes to enriching the understanding on the WAAM process of steels and helping operators to select a proper heat input for specific applications.

Keywords. WAAM, 308L, heat input, weld bead, microstructures, mechanical property.

1. Introduction

Additive manufacturing (AM) technologies are strongly developed and have become one of the key technologies of Industry 4.0. With the layer-by-layer manufacturing principle, AM technologies enable very complex structures to be produced from different materials [1]. Nowadays, AM technologies (particularly metal AM) are widely applied in various industrial sectors from aeronautics [2] to construction [3–5] and biomedical engineering [6].

The metal AM technologies can be categorized based on the energy source and feedstock form or the fabrication methods. Based on the feedstock form and the fabrication methods, there are two main groups of metal AM - powder bed fusion (PBF) and directed energy deposition (DED) [7]. Among DED technologies, WAAM (wire-and-arc additive manufacturing) uses a metallic wire as the raw material and an electrical arc as the energy source to fabricate metallic parts layer-by-layer [8]. Compared to other metal AM technologies, the WAAM processes feature a high rate of material deposition (about 2-10 kg/h) and low investigating costs of devices and equipment [8–10]. The working space of a WAAM system is also wide and flexible. Therefore, the WAAM process is increasingly adopted to produce large-dimensional components in different industries, such as aerospace, automobile, and shipbuilding [11]. The energy sources used in WAAM processes can be gas-metal-arc welding (GMAW), cold-metal transfer (CMT), plasma-arc welding (PAW), and gas-tungsten-arc welding (GTAW). Compared to GTAW-WAAM and PAW-WAAM processes, the GMAW-WAAM process is less stable. However, it features the highest deposition speed [8,12]. Therefore, the GMAW-WAAM process is the most preferable to be used to produce wide-dimensional components.

The main challenge related to the WAAM processes is how to control the processing parameters to achieve the expected quality of parts [8,13,14]. Many research works were conducted to observe geometrical characteristics (e.g., surface waviness) [15,16] and material

properties of the WAAMed components [15,17,18], while other works focused on the influences of the processing variables on the part quality and trying to establish the process maps [8,19]. Xiong et al. [20] investigated the relationship of process variables on the surface quality of GMAW-WAAMed low-alloy steel walls. They stated that the surface qualities of the as-fabricated wall could be enhanced by augmenting the travel speed. However, the surface waviness increased if the travel speed rise beyond 0.42 m/min. As the wire feed speed/travel speed ratio was fixed, the surface waviness also elevated with a rise in the wire feed speed. Dinovitzer et al. [21] studied the impact of process variables on the geometry of weld beads (WBs) and microstructures of deposited nickel alloy (Hastelloy X alloy). They showed that the current and traveling speed were two major variables that influence the WBs' shape and microstructure characteristics of the as-fabricated material. A rise in the travel speed and/or a reduction in the current could lead to a diminution in the penetration depth. Sarathchandra et al. [22] observed the influence of the input variables on the WBs' geometry of SS304 fabricated by the CMT-WAAM process. They found that the welding speed was an essential factor that affected the homogeneity and continuity of the WBs. Meanwhile, the current mostly influences the penetration depth of WBs, and the standoff distance between the torch and the deposited surface corresponded to the WBs' width. Nemani et al. [23] investigated the influences of interpass temperature on the microstructure of martensitic stainless steel fabricated by the WAAM process. They observed that the microstructures of the as-built material strongly depended on the thermal cycles during the deposition process. Moreover, a higher percentage of residual austenite could be achieved when the higher interpass temperature was applied. Yang et al. [24] stated that the proportion of N_2 in the shielding gas had vital effects on the forming quality of high-nitrogen austenite-stainless steels. A volume percentage of N_2 over 10% enabled achieving a good forming quality. Tawfik et al. [25] explored the influences of the welding velocity on the metallurgical

characteristics of Al-Mg alloys produced by the GTAW-WAAM process. They focused on the variation in the travel speed and found that the microstructure of the as-fabricated sample exhibited fine and equiaxial grains with high travel speeds, compared to the cases of the lower travel speeds. As a result, the sample fabricated with the travel speed of 0.5 m/min featured the most elevated tensile strengths. Tonelli et al. [26] and Le et al. [27] found that the interpass-forced-air cooling method had remarkable effects on both geometrical attributes and material properties of WAAMed components. This method could be used to enhance both the part quality and productivity.

Among the process parameters, the heat input plays an important role [8]. Le et al. [28,29] observed that the forming quality of the walls made of mild steel was significantly affected by the heat input. However, in these studies, the heat input showed less effect on the microstructure evolution. Similarly, Rodrigues et al. [30] also stated that the relationship of the heat input and microstructures of high-strength low-alloy steel was not visible. Meanwhile, the effect of this parameter on the surface waviness was remarkable. The surface waviness of the walls could also be enhanced as the heat input increased. In cases of the WAAM process of austenite-stainless steels [31], raising the heat input could lead to the augmentation in the spacing of secondary dendrites. The same tendency was found in the study of Wen et al. [32], where the influence of heat input on microstructural characterization and corrosion resistance of WAAM SS316L. The authors claimed that a rise in heat input led to the increase in corrosion resistance and primary dendrite spacing in microstructures. Wu et al. [33] showed that the welding-current processes (i.e., speed-cold and speed-arc processes) in WAAM produced different heat inputs and cooling rates, leading to the anisotropy in mechanical characteristics and microstructures of SS 316L. Su et al. [34] investigated the influence of the heat input on metallurgical properties of WAAM thick Al-Mg alloy components. They reported that the grain size could be modified from large columnar to

equiaxed grains by changing the heat input. Aldalur et al. [35] observed the impact of heat input and the working modes (including pulsed-AC, pulsed-GMAW and cold arc) on the porosity and microstructures of GMAW-WAAM single-track multi-layer 5356 aluminum alloy walls. The authors showed that the pulsed-GMAW mode generated the highest heat input, followed by the pulsed-AC and the cold arc modes, respectively. As a result, the cold arc and pulsed-AC mode produced lower penetration than that of the pulsed-GMAW mode. In terms of microstructures and porosity, the pulsed-AC model enabled increasing the average grain size, and significantly reducing the porosity. Nagasai et al. [36] recently examined the influences of the heat input on the metallurgical characteristics of cylindrical parts produced by GMAW-WAAM and CMT-WAAM. They stated that the CMT-WAAM process enabled achieving finer grain size compared to the GMAW-WAAM process. Consequently, the tensile properties of the CMT-WAAMed components were higher than those of the GMAW-WAAMed components. This was due to the CMT-WAAM process generating a heat input that was significantly lower than the one in the GMAW-WAAM process. Veiga et al. [37] compared the shape and metallurgical properties of Invar 36 alloy walls fabricated by PAW-WAAM and GMAW-WAAM processes. The findings in their study could be used to construct the relationship between the heat input and the metallurgical properties of the fabricated materials.

308L is a grade of austenite stainless steels, and it is widely utilized in various industries - e.g., in natural gas, petroleum, and shipbuilding, thanks to its superior corrosion resistance and mechanical characteristics [38]. Previous research in the literature mostly concentrated on analyzing microstructural and mechanical characteristics of deposited 308L stainless steels [38–42]. In the current investigation, the connection of the heat input to the geometrical dimensions of WBs and the quality of single-bead thin walls made of 308L stainless steel was studied. The findings of this work contribute to enriching the

understanding on the WAAM process of steels and helping us to select appropriate heat input levels for specific applications.

2. Materials and experimental procedure

2.1. Materials and the WAAM system

In this research, the 1-mm-diameter 308L-stainless-steel wire was used as the raw material, and the low-carbon steel substrates with dimensions of 250 x 150 x 10 mm were utilized in the GMAW-AM process. The chemical elements of the substrate and wire were given in Table 1.

Table 1. Chemical elements of the substrate and wire materials.

Element	Ni	Cr	Mo	Mn	C	Si	P	S	Fe
Substrate	-	-	-	0.4-0.57	0.11-0.18	0.12-0.17	0.02	0.03	Bal.
wire	9.71	19.84	0.17	1.9	0.04	0.41	≤ 0.03	≤ 0.03	Bal.

Fig. 1a shows the GMAW-WAAM system, which is composed of a GMAW source, a welding torch, a welding wire feeder, a 6-axis robot, and a shielding gas feeder. The relationships between the wire feed speed and welding current in this GMAW-WAAM system is approximately increasing linear [20]. The argon gas with 16 L/min in flow speed was applied during the GMAW-WAAM process.

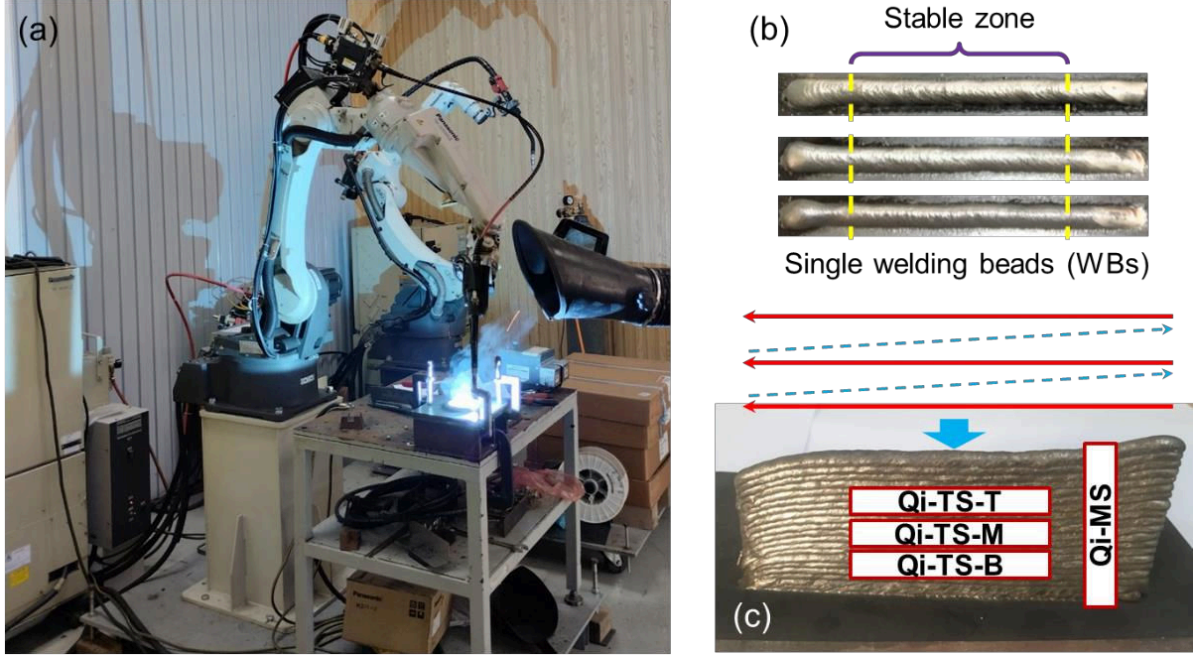


Fig. 1. (a) The real GMAW-WAAM system, (b) welding beads, and (c) the building deposition paths (welding lines in red and rapid travel without welding in blue) and the printed part with the locations for cutting microstructural and tensile specimens.

2.2. Experiments

To achieve the above-mentioned goals of the study, the experiments were carried out through two steps:

Step 1: The single welding beads (WBs) were fabricated with various levels of the heat input Q (Table 2). Herein, the experimental series was designed using the Box-Behnken method [43]. Q was computed by Eq. (1).

$$Q = \epsilon \times \frac{60 \times U \times C}{V} \text{ (J/mm)} \quad (1)$$

where ϵ is the coefficient of the arc source efficiency ($\epsilon = 0.8$ for the GMAW source), U (V) is the voltage, C (A) is the current, and V (mm/min) is the traveling speed.

After the fabrication of WBs, the width (BW) and the height (BH) of weld beads were measured in the stable area of the WBs (Fig. 1b). The BW and BH had been measured by a digital caliper of Mitutoyo. Its precision and resolution are ± 0.02 and 0.01 mm, respectively.

For each WB, the BH and BW were measured at five positions, and the average value was adopted for the analysis, as revealed in Table 2.

Table 2. Scenarios used to investigate the effect of Q on the WB's dimensions.

No	C (A)	U (V)	V (mm/min)	Q (J/mm)	BW (mm)	BH (mm)
1	100	17	400	204.00	3.79	3.01
2	140	17	400	285.60	4.03	3.98
3	100	23	400	276.00	5.08	2.49
4	140	23	400	386.40	5.41	3.46
5	100	20	300	320.00	4.96	3.29
6	140	20	300	448.00	5.42	4.22
7	100	20	500	192.00	4.19	2.02
8	140	20	500	268.80	4.53	3.14
9	120	17	300	326.40	4.31	4.21
10	120	23	300	441.60	6.02	3.66
11	120	17	500	195.84	3.81	3.12
12	120	23	500	264.96	5.21	2.62

Step 2: To assess the influence of Q on geometrical and metallurgical characteristics of the as-produced components, two walls were fabricated with two levels of the heat input – i.e., $Q_1 = 230$ J/mm for the low level and $Q_2 = 576$ J/mm for the high level. The walls were produced using the monodirectional deposition method (Fig. 1c) with twenty layers. A dwell time of 60 s was also utilized between the adjacent layers for the natural cooling. The arc voltage, the welding current, and the dwell time between two adjacent layers were fixed at the same values in both cases.

For evaluating the effects of Q on the wall shape, the walls were scanned by a 3D laser scanner (Range 7, Konica Minolta). The average value of the layers' width, layer height, and wall height were estimated from scanned data with the help of Geomagic-Design-X software [44]. The morphology of the walls' side surfaces was also evaluated and compared.

To assess the impact of the heat input on the metallurgical properties, a metallurgical sample (Qi-MS) and three horizontal tensile testing specimens (Qi-TS-T, Qi-TS-M, and Qi-TS-B) were prepared for each wall (Fig. 1d). A series of operations, including grinding, polishing, and chemical etching was applied for the Qi-MS samples. The etching operation was performed by a mixture of nitric acid, hydrochloric acid, and water with a ratio of 10:20:30 ml, respectively, within 10s.

The microstructure of the as-fabricated material was observed using an optical microscope. The secondary dendrite arm's spacing (SDAS) was estimated from micrographs, using the ImageJ software and the line-intercept approach [45,46]. Moreover, an XRD (X-ray diffraction) machine, a SEM (scanning-electron microscopy) with EDS (energy-dispersive spectroscopy) technique, and an optical emission spectrometer (OES) of Thermo Scientific (ARL3460) were used to analyze the phase and chemical compositions of the as-built material.

The microhardness was tested using a microhardness tester. A 980.7mN load and a 10s idle time were used for each test. The tensile specimens (Qi-TS-T, Qi-TS-M, and Qi-TS-B) were designed respecting to the ASTM E8M-13 standard [47]. The tensile testing was performed at room temperature using a tensile testing machine. The crosshead's velocity was 0.01 m/min. The morphologies of fractured surfaces were captured by the SEM machine.

2.3. Numerical simulations

In this investigation, the finite element (FE) method was adopted to predict the temperature field and the thermal cycle at a point of the walls, which were fabricated with Q_1 and Q_2 (Subsection 2.2). The commercial Ansys LS-DYNA software [48] was used to carry out the numerical simulations. For the WAAM process, the techniques such the element activation/deactivation, the energy source movement, and the models of materials behavior were based on previous studies [49–54]. The Goldak's double ellipsoidal model [55] was utilized to describe the arc source.

The geometry model was designed based on the actual dimensions of the walls and the substrates (e.g., the average wall width, the average layer height, the average wall length, and so on). Each wall was simulated with twenty layers. Linear-brick eight-node elements were applied to generate the mesh. The mesh used for the added layers and the substrate areas close to the wall is finer than other areas of the substrate (Fig. 2). The details of the simulation model can be found in [56]. The thermal cycles were observed at the points (P₁, P₂, and P₃), which are in the middle of the layers L2, L10 and L18, as shown in Fig. 2.

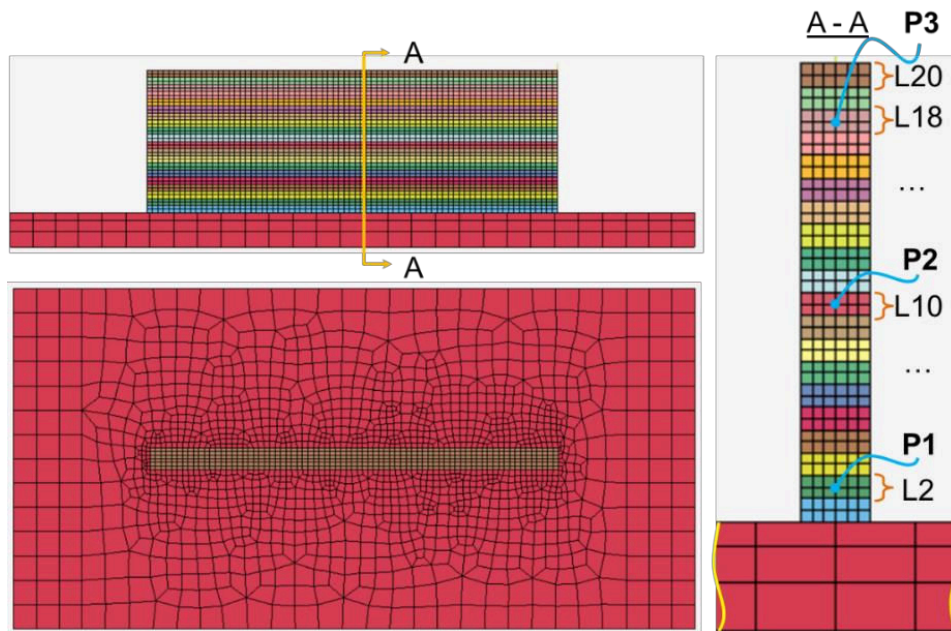


Fig. 2. The FE model used for the case studies in the step 2.

3. Results and discussion

3.1. Correlation between the heat input on the bead geometry

Fig. 3 depicts the relations between the heat input Q and the weld bead dimensions. It is clearly revealed that the bead width and bead height of WBs strongly depend on the heat input Q . In both cases, less dispersion of experimental data compared to the trend line indicates that the replication of experiments is consistent. The bead width BW of weld beads reveals an increasing tendency with an increase in the heat input Q from 192 to 448 J/mm. The bead width reaches the highest value of 6.02 mm at $Q = 441.60$ J/mm, and the minimum bead width of 3.79 mm was achieved at $Q = 204$ J/mm. Indeed, the augmentation in heat input is related to the increment in voltage U or current C , or the decrease in traveling speed V , as shown in Table 2 and Eq. (1). The effect of the current C on the heat input Q is more important than that of voltage U , while the traveling speed V significantly contributes to the heat accumulation and the time of welding [57–59]. Higher values of the heat input generally generate more important volumes of the welding pool [57]. In this case, the melted material was spread on the substrate. As a result, at higher levels of the heat input, the melted material is spread much more on the substrate and larger bead width was formed.

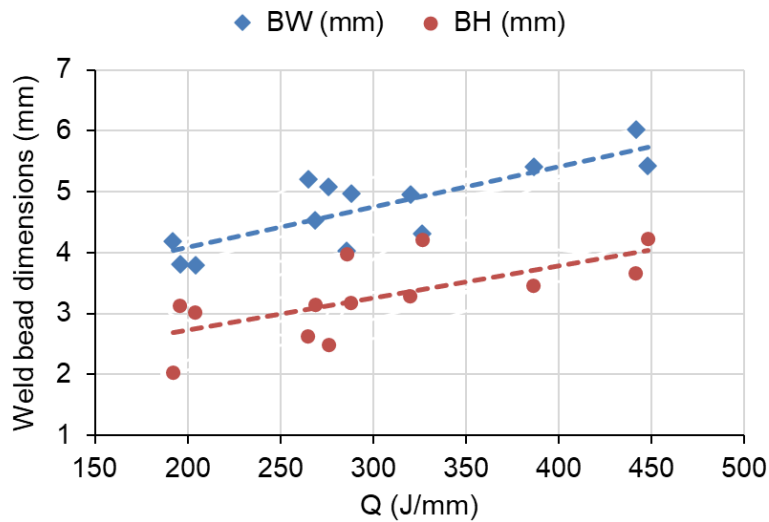


Fig. 3. Relationship between the heat input Q and the WB dimensions (BW and BH).

Similarly, the bead height BH also exhibits a visible tendency to augment with an increase in the heat input Q . The highest value of BH was achieved at $Q = 448$ J/mm and the lowest BH was obtained at $Q = 192$ J/mm. As aforementioned, an increment in the heat input Q involves an increase in the voltage U and the welding current C , or a decrease in the welding speed V , as depicted in Eq. (1). Increasing the heat input Q results in a higher energy to the welding region, leading to higher volumes of melted material and bigger weld pool volumes [57]. Consequently, the bead height BH also shows an increasing tendency with a rise in heat input.

3.2. Effect of the heat input on the wall shape

As shown in Fig. 4, the wall fabricated with a lower heat input features smoother surfaces than the wall built with a higher heat input. The values of S_a , S_q and S_z of the wall surfaces in the case “ $Q_1 = 230$ J/mm” ($S_a = 106$ μ m, $S_q = 131$ μ m, and $S_z = 796$ μ m, Fig. 4a) are significantly lower than those in the case “ $Q_2 = 576$ J/mm” ($S_a = 179$ μ m, $S_q = 226$ μ m, and $S_z = 1434$ μ m, Fig. 4b). This result indicates that decreasing the heat input can enhance the waviness of side surfaces of the as-built walls. This improvement can be explained by the higher levels of heat input that lead to a rise in the interpass temperature. As a result, the wettability of the next layers increases [30,60], thus the surface waviness of the walls is more visible, as presented in Fig. 4b.

Meanwhile, the average layer height and average wall width augment as the heat input increases. The wall fabricated with the low heat input ($Q_1 = 239$ J/mm) features the average value of layer width and height of 5.73 ± 0.18 and 2.04 ± 0.11 mm, respectively, while the wall fabricated with $Q_2 = 576$ J/mm shows an average value of 7.93 ± 0.57 mm for the layer width and an average value of 3.32 ± 0.14 mm for the layer height. As discussed above, the heat input increases, leading to a rise in interpass temperature and the solidification time, and

the wall width will enlarge. The increase in travel speed reduces the volume of material deposited per length and the heat input; thus, reducing the layer height and the stair-stepping effect, as well as the surface waviness [9]. These outcomes emphasized an important effect of heat input Q in the stability and solidification of the melting pool. Higher levels of the heat input can promote an increase in the workpiece temperature. This in turn hampers the dissipation of heat, leading to the difficulty of controlling the height of deposited material.

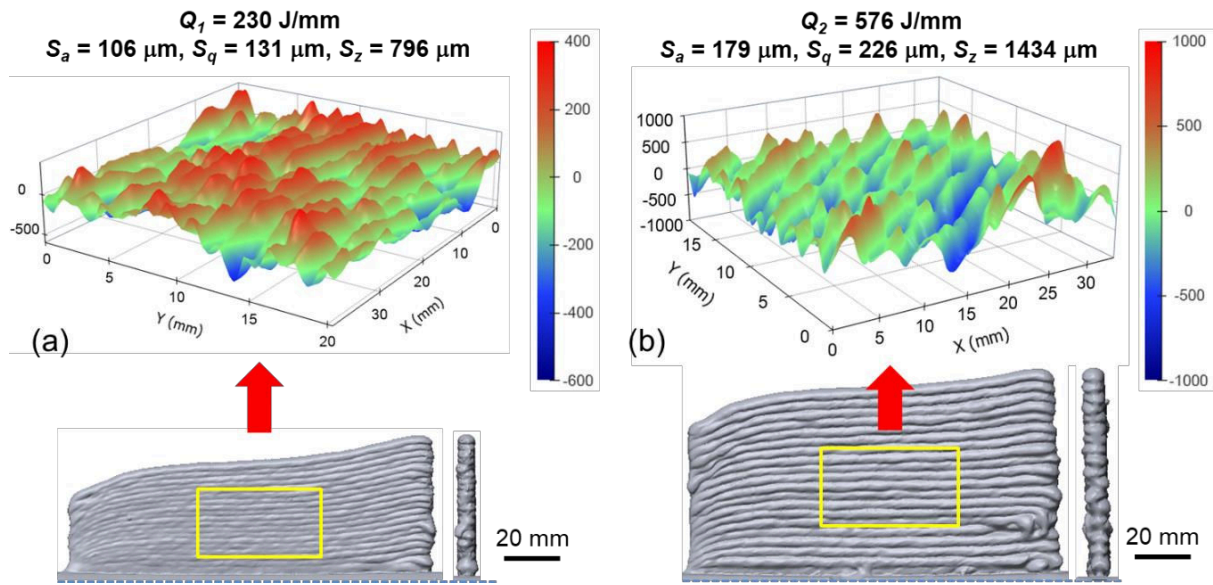


Fig. 4. Surface characteristics of the walls built with (a) $Q_1 = 230$ and (b) $Q_2 = 576$ J/mm.

3.3. Relationship between the heat input and the interpass temperature and cooling rate

Fig. 5 shows the numerical results on the thermal simulation of the WAAM process of the thin walls built with Q_1 and Q_2 . It is first found that the head accumulation occurs in the WAAM process. The interpass temperature and the temperature of the first peak in the thermal cycles increase with the layers' number.

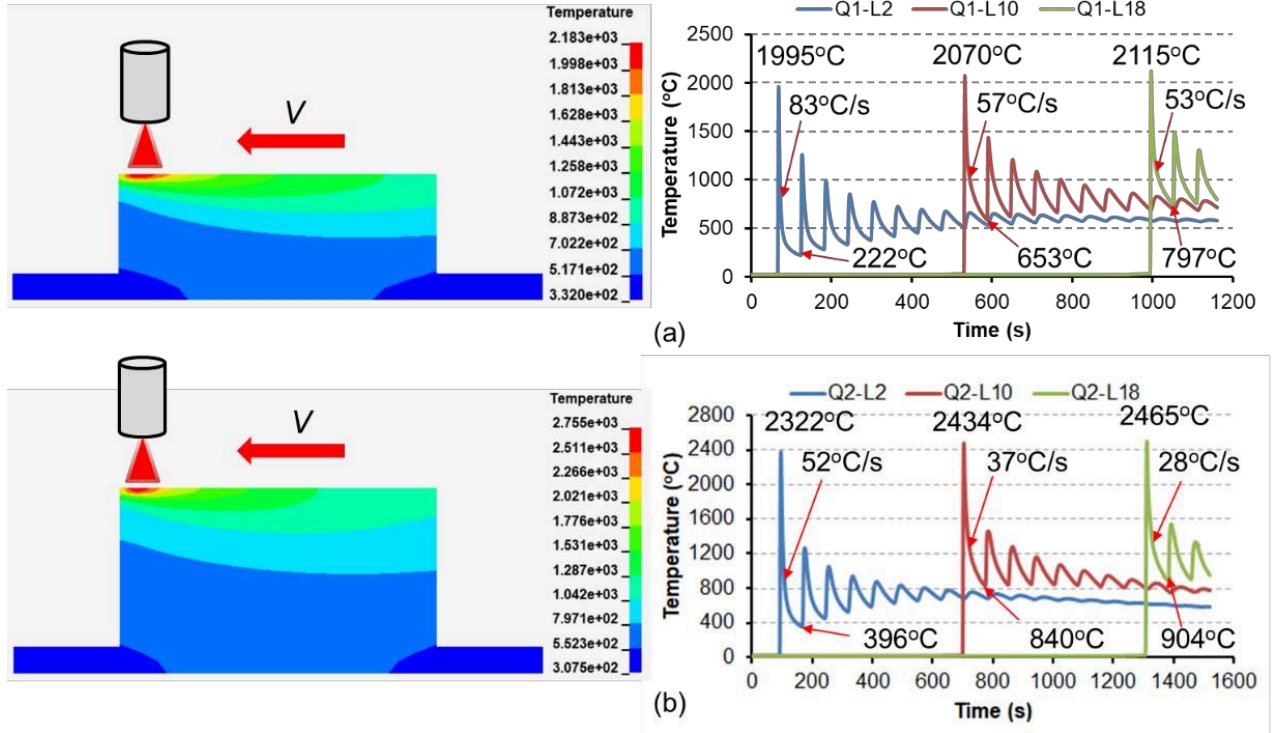


Fig. 5. Thermal field (on the left) and thermal cycles (on the right) of the middle points of the 2nd layer (L2), the 10th layer (L10), and 18th layer (L18) of the wall fabricated with (a) $Q_1 = 230$ J/mm and (b) $Q_2 = 576$ J/mm.

When the $Q_1 = 230$ J/mm was applied, the interpass temperature of the layers L2, L10, and L18 is about 222, 653, and 797 °C, respectively (Fig. 5a). For the case of the wall built with $Q_2 = 576$ J/mm, the interpass temperature also increases along the level of the layers (396, 840, and 904 °C for the layer L2, L10, and L18, respectively, Fig. 5b). A similarly increasing tendency was observed for the temperature of the first peak in the thermal cycles. According to this result, it is proved that with a higher heat input Q applied a higher interpass temperature is accumulated.

The heat accumulation and the augmentation of the interpass temperature in turn cause a reduction in cooling rate. It is shown that the cooling rate decreases along the level of the layers. For instance, the cooling rate at the second, 10th, and 18th layers of the wall fabricated with Q_1 is about 83, 53, and 53 °C/s, respectively (Fig. 5a). The decreasing tendency in the

cooling rate along the level of the layers is also observed in the wall produced with Q_2 (Fig. 5b). However, the cooling rate at a layer of the wall built with Q_2 is remarkably lower compared to that of the wall built with Q_1 .

In summary, the heat input plays a very important role during the WAAM process. The augmentation of the heat input leads to an increment in heat buildup and interpass temperature, thus leading to an augmentation in cooling time and reducing the cooling rate. To maintain the similar interpass temperature for both cases, the idle time between two adjacent layers should be augmented for the wall built with Q_2 . Nevertheless, this method causes an increase in lead times and reducing the productivity.

3.4. Effects of the heat input on microstructures

From the chemical composition analysis by OES (Table 3) and EDX technique (Fig. 6), it is shown that the main chemical elements (e.g., Cr, Ni, Mo, Mn, Si, and Fe) feature a similar percentage in weight as compared to that of the consumable wire (SS 308L) (Table 1). In terms of oxygen and hydrogen elements, in this study, the percentage of oxygen is very low (smaller than 0.01%), whereas the hydrogen element is not detected by either the OES or EDX techniques (Table 3 and Fig. 6). This indicates that the melted metal was well protected by the local torch shielding gas of pure argon. However, Elmer and Gibbs [61] observed that the percentage of oxygen in the WAAMed 308L SS in the air condition reached 0.041% compared to 0.007% of oxygen in the wire. This could be explained by the formation of slag (mostly SiO_2) after removing the welding soot [61]. The slags could contribute to the formation of inclusions in microstructures (Fig. 7a). However, the oxygen segregation to grains or phase boundaries was not found in this study. To effectively prevent the penetration of oxygen from the air ambient, the fabrication of part should be performed under the inert atmosphere condition [61].

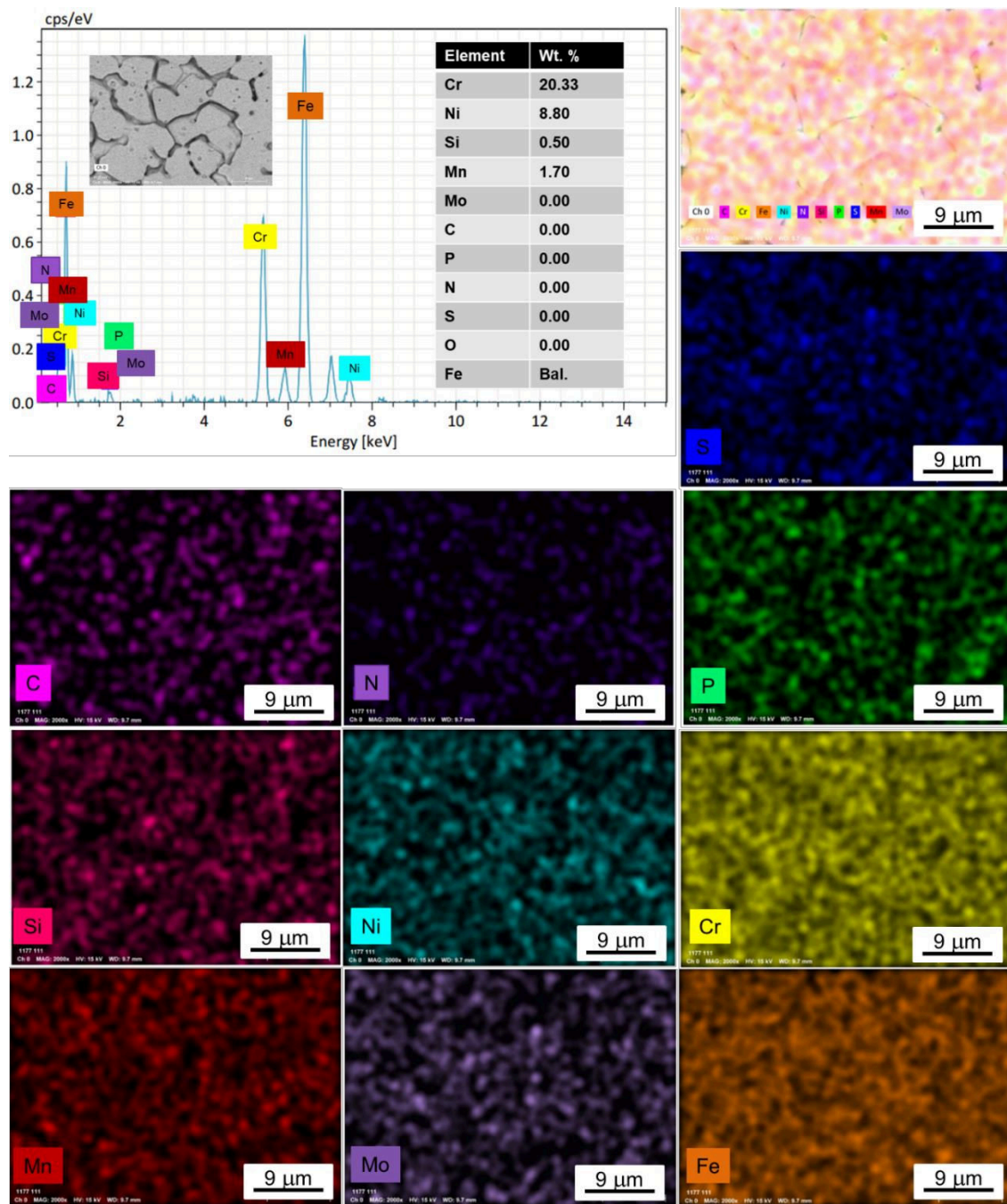


Fig. 6. EDX results and element mapping in the middle layer of the as-built wall.

Table 3. OES analysis results of chemical composition of as-built SS 308L by WAAM.

Material	Units	Cr	Ni	Si	Mn	Mo	Cu	Co
WAAM as-built SS 308L	wt. %	18.67	8.67	0.57	1.74	0.023	0.149	0.207
Material	Units	V	C	N	O	P	S	Fe
WAAM as-built SS 308L	wt. %	0.064	0.013	-	-	0.014	0.008	69.84

Based on the microstructure observation on the entire height of the wall along the building direction, it is found that both walls built with low and high levels of the heat input feature a similar formation and evaluation in microstructures. The microstructures of as-built walls consist of two phases, including dominant austenite phases (white), which present in form of columnar dendrites and equiaxed grains, and residual ferrites (black) with various morphologies, for example, lathy, vermicular, columnar, and globular (Fig. 7b to 7d). Fig. 7a shows a SEM image of microstructure, which indicates austenite and ferrite phases, as well as micro inclusions.

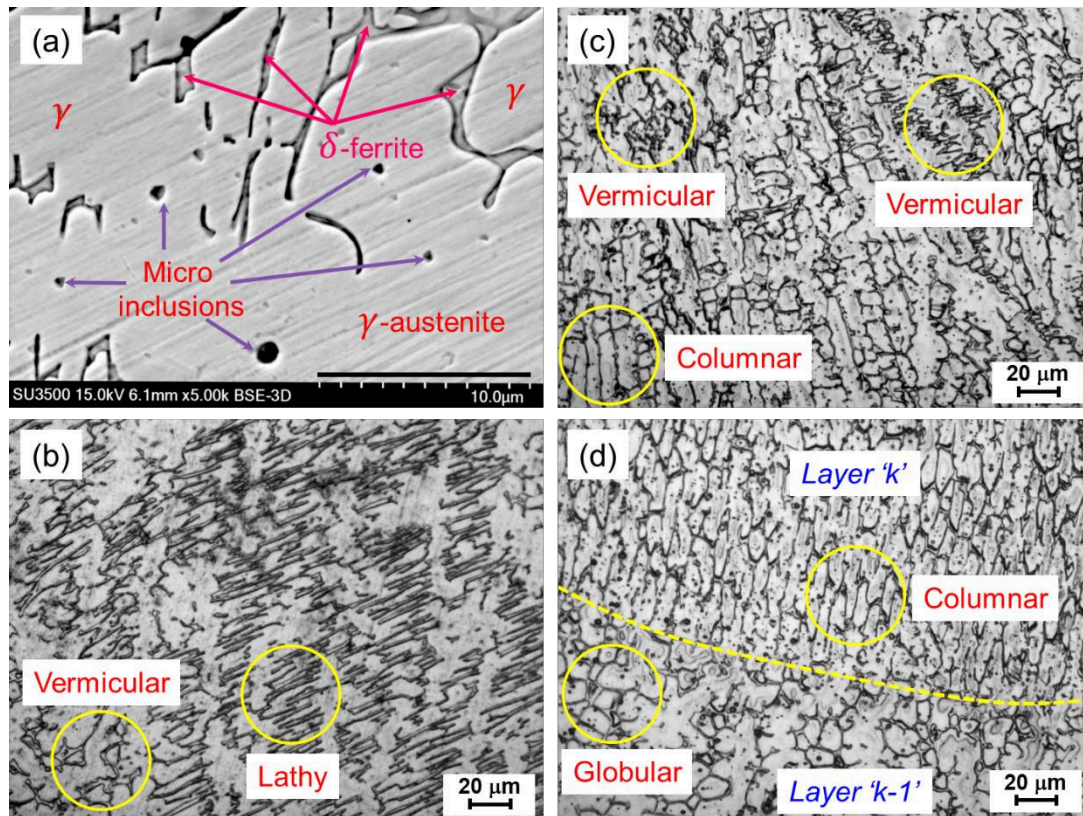


Fig. 7. (a) A SEM image, (b) to (d) OM images of microstructures of WAAMed SS 308L.

Theoretically, the solidification of stainless steels can be occurred respecting to either the A (austenite), AF (austenite-ferrite), or FA (ferrite-austenite), or F (ferrite) mode [62], as shown in Eq. (2) to Eq. (5), respectively.

$$A \text{ mode} : Cr_{eq} / Ni_{eq} < 1.25, L \rightarrow L + \gamma \rightarrow \gamma \quad (2)$$

$$AF \text{ mode} : 1.25 < Cr_{eq} / Ni_{eq} < 1.48, L \rightarrow L + \gamma \rightarrow L + \gamma + \delta \rightarrow \gamma + \delta \quad (3)$$

$$FA \text{ mode} : 1.48 < Cr_{eq} / Ni_{eq} < 1.95, L \rightarrow L + \delta \rightarrow L + \delta + \gamma \rightarrow \delta + \gamma \quad (4)$$

$$F \text{ mode} : 1.95 < Cr_{eq} / Ni_{eq}, L \rightarrow L + \delta \rightarrow \delta \quad (5)$$

where L indicates the liquid phase, δ and γ are the ferrite and austenite phases, respectively. Ni_{eq} and Cr_{eq} are nickel and chromium equivalent that are calculated by the Schaeffler formula [62]:

$$Cr_{eq} = Cr + 0.5Nb + 1.5Si + Mo \quad (6)$$

$$Ni_{eq} = Ni + 0.5Mn + 30C \quad (7)$$

Based on the results of EDX (Fig. 6) and OES analysis (Table 3) on the WAAMed 308L stainless steel sample, the ratio Cr_{eq}/Ni_{eq} is about 1.82. Hence, as shown in Eq. (4), the solidification process of melted 308L stainless steel respects the FA mode. The primary ferrite phases are first secreted from the liquid. Thereafter, the austenite is then formed and gradually grows to replace the ferrite phases according to the solidification time. As the disequilibrium solidification and fast cooling rate in the material deposition, ferrite (δ) phases cannot be transformed into austenite (γ) phases completely. Therefore, at room temperature, there exists a small amount of ferrite remaining in austenite boundaries with different morphologies - e.g., columnar, lathy, vermicular, and globular ferrites. The morphologies of ferrite strongly depend on the solidification and cooling rate. For example, the vermicular morphology is generated at low cooling rate, whereas the lathy is formed at high cooling rate [62–64]. The heat input is directly involved in the cooling and solidification rates; therefore, it will significantly influence the generation and distribution of ferrite morphologies.

The phases of WAAMed 308L stainless steel were also analyzed using the XRD method (Fig. 8). It is shown that the highest peak with the highest integral strength is the

austenite γ (111) peak that is corresponding to austenite matrix. There are also two lower austenite peaks γ (200) and γ (220), indicating columnar austenitic grains, and a small peak δ (110) corresponding to ferrite phases. The same XRD results were presented by Li et al. [40], where the 308L stainless steel was coated by the laser hot-wire cladding process. This again demonstrated that the microstructure of WAAMed 308L stainless steel is dominantly composed of austenite and a little amount of ferrite.

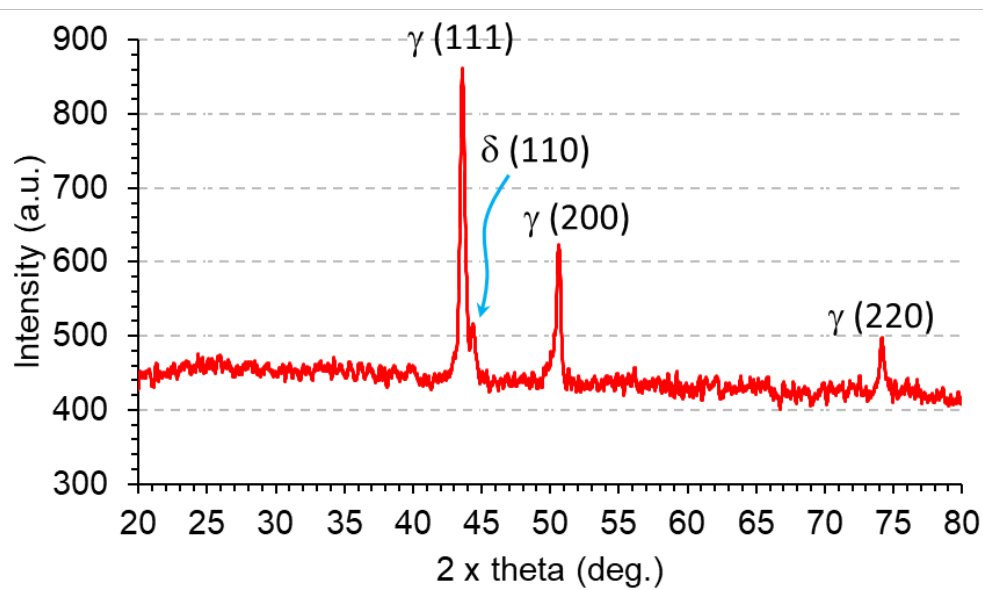


Fig. 8. XRD results analyzed in the middle layers of the as-built wall.

Moreover, the microstructures of the as-built walls can be classified into three regions, namely (i) the top of the wall with free cooling and no heat effects due to the deposition sequence of layers (Fig. 9a), (ii) the middle of the wall consisting of the layers that the fore layer is affected by the thermal cycles of later layers (Fig. 9b), and (iii) the bottom of the wall with a high-cooling rate (Fig. 9c).

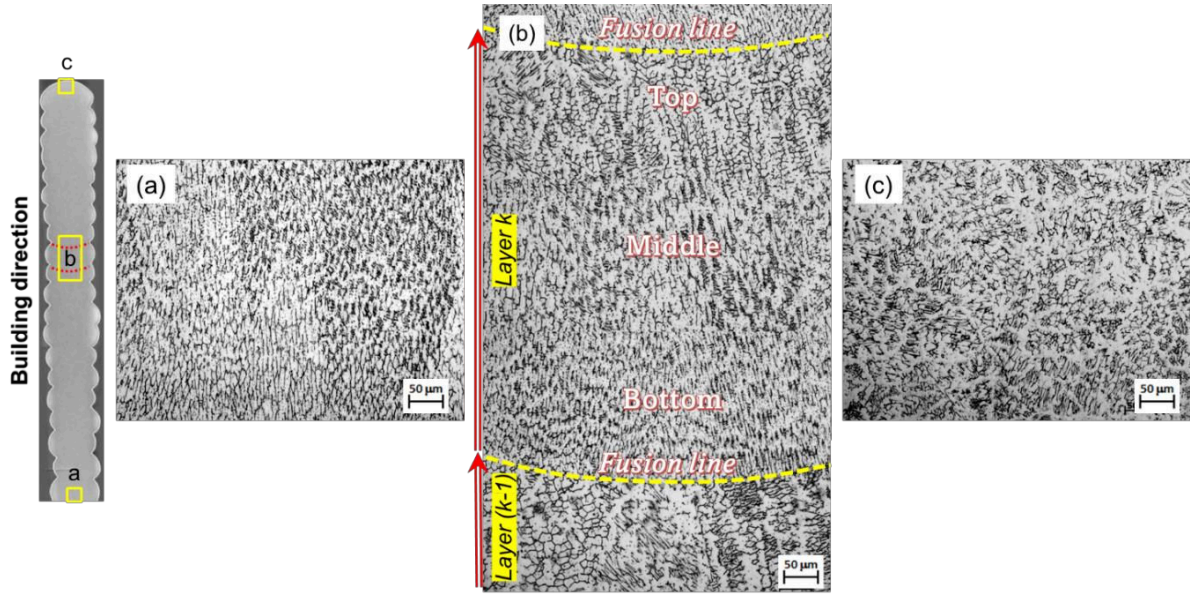


Fig. 9. Microstructures in the three regions of the wall: (a) top, (b) middle, and (c) bottom.

The top layer features granular and equiaxial austenite grains, and ferrite presents in vermicular and lathy forms (Fig. 9c). Indeed, this layer was not affected by the heat of any deposited layer. It was cooled down to room temperature with the nature air, and the cooling rate was low. The temperature gradient in the top layer was also low. Therefore, equiaxed grains were formed, as reported in previous studies [40–42].

The bottom layer, on the other hand, shows thin columnar dendrites, which developed nearly vertically (Fig. 9a). This is the result of the high temperature gradient and high cooling rate that occurred in the first layers. The first layer is built directly on the substrate; thus, the heat is rapidly transmitted. As the temperature gradient decreased, the growing rate of dendrites is accelerated, leading to increasing the columnar dendrites' quantity along the direction of temperature gradient.

The microstructure within a layer in the middle region was shown in Fig. 9b. Fine columnar structures were formed in the bottom of the layer. They were then enlarged progressively. The heat of the layer ' k ' partially melted the top of the layer ' $(k-1)$ ', leading to recrystallization. As a result, the austenite grains close to the fusion line were coarse.

The difference in terms of microstructure between the walls was not visible in the top and bottom of the walls. Nevertheless, the effects of heat input on the grain size can be seen in the middle of the walls. Fig. 10-a-b-c and Fig. 10-d-e-f show representative microstructures in the three regions of the 10th layer of the as-built walls fabricated with Q_1 and Q_2 . It is shown that the secondary-dendrite-arm spacing (SDAS) in microstructure enlarges along the building direction.

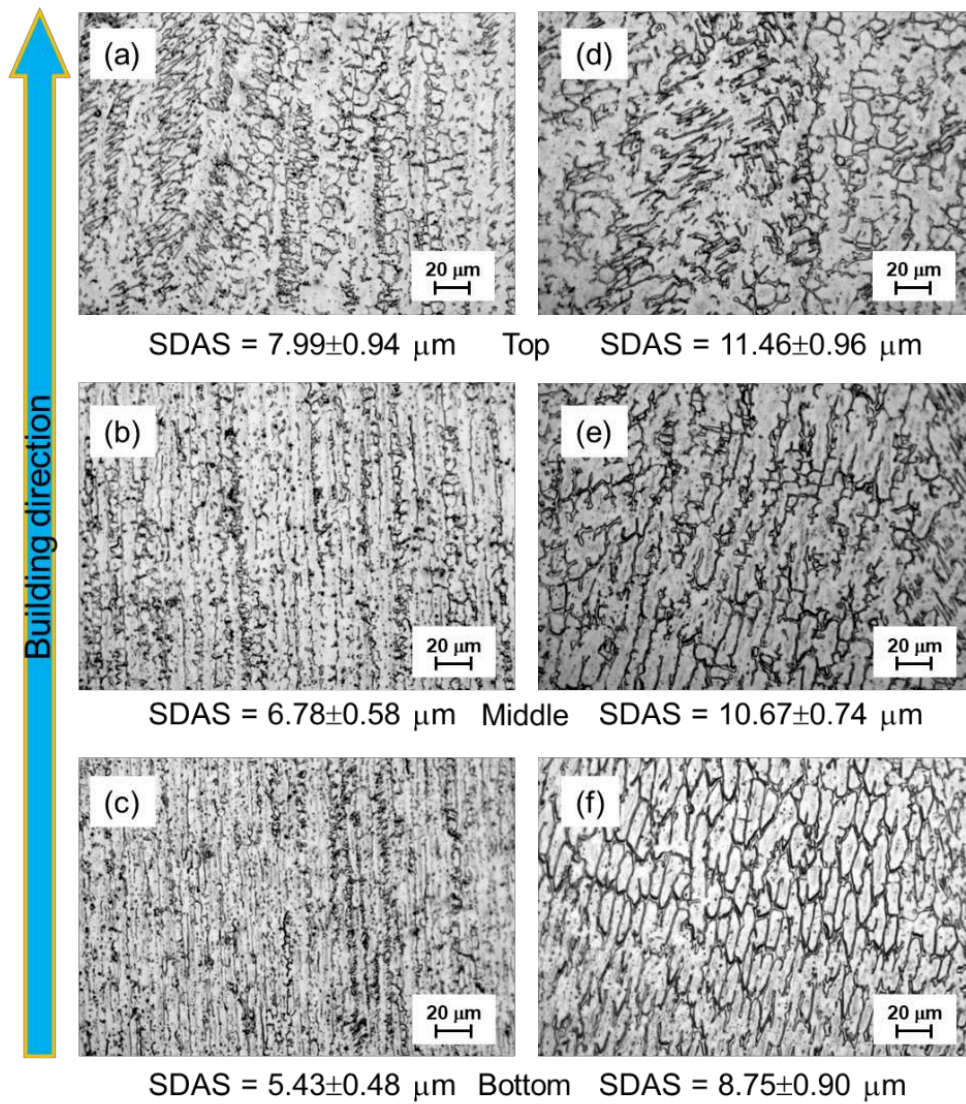


Fig. 10. Microstructures of the 10th layer of the walls: (a, b, c) for the case of “ $Q_1 = 230$ J/mm” and (d, e, f) for “ $Q_2 = 576$ J/mm”.

Comparing the SDAS value in microstructures between two cases of Q_1 and Q_2 , it is noticed that the SDAS increases with an increase in heat input Q . In the case of Q_1 , SDAS =

5.43 ± 0.48, 6.78 ± 0.58, and 7.99 ± 0.94 μm in the bottom, middle, and top of the layer, while these values increase to 8.75 ± 0.90, 10.67 ± 0.74, and 11.46 ± 0.96 μm, respectively, in the case of Q_2 . This phenomenon can be explained by the transformation of vermicular and lathy ferrite into globular and faceted ferrite, respectively, when the heat input increased [61]. As the amount of vermicular ferrite increased, the grain size increased, meanwhile the grain size reduced when the lathy ferrite increased. As shown in Fig. 10-a-b-c, the quantity of lathy ferrite in the sample fabricated with the low heat input (Q_1) is much higher than that in the sample fabricated with the high heat input (Q_2) (Fig. 10-d-e-f). Hence, the grain size in the case of Q_1 is smaller. The obtained result in this study is consistent with previous studies in literature. Wen et al. [32] also stated that the SDAS of the stainless steel specimen fabricated with the higher heat input was larger than that of the one produced with the lower heat input. Similarly, in the work of Wang et al. [65], the speed-pulse WAAM with a higher heat input produced microstructures with larger SDAS than the speed-arc WAAM process, which generated a lower heat input. According to the following relationship, $SDAS = 50 \times \lambda^{-0.4}$, where λ is the cooling rate [65,66], the SDAS increased when the cooling rate decreased. In this investigation, as shown in Fig. 5, the cooling rate decreased when the heat input increased. For example, for the layer L10, the cooling rate in the case of $Q_2 = 576$ J/mm was 37 °C/s vs. 57 °C/s in the case of $Q_2 = 230$ J/mm. Consequently, the SDAS increased with the increase in the heat input.

3.5. Influence of the heat input on hardness and tensile characteristics

As mentioned above, there was insignificant difference in microstructures in the bottom and the top regions between the walls built with low and high heat inputs. Hence, the mechanical properties were analyzed in the middle regions of the walls to clarify the influence of the heat input on the mechanical characteristics.

In terms of hardness, the microhardness (HV0.1) was measured at six positions along the building direction in each layer, L9 to L12. Fig. 11 presents the microhardness results on the cross-section of the walls built with Q_1 and Q_2 . It is shown that the microhardness of the wall built with Q_1 is higher. The average microhardness in the middle layers (L9 to L12) of the wall part built with Q_1 is 166 ± 7.81 HV0.1, whereas that value in the case of Q_2 is 158 ± 5.22 HV0.1. The microhardness measured in each layer of the wall built with the higher heat input (Q_2) was also relatively lower than the microhardness of the wall built with the lower heat input (Q_1).

This observation in microhardness is coherent with the microstructure presented in the previous section. It is well known that the microhardness of the material strongly depends on the chemical elements, phases, and grain size in microstructures. Based on the Hall-Petch relation, as depicted by Eq. (8) [67,68], the microhardness increases with a decrease in the grain size.

$$HV = HV_0 + \frac{K_H}{\sqrt{d}} \quad (8)$$

where HV_0 and K_H are the constants related to the hardness measurement and the tested material, d is the grain size.

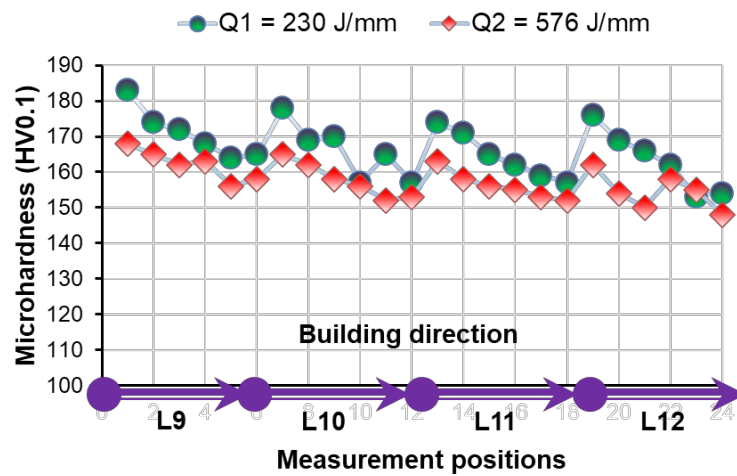


Fig. 11. Microhardness comparison in the middle regions of the walls.

In this study, the SDAS is characterized for the grain size in microstructure of the as-built 308L stainless steel. As revealed in Fig. 10, the SDAS in all regions of a specific layer in the middle location of the wall built with a lower heat input was narrower (Fig. 10-a-b-c vs. Fig. 10-d-e-f). As mentioned above, when a larger heat input was applied, the dissipation of the heat and the cooling rate were slower, resulting in larger SDAS in microstructure. Consequently, the microhardness of the as-built material decreases when the heat input increases. This results in microhardness also agree with those reported in the literature for the WAAM process of stainless steels. Wu et al. [33] demonstrated that the 316L stainless steel wall built with the speed arc welding process, which generated a higher heat input, led to coarser microstructures and lower microhardness compared to the wall built with the speed cold welding process, which generated a lower heat input. Wang et al. [65] also stated that the 316L stainless steel wall fabricated by the WAAM process with the speed-arc mode exhibited higher microhardness than the wall built by the speed-pulse WAAM process. This was due to the speed-arc mode producing less heat input than the speed pulse-mode. However, in the case of high strength low alloy steels, the wall built with lower heat inputs featured relatively lower microhardness [69]. In that case, the different levels in the heat input did not cause significant difference in grain size. The microstructure of the walls built with the high and low levels of heat input exhibited the same microstructures [69].

In terms of the tensile properties, the wall produced with a lower heat input also features higher values of tensile strengths (UTS, $YS_{0.2\%}$) than those of the wall produced with a higher heat input (Fig. 12). The average value of UTS, $YS_{0.2\%}$ of the wall built with Q_1 are 523 ± 4.24 MPa and 292 ± 4.95 MPa, respectively, whilst $UTS = 482 \pm 5.66$ MPa and $YS_{0.2\%} = 273 \pm 4.24$ MPa for the wall built with Q_2 . These observations are also in line with the microstructure differences in terms of grain size. As the SDAS in the middle part of the wall

built with lower heat inputs was narrower, its tensile strengths were higher. This can be explained by the Hall-Petch relationship, Eq. (9).

$$\sigma_{(YS)} = \sigma_{(0)} + \frac{k_{(YS)}}{\sqrt{d}} \quad (9)$$

where $\sigma_{(0)}$ and $k_{(YS)}$ are the constants of the metals, $\sigma_{(YS)}$ presents the yield strength, and d is the grain size.

According to this relationship, the tensile strengths (YS and UTS) become better when the SDAS in microstructures is smaller. As discussed above, the wall built with lower heat inputs featured smaller SDAS in microstructures. Therefore, the tensile strengths (YS and UTS) of the wall built with the lower heat input were higher. Nevertheless, the elongation and toughness showed an opposite tendency. The augmentation in grain size leads to an increase in elongation and toughness [70,71]. As a result, the wall built with higher heat inputs revealed higher elongation ($51 \pm 1.42\%$) compared to that of the wall built with the lower heat input ($42 \pm 2.83\%$). The mechanical characteristics of WAAMed stainless steels (308L and 316L) in the current study and several previous works were summarized in Table 4.

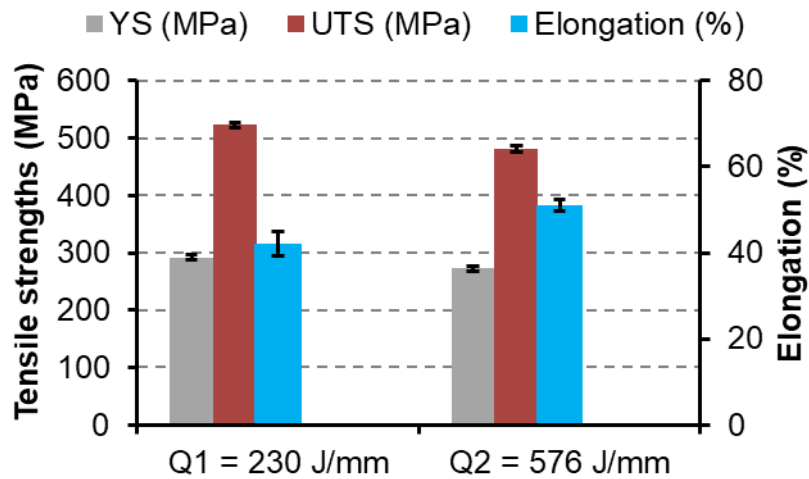


Fig. 12. Comparison of the tensile properties between the walls.

Table 4. Summary of mechanical characteristics of WAAMed stainless steels.

Process	Material	YS (MPa)	UTS (MPa)	El. (%)	Microhardness (HV)	Reference
GMAW-WAAM	308L SS	292 ± 4.95 273 ± 4.24	523 ± 4.24 482 ± 5.66	42 ± 2.83 51 ± 1.42	166 ± 7.81 158 ± 5.22	The current study
CMT-WAAM	308L SS	269	580	68	224	Li et al. [72]
CMT-WAAM	308L SS	368.5	593	50.5	220	Nagasai et al. [73]
GMAW-WAAM	316L SS	345 ± 16.31 355 ± 15.23	579 ± 30.46 607 ± 44.87	-	174.9 177.4	Wu et al. [33]
GMAW-WAAM	316L SS	418.0 417.9	550 ± 6 553 ± 3	-	175 – 200	Wang et al. [65]

In both cases, equiaxed dimples on the fractured surfaces of tensile specimens indicate that the stress was distributed equally over the entire fractured surface during the tensile testing (Fig. 13). The distribution of dimples is also relatively homogenous, indicating a ductile fracture mode, and the as-built material has fine toughness. The dimples of the specimen fabricated with the higher heat input (Fig. 13a) were deeper and greater than those of the wall built with the lower heat input (Fig. 13b). This again proves that the ductility of the wall built with the higher heat input is better. Similarly, Wu et al. [33] found that the 316L stainless steel sample fabricated with the higher heat input featured bigger and deeper dimples on the fractured surface than the one fabricated with the lower heat input. As a result, the sample produced with the higher heat input had a higher elongation, while its UTS and YS were lower. The results in terms of dimples observed on both cases in the current investigation also agree with the results presented in the work of Wang et al. [65], where

dimples with larger dimensions and depth were found on the fractured section of the specimens fabricated with the speed-arc mode (i.e., with the lower heat input).

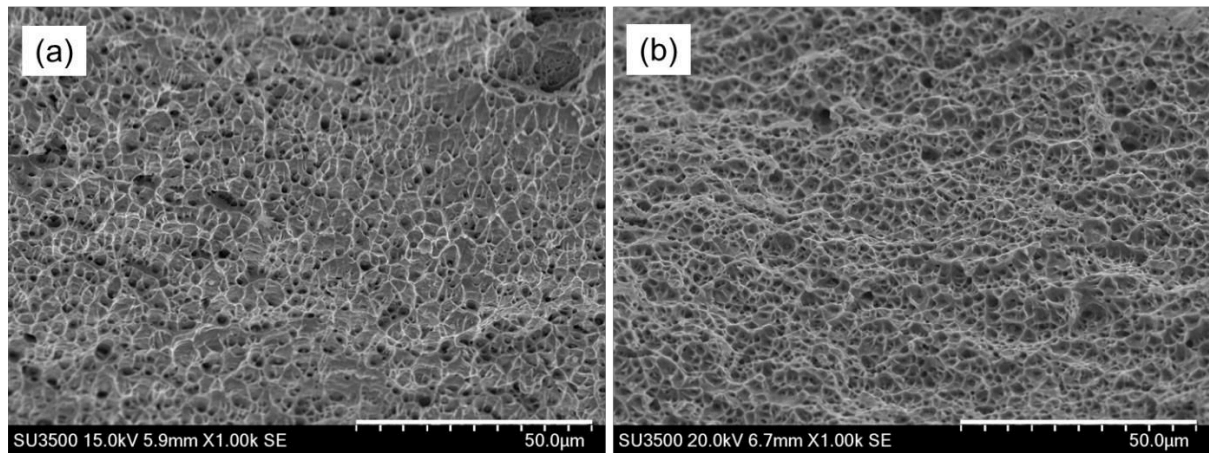


Fig. 13. Fractured surface morphology of the tensile specimen of the wall built with (a) “ $Q_2 = 576 \text{ J/mm}$ ” and (b) “ $Q_1 = 230 \text{ J/mm}$ ”.

4. Conclusions

This study explored the connection of the heat input in the GMAW-WAAM process to the forming quality of 308L stainless steel. The influences of the heat input on both the WBs quality and the multilayer deposition quality were analyzed based on the experimental data. The main findings of the study are listed below:

- The walls built with the lower heat input reveal smoother surfaces, lower height, and narrower width.
- The microstructure of GMAW-WAAM 308L stainless steel features dominant austenite and small amount of ferrite remaining in austenite boundaries. The microstructures of the wall vary along the building direction and strongly depends on the temperature gradient and cooling rate.
- In the middle part of the walls, the average SDAS of the walls produced with the lower heat input is smaller, resulting in stronger tensile strengths and higher microhardness.

- The surface quality and the mechanical properties of thin walls fabricated by the GMAW-WAAM process can be improved by using the lower heat input. A reduction of about 41% in surface waviness (S_a) and an improvement of about 9% in UTS were obtained when decreasing the heat input from 576 J/mm to 230 J/mm.

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2023-03-01

Elsevier

<https://doi.org/10.1016/j.vacuum.2023.111807>

Le VT, Bui MC, Nguyen TD, et al., (2023) On the connection of the heat input to the forming quality in wire-and-arc additive manufacturing of stainless steels. Vacuum, Volume 209, March 2023, Article number 111807