Effects of processing parameters and heat treatment on thermal conductivity of additively manufactured AlSi10Mg by selective laser melting

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A R T I C L E   I N F O
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Aluminium alloy AlSi10Mg
Thermal conductivity
Heat treatment

A B S T R A C T
This work investigates the thermal conductivity of parts which have been additively manufactured using the aluminium alloy AlSi10Mg by selective laser melting, a laser-based powder bed fusion technique. Thermal conductivity characterisation is of particular importance to thermal engineers wishing to make use of additive manufacturing in next generation thermal management solutions.

A number of processing parameters and scanning strategies were employed to fabricate samples for experimental characterisation. While the porosity of produced parts had a significant impact on thermal conductivity, after an anneal heat treatment post-processing step, thermal conductivity increased by 18–41% without any measurable change in porosity. Even though the parts produced with the “contour-hatch” strategy have higher levels of porosity compared to the “contour-hatch” strategy, it has been found that after the heat treatment step, its thermal conductivity can be increased up to the “contour-hatch” strategy. Analysis of the resulting microstructures using scanning electron microscope and energy-dispersive X-ray showed precipitation and coalescence of Si with increasing heat treatment temperature, with dwell time having a lower impact.

While there is a desire for additively manufactured parts with little to no porosity, it has been shown in this study that it is possible to reduce laser energy density requirements by approximately one order of magnitude and still produce parts with acceptable levels of thermal conductivity which could be used for components that are not subjected to strenuous loading conditions, such as heat sinks.

1. Introduction
The additive manufacturing (AM) laser-based powder bed fusion process of selective laser melting (SLM) has become a subject of intense research over the past two decades. In the review by Frazier [1], it has been described as an important emerging commercial manufacturing technology, with the potential to provide customised parts on demand when and where they are needed. Compared to conventional manufacturing methods for metals, such as casting or machining, AM methods allow for the production of complex shapes with an unconstrained level of design freedom. Ngo et al. [2] discussed additional benefits of the technology whereby designers can incorporate porous or lattice weight saving regions, which is considered to be not straightforward and time-consuming for traditional methods. AM has now found uses in a wide range of industries, particularly in aerospace, automotive and for biomedical applications. In the case of the aerospace industry, Liu et al. [3] outlines how it is now becoming to be of strategic importance as it allows for rapid prototyping, direct manufacture, and repair of components. In the biomedical field, Mullen et al. [4] describes how AM orthopedic parts have been shown to have significant advantages as they can be made to more closely mimic natural bone structures.

Much of the focus in literature has been on the mechanical properties of AM parts and components because as noted by Seifi et al. [5], variation in quality of produced parts can limit their use in high-value or mission critical applications. To address this issue, studies have examined the effects of SLM processing parameters on the mechanical and microstructural properties and the corresponding densification of AM...
conventionally processed material. For the aluminium alloy Al–stainless steel, and found that with the correct combination it was possible to produce samples with mechanical properties, such as ultimate parts. Liverani et al. [6] demonstrated this during manufacture of 316 L stainless steel, and found that with the correct combination it was possible to produce samples with mechanical properties, such as ultimate tensile strength and elongation to failure, greater than the conventionally processed material. For the aluminium alloy Al–12Si, Prashanth et al. [7] found a similar result for this material, where the yield and tensile strengths were significantly higher then the corresponding values for the cast material. They did however note a reduction in the fracture strain of the AM processed material. Aboulkhair et al. [8] investigated a wide range of processing parameters and laser scanning strategies and found it was possible to produce parts with almost no porosity, described by Maskery et al. [9] as the main defect resulting in crack propagation. Read et al. [10] developed a statistical methodology to optimise processing parameters in order to minimise part porosity.

To date, however, comparatively less attention has been given to the thermal properties of AM parts, even though they can potentially offer significant benefits in the area of thermal management technologies. These benefits largely stem from the generation of complex geometries. Fasano et al. [11] demonstrated the heat transfer augmentation of a heat sink as a result of the integration of internal channels. Ventola et al. [12] described how the artificial roughness induced during the manufacture of AM parts leads to significant enhancement of natural convection heat transfer. Dede et al. [13] used an optimisation process to produce an organic-like branched fin networks using AM which improved heat transfer performance compared to traditional designs. Until recently, the main materials widely available for AM have generally had poor thermal properties, such as plastics, titanium alloys and stainless steel; but the advent of processes using aluminium alloys, and even more recently, pure copper by Ikeshoji et al. [14], makes AM more and more attractive for thermal solutions.

Aluminium alloys have generally been more difficult to use for SLM compared to titanium or stainless steel. Louvis et al. [15] lists factors such as their high reflectivity, high transfer of heat away from the melt pool, and formation of oxide layers as particular issues. However, coupled with the benefits of low density, good mechanical properties and high corrosive resistance have lead to intensive research and its now widespread use in AM. The main aluminium alloys used for AM are silicon based, with AlSi10Mg being described by Trevisan et al. [16] as one of the most popular in widespread use. The addition of silicon helps to improve fluidity [17], reduce shrinkage and lower the melting point [18], whereas magnesium makes it possible to age harden the alloy [17]. In the context of thermal management applications, studies such as Ameli et al. [19] have investigated the feasibility of producing AM aluminium/ammonia heat pipes with AlSi10Mg, where the capability of SLM to produce heat pipes with integrated wicking structures was demonstrated.

The post-processing of parts built using SLM has also been identified as an important area of research and Fiocchi et al. [18] stated that it is as yet unclear if the same types of heat treatments developed for cast materials are useful for those produced by SLM. Indeed, work by Aboulkhair et al. [20] demonstrated that a standard T6 heat treatment of SLM produced AlSi10Mg, which was believed to improve hardness, actually lead to a softening of the material and it was concluded that a new set of heat treatment procedures should be designed and tailored specifically for SLM.

A precise understanding of the thermal properties of AM components produced from AlSi10Mg is necessary for aiding in the design and characterisation of heat transfer devices that wish to exploit the advantages AM has to offer in the field of thermal management which has been the focus of some recent studies in the literature. Kimura et al. [21] presented thermophysical property measurements of AM Al–xSi binary alloys for what they considered their best SLM processing parameters. The Si content was found to play an important role in leading to either increasing or decreasing of properties including tensile strength, thermal conductivity or the input energy density required for densification. Strumza et al. [22] conducted measurements of thermal conductivity, diffusivity, heat capacity and thermal diffusion of AM AlSi10Mg from room temperature up to 550 °C and found differences compared to the conventionally processed materials in some cases. Wu et al. [23] performed experiments with a AM liquid-cooled heat sink and observed a performance boost after an anneal heat treatment. During studies of a micro channel heat sink, Collins et al. [24] concluded that an apparent mismatch between the printed and nominal AM material thermal properties can lead to inaccurate estimations of device performance. A recent study by Sêlo et al. [25] presented results of the thermal conductivity enhancement of AM AlSi10Mg due to different post-processing heat treatments in gyroid lattice structures intended for heat transfer applications.

The aim of this work is to experimentally investigate the effects of the AM processing parameters on the thermal conductivity of manufactured parts with the goal of optimising the manufacturing process in order to achieve heat transfer performance enhancement. Although AlSi10Mg is one of the most widely available commercial alloys, the reported values of its thermal conductivity can vary widely which can lead to some

\[ \begin{align*}
\phi & \quad \text{Porosity (–)} \\
\rho & \quad \text{Electrical resistivity (Ω-m)} \\
\Delta U & \quad \text{Voltage drop (V)} \\
V & \quad \text{Volume (m}^3)\end{align*}\]

<table>
<thead>
<tr>
<th>Nomenclature</th>
<th>Description</th>
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<tr>
<td>(c_p)</td>
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</tr>
<tr>
<td>(D)</td>
<td>Density (kg/m³)</td>
</tr>
<tr>
<td>(d)</td>
<td>Diameter (m)</td>
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<tr>
<td>(E_d)</td>
<td>Energy density (J/mm³)</td>
</tr>
<tr>
<td>(I)</td>
<td>Electrical current (A)</td>
</tr>
<tr>
<td>(\nabla I_G)</td>
<td>Gradient of grey level intensity (–)</td>
</tr>
<tr>
<td>(k)</td>
<td>Thermal conductivity (W/(m·K))</td>
</tr>
<tr>
<td>(L)</td>
<td>Lorenz constant (W·K²)</td>
</tr>
<tr>
<td>(l)</td>
<td>Length (m)</td>
</tr>
<tr>
<td>(m)</td>
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<td>Electrode spacing (m)</td>
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</tr>
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<tr>
<td>(t)</td>
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<tr>
<td>(t_{\text{exp}})</td>
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<td>(\Delta U)</td>
<td>Voltage drop (V)</td>
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<td>(V)</td>
<td>Volume (m³)</td>
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Greek:

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<td>(\phi)</td>
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Subscripts:

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<td>(e)</td>
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</tr>
<tr>
<td>(f)</td>
<td>Fluid</td>
</tr>
<tr>
<td>(g)</td>
<td>Phonon</td>
</tr>
<tr>
<td>(i)</td>
<td>Inside</td>
</tr>
<tr>
<td>(o)</td>
<td>Outside</td>
</tr>
<tr>
<td>(s)</td>
<td>Solid</td>
</tr>
<tr>
<td>0</td>
<td>Material with zero porosity</td>
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uncertainty in the selection of the correct value for thermal analyses or simulations. In addition, the significant impact of porosity on the resulting effective thermal conductivity will be discussed, and recommendations are presented for the correct selection of empirical relationships.

2. Experimental method

2.1. Materials and manufacturing

The samples in this study were manufactured on a Renishaw RenAM 500 M, a metal powder bed fusion additive manufacturing system. It is equipped with a Gaussian beam continuous-wave fibre laser (max. 500 W, 75 μm spot diameter, 1.03 μm wavelength). It operates on a “move-fire” based technique, whereby the laser is held at a point, fires for a fixed exposure time before turning off and moving rapidly a set distance to the next point in sequence. The printing process was performed in an environment which was initially vacuumed to ~960 mbar and then back filled with pure argon, with an oxygen content of approximately 0.1%, to 10 mbar. AlSi10Mg powder, the composition of which is given in Table 1, was used to manufacture the samples in this study. All tube samples were printed with their axial direction aligned to the z-axis of the machine, i.e. vertically.

As the SLM process is controlled by a range of different processing parameters that affect the properties of manufactured samples, a parametric study of these parameters was undertaken. Six different manufacturing configurations, A-F, listed in Table 2, were used to produce the tube samples. The major processing parameters of layer thickness, laser power, point distance, exposure time and hatch spacing were varied, and the widely used energy density parameter was calculated for each configuration. The energy density parameter groups the processing parameters and allows for a comparison between strategies, as detailed by Olakanmi [27] and Yakout et al. [28]. The energy density is given by

\[ E_d = \frac{P_{\text{top}}}{p s t} \]

where \( P \) is the laser power, \( t_{\text{top}} \) is the exposure time, \( p \) is the point distance, \( s_h \) is the hatch spacing and \( t \) is the layer thickness. While the energy density can be used to estimate the amount of energy transferred to the powder bed, Bertoli et al. [29] and Prashanth et al. [30] have noted that it is considered an approximation as it does not take into account fully the complex physics in the melt pool or differences in material properties. Nonetheless, it provides a useful means of comparison between different configurations.

In addition to varying the processing parameters, two different hatch scanning strategies were employed during the manufacturing step. For samples built using configuration A, the commonly-used “contour-hatch” strategy was used. This strategy generally consists of tracing the original CAD geometry contours with the laser, and infilling the areas between the contours with hatch scans. For the remaining configurations (B—F), a novel single-exposure “points” strategy was used. It differs from the previous method in that instead of tracing and infilling, the CAD geometry is transformed into a point cloud and the laser fires a single exposure for a specified period of time at each point in a continuous sequence. The geometry is therefore built up from overlapping areas equal to the width and shape of the melt-pool at each point. This approach, developed by Betatype Ltd., has the benefits of offering more control over the minimum feature size and the reduction of build time by reducing the number of laser head movements required for hatch infilling. A complete description of this scan strategy, as well an illustration, is provided by Ghouse et al. [31,32].

The samples built in this study consisted of tubes of circular cross section with an external diameter of 6 mm and length of 80 mm. Samples were manufactured with different wall thicknesses ranging from 0.2–1.2 mm in order to later examine the feasibility of their use as potential heat pipes or thermosyphons - two-phase heat transport devices used in thermal management applications. Fig. 1 shows an example of one such additively manufactured tube in this study, where grooved wicking structures are included on the internal tube surface. The wicking structure is used to transport the working fluid from one end of the pipe to the other by capillary action, where each groove is 150 μm wide. Because of the requirement for such small feature sizes, the “points” strategy has been primarily implemented in this study as it is described by Ghouse et al. [31] to be more suited for building structures in terms of both minimum feature size, computational cost and build time. Development of such devices by AM with integrated wicking structures is a topic of active interest in industry, with different approaches and strategies being used, with some studies being reported in literature. Ameli et al. [19] demonstrated the production and characterisation of AM aluminium/ammonia heat pipes from AlSi10Mg. A

![Fig. 1. Additively manufactured tube sample (external diameter of 6 mm), showing grooved wicking structures on internal tube wall. Inset: diagram of heat pipe grooved wicking structure section.](image-url)
regular repeating pattern was used to include porous structures along the inner pipe wall. Richard et al. [33] utilised AM to fabricate the evaporator wick in a loop heat pipe which was made bi-porous by varying the AM geometry in specific locations in order to optimise its performance. Furst et al. [34] developed a mechanically pumped fluid loop which contains an evaporator with integrated AM wicking structures. The structures employed could not have been implemented using conventional processing methods. Further information on heat pipes is not presented in this paper as it is not relevant to the current study which focuses on the effect of the SLM and heat treatment processing parameters on the porosity and thermal conductivity of the material.

2.2. Anneal heat treatment

For the post-processing heat treatment step, the parts were placed in a Carbolite GHA 12/450 horizontal tube furnace. This furnace allows for the control of heat treatment temperature, dwell time and the atmosphere inside the furnace tube. In this study, the tube was held under vacuum (absolute pressure of 0.01 mbar) using an Edwards RV5 vacuum pump. Three different temperatures (300 °C, 400 °C, 500 °C) and dwell times (1 h, 2 h, 5 h) were selected based on the stress relieving cycle of 300 °C recommended in Tang and Pistorius [35] for AlSi10Mg. When power was supplied to the furnace heaters, the temperature increased at a rate of 5 °C/min up to the selected heat treatment temperature, at which it was held for the selected dwell time, before the power to the heaters was switched off and the samples were allowed to cool slowly in the furnace back to room temperature before switching off the vacuum pump.

2.3. Thermal conductivity measurements

Heat is conducted through metals by electrons and lattice waves (phonons), with the overall thermal conductivity, \( k \), written as

\[
 k = k_c + k_v
\]

where \( k_c \) and \( k_v \) are the electron and phonon contributions to the overall thermal conductivity respectively. In pure metals the electron contribution is dominant, whereas in impure metals or disordered alloys the electron mean free path is reduced by collisions with impurities, resulting in an increased phonon contribution, as described by Kittel [36]. The electron contribution to thermal conductivity and electrical conductivity are approximately related by the Wiedemann-Franz law, given in Olafsson et al. [37] by

\[
k_e = \frac{LT}{\rho}
\]

where \( L \) is the Lorenz constant, \( T \) is temperature and \( \rho \) is the electrical resistivity. The theoretical value for \( L \) is 2.44 × 10⁻⁸ W K² but extensive measurements in the literature have found that this value can vary slightly between different materials. For aluminium and its alloys, \( L \) has been found in literature to be 2.1 × 10⁻⁸ W K² and the value of \( k_L \) varies between 10.5 [38] and 13.6 [39] W/(m K).

To determine \( k \) from Eq. (2), the electrical resistivity (the reciprocal of electrical conductivity) of the samples were measured experimentally at ambient temperature (\( T = 20 \) °C) using the four-wire technique. This is a well-known method detailed in Northrop [41] which is used to measure very low electrical resistances and removes from the measurement the effect of lead and contact resistances. A Keithley 2450 SourceMeter was used to supply a constant current of 1 A across a sample and a Keithley 2182A Nanovoltmeter was used to measure the voltage drop over a fixed distance. The electrical resistivity can then be determined from

\[
\rho = \frac{\Delta U}{I} \pi (d_e^2 - d_i^2) / 4s
\]

where \( \Delta U \) is the voltage drop, \( I \) is the current, \( d_e \) is the external diameter, \( d_i \) is the internal diameter and \( s \) is the spacing between the voltage sensing probes. All sample geometrical dimensions were measured by a Mitutoyo vernier caliper, and verified by micrometry measurement on the microscope samples, discussed later in Section 2.6.

Thermoelectric voltage effects between sample and measurement probes were eliminated by using the technique of offset-compensation, described by Lipták [42] and the specific procedure outlined in the equipment documentation [43]. This involves measuring the voltage drop across the sample with the source current switched off, and measuring the voltage drop again with the source current switched on. The offset associated with the thermoelectric effect is determined during the first measurement and it was then subtracted from the second total measurement. The average thermoelectric voltage was found to be 3 nV, which was less than 1% of the voltage measurement with the current supply switched on.

2.4. Porosity measurements

The porosity, defined by Fayed and Otten [44] as the fraction of void space compared to the bulk volume, was determined by the density method. From Aghion and Perez [45], this depends on determining the bulk density of the sample and comparing it to the density of the solids in the sample. The bulk volume was determined by accurate measurement of the sample dimensions, as described in Section 2.3. The volume of solid material was determined by measuring the mass of the sample and the knowledge of the true particle density of the material. The porosity, \( \phi \), is therefore defined as

\[
\phi = 1 - (D_b / D_s)
\]

\[
= 1 - (V_b / V_s)
\]

\[
= 1 - m / (D_s \pi (d_e^2 - d_i^2) l)
\]

where \( D_b \) and \( D_s \) are the bulk and solid particle densities respectively, \( V_b \) and \( V_s \) are the bulk and solid volumes respectively, \( m \) is the mass of the sample measured by a Sartorius CPA225D precision electronic balance and \( l \) is length of the sample. The solid particle density of AlSi10Mg given by Thijis et al. [46] is 2680 kg/m³.

As well as the density method, optical measurement of the porosity was also performed on the polished samples prepared for scanning electron microscope (SEM) analysis (see Section 2.6). The areal porosity of each sample was determined from the acquired images using Matlab (version R2019b). The 8-bit greyscale images were imported into the software, and as their histograms took the form of a bimodal distribution (dark areas for pores, light areas for solid material), a Gaussian mixture model (fimgdist) was used to determine the threshold for image segmentation. The areal porosity can therefore be calculated from the ratio of the number of dark pixels compared to the total number of pixels. Where necessary, the mounting material surrounding the sample visible in the images is excluded from this calculation by masking it prior to the segmentation process.

While other methods, such as the Archimedes imbibition and gas expansion, are often used in the literature for the determination of porosity, the presence of both open and closed voids in the samples in this study made its use infeasible. Since this method only has access to open pore networks, for materials with low or medium porosity (\( \phi < 0.3 \)), Zou and Malzbender [47] state that the existence of closed porosity limits the ability of Archimedian porosimetry in the measuring the total porosity.
2.5. Experimental uncertainty

The uncertainty in the experimental measurements was determined from the technique described in Holman [48] which estimates how the uncertainties in the primary experimental measurements propagate through the data analysis. These primary uncertainty values are listed in Table 3 and are included as error bars in the results in Section 3.

A sensitivity analysis for the values of porosity found that they were most sensitive to the uncertainty of \( d_i \), whereby an increase in \( d_o \) by 0.03 mm leads to a decrease in \( \phi \) by 0.010, and an increase in \( d_i \) by 0.03 mm leads to an increase in \( \phi \) by 0.017.

2.6. Microscopic & chemical characterisation

For microscopic characterisation, the as-built and heat-treated tubes were sectioned samples in the radial and axial directions, mounted in resin, and then ground and polished using standard metallographic techniques, detailed in Vander Voort [49], with a Struers LaboPol-5 machine before being gold-coated by an Emitech K550 sputter coater. The microstructure images were acquired using a Hitachi TM-1000 scanning electron microscope (SEM). The chemical composition was achieved by image processing techniques, and presented in Fig. 2.

Micrometry measurement of the polished samples geometry (i.e. \( d_o \) and \( d_i \)) was performed in order to verify the caliper measurements, discussed in Sections 2.3 and 2.4. An optical system with a 6.5× zoom lens attached to an Imaging Source DFK 31 AU03 camera was used to capture images. These images were then imported into Matlab where a script was used to determine the nominal values of \( d_o \) and \( d_i \). This was achieved by image processing techniques, and presented in Fig. 2. Starting from the centre of the sample, the gradient of the grey level intensity, \( \nabla I_o \), along a radial line was calculated. The sample inner wall location is associated with a local maximum peak of \( \nabla I_o \), whereas the outer wall is location is associated by a local minimum peak \( \nabla I_o \).

![Fig. 2. Microscope image of tube sample used for determination of inner and outer diameters from profile of gradient of grey level intensity.](image_url)

### Table 3

<table>
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<th>Variable</th>
<th>Uncertainty</th>
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<tr>
<td>( d, l, s )</td>
<td>0.03 mm</td>
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<tr>
<td>( m )</td>
<td>1 ( \mu )g</td>
</tr>
<tr>
<td>( \Delta U )</td>
<td>((\Delta U \times 0.0001% + 10 , mV \times 0.0001%))</td>
</tr>
<tr>
<td>( I )</td>
<td>((100 , mA \times 0.025% + 15 , \mu A))</td>
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<tr>
<td>( T )</td>
<td>0.1 °C</td>
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### Table 4

<table>
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<tr>
<th>Material</th>
<th>( k )</th>
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<th>Constants for Eqs. (2) &amp; (3)</th>
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<td></td>
<td>( L )</td>
<td>( k_f )</td>
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<tr>
<td>AISI 304</td>
<td>14</td>
<td>2.25×10⁻⁸</td>
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<tr>
<td>Fe</td>
<td>77</td>
<td>2.62×10⁻⁸</td>
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<td>Al 6082</td>
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In order to first verify the efficacy of the experimental thermal conductivity measurement setup outlined in Section 2.3, a number of samples of known thermal conductivity were measured. Six samples, listed in Table 4, were selected which gave a wide range of thermal conductivity values. The samples were all of circular cross section with external diameters between 2 and 10 mm, and were either rods (i.e. solid) or tubes. All verification samples were sourced from Goodfellow (https://www.goodfellow.com).

The results of the experimentally measured values of \( k \) are shown in Fig. 3 plotted against their known values from literature. It can be seen that there is excellent agreement between the measured and literature values, with the maximum difference equal to 6.7%. The presence of two measured values of \( d_i \) in Eq. (4) results in the uncertainty values for OFHC and Ag being greater than those for the other three materials as these two samples were tubes while the remaining samples were rods.
3.2. Porosity measurement verification

A comparison of the measured porosity values are presented in Fig. 4. Shown in this figure, are the values obtained from the density method, $\phi_D$, and analysis of the SEM images, $\phi_{SEM}$, outlined in Sections 2.4 and 2.6. It can be seen that the two techniques report different results, with the values of $\phi_{SEM}$ lower than those of $\phi_D$. There was found to be two main reasons for this difference.

Firstly, the pore structure of the samples reduces the effectiveness of the optical method to estimate the total porosity. In Fig. 4 it can be seen that configurations D–F have the highest values of $\phi$, which is caused by the higher $s_h$ values used during the SLM process. The resulting pore structure in these configurations follows the path taken by the laser, as seen in Fig. 5 (d)-(f) which presents SEM images of radial cross-sections (see also the additional images provided in the supplementary information) For these cases, which utilises the “points” scanning strategy, an “onion”-like layered radial structure is visible, with partially melted powder present between the layers. This structure gradually disappears with increasing $E_d$ as a result of the lower $s_h$ and increased $t_{exp}$ moving from configuration F to B. The structures for configurations D–F can therefore no longer be considered “random”. Dullien [58] states that optical methods of determining the sample porosity from the areal porosity are only applicable if the pore structure is random.

In contrast, the pore structure for the remaining configurations, A-C, exhibit more random-like structures, as seen in Fig. 5(a)-(c), consisting of spherical pores, for cases using both the “contour-hatch” and “points” strategies. The values of $\phi$ for the two measurement techniques are not equal however due to the multi-scale pores present in the samples. Analysis of the SEM images acquired at 250× do not have sufficient resolution to detect pores <10 μm in diameter, which were visible in images acquired at 2000× (see Fig. 11 and additional images provided in the supplementary information). The presence of multi-scale pores is one of the main reasons given by Dullien [58] why porosities obtained by optical methods differ significantly from the results obtained by other methods.

The error bars presented in 4 for $\phi_{SEM}$ were determined from the standard deviation of the porosity from analysis of different images of the same sample configuration. The error bars for $\phi_D$ were calculated using the methodology described in 2.5. The difference between the values of $d_i$ and $d_o$ obtained from the vernier and microscopy
measurements was found to be within the uncertainty value of ±0.03 mm listed in Table 3, leading to the conclusion that the values of porosity will fall within the uncertainty bounds. Based on these verification measurements, it was therefore decided to present the porosity of the samples in the remainder of this study using the values determined from the density method.

3.3. Thermal conductivity and porosity of as-built additively manufactured samples

The thermal conductivity and porosity of the as-built (non-heat treated) AM samples were determined from the procedure outlined in Sections 2.3 and 2.4, and are presented in Fig. 6.

It can be seen that the thermal conductivity decreases with increasing porosity. The samples produced with the contour-hatch strategy (A) have greater thermal conductivity and lower porosity compared to most of the samples produced with the points strategies (B, C, D, E and F). The results of configurations A and C are however, very similar and Table 2 shows that they were built with the same value of energy density. The average uncertainty values for $k$ and $\phi$ are 3.0% and 12.8% respectively.

In order to understand the effects of individual processing parameters, a sensitivity analysis was performed. The use of regression techniques allows the sensitivity ranking to be determined based on the relative magnitude of the resulting regression coefficients. These values give an indication of the amount of influence each parameter has on the result [59]. It is a common technique used during experiments to facilitate understanding, interpretation and implementation [60].

The values in Table 5 show that the processing parameter $P$, which has the largest magnitude, will have the greatest effect on $\phi$, i.e. increasing $P$ has the greatest effect in reducing $\phi$, where an increase or decrease in $\phi$ is indicated by the sign of the coefficient. $t_{exp}$ has a magnitude slightly smaller than $P$, but similarly, increasing its value will result in reducing $\phi$. These effects can be seen in the cases of configurations B and C. Configuration C has the highest value of $P$ compared to the other “points” strategy configurations resulting in the lowest values of $\phi$ for these cases. For configuration B, although it has the highest value of $t_{exp}$ it has the lowest value of $P$, it still however results in lower values of $\phi$ when compared to the remaining configurations. Increasing $p$ has a lesser effect on reducing $\phi$ when compared to the two previously discussed parameters. The large reduction in $t_{exp}$ can be seen to have a detrimental effect on the as-built samples for configurations D–F whereby it results in the large radial pores as seen in Figs. 5(d)–(f) which were discussed earlier in Section 3.2. Finally both $t$ and $s_b$ can be seen to have minimal effect on $\phi$ compared to the other parameters. In general, the results in Fig. 3.3 show that having a higher value of $E_0$ results in lower $\phi$, but the sensitivity analysis highlights that correct manipulation of the individual processing parameters can also aid in achieving the desired output or reduced porosity.

Also plotted in Fig. 6 are some existing relations in the literature for porous and sintered materials:

$$k = \begin{cases} 2 \left( 1 - \frac{4}{3\phi^2} \left( j_t - k_0 \right) \right) k_0 + \frac{3\phi}{2} \left( j_t - k_0 \right) \right)^{0.5} \
\tan^{-1} \left[ \sqrt{\phi} \left( 1 + \sqrt{\phi} \left( j_t - k_0 \right) \right) \right]^{-1} \left( k_f + \frac{3\phi}{2} \left( j_t - k_0 \right) \right) \right] \cdot \sqrt{\phi} \left( 1 + \sqrt{\frac{3\phi}{2} \left( j_t - k_0 \right) \right) \right] \right] \cdot \sqrt{\phi} \left( 1 + 11\phi^2 \right) \end{cases}$$

$\rho$, $p$, and $s_b$ are the density, porosity, and solid fraction, respectively. The large reduction in $t_{exp}$ can be seen to have a detrimental effect on the as-built samples for configurations D–F whereby it results in the large radial pores as seen in Figs. 5(d)–(f) which were discussed earlier in Section 3.2. Finally both $t$ and $s_b$ can be seen to have minimal effect on $\phi$ compared to the other parameters. In general, the results in Fig. 3.3 show that having a higher value of $E_0$ results in lower $\phi$, but the sensitivity analysis highlights that correct manipulation of the individual processing parameters can also aid in achieving the desired output or reduced porosity.

Also plotted in Fig. 6 are some existing relations in the literature for porous and sintered materials:
packed spheres (see Fig. 7). Even if two samples from each configuration were to have the same level of porosity, the thermal properties would be significantly different due to the different pathways for conduction.

During the SLM process, the metal powder particles are melted and fuse together, with small spherical pores generated as a result of gas becoming entrapped during melting and solidification, or irregular shaped pores generated by unmelted powder or insufficient overlap between scan hatches as outlined by Kempen et al. [67]. Eqs. (6)–(8) fit the experimental measurements more closely compared to Eq. (9). The goodness of fit ($R^2$) between the experimental measurements and Eqs. (6)–(8) are 0.92, 0.92 and 0.91 respectively compared to a negative $R^2$ value for Eq. (9). Therefore it can be contended that while AM parts are produced from metal powders, they are better represented as convex porous materials. Furthermore, it is possible to estimate the value of $k_0$ from Eqs. (6)–(8), which was found to be equal to an average value of 143.9 W/(m·K). This is very similar to the value of 146 W/(m·K) reported by Aboulkhair et al. [8] for $\phi = 0$.

As noted previously, samples A (Fig. 5(a)) and C (Fig. 5(b)) which were found to have similar values of $k$ and $\phi$, also have similar pore structures with small, apparently spherical pores of various sizes present. These pores can be attributed to gas entrapment during the SLM process resulting in residual porosity. For “point” configurations, however (Fig. 5(c)), there are clear radial gaps visible that follow the contours of the employed scanning pattern, as well as the presence of partially melted powder. It is clear that the processing parameters in this case resulted in insufficient melting and hence the greater levels of porosity observed in Fig. 6.

3.4. Thermal conductivity and porosity of additively manufactured samples after annual heat treatment

After initial measurement, the as-built AM samples were subsequently heat treated as outlined in Section 2.2 and the measurements for thermal conductivities and porosities were repeated. Results for the relative change in thermal conductivity for the different heat treatment parameters are shown in Figs. 8 and 9, where $k'/k$ is the ratio of thermal conductivity before and after heat treatment. It can be seen that for all sample configurations there is an increase of 18–41% in the value of $k$ after heat treatment, with the “points” strategies showing the largest relative increases. Note that due to a limited number of B and C samples, these parts were only investigated at a heat treatment time of 5 h for the three different temperatures (Fig. 8(c)).

Interestingly however, there is no discernible trend or optimal parameter set visible in the data other than in Fig. 8(a) whereby an increase in temperature results in a corresponding increase in $k$ for a 1 h dwell time. As stated in Section 2.2, the heat treatment parameters were based on the recommended stress relieving cycle of 300 °C and 2 h, but in the case of thermal conductivity, these values also appear to be sufficient for achieving almost maximum improvement in the range examined. Wu et al. [23] showed that a minimum temperature of approximately 200 °C is necessary for some thermal performance enhancement. For the dwell time, in the case of die-cast aluminium alloys, Cingi et al. [68] observed that an increase from 3 to 6 h during anneal heat treatment had no effect on thermal conductivity.

The measurements of porosity after the heat treatment step showed no change compared to the as-built parts. Additionally, examination of the SEM images for the range of SLM and heat treatment processing parameters (see images provided in supplementary information) showed no structural differences between the as-built and heat treated samples. This is true for all configurations, where the small spherical pores are still visible for A-C, and partially melted powder and radial structures remain unchanged in D–F. If re-sintering or structural change did take place, this effect would be expected to cause an increase in $k$ with increasing heat treatment temperature or time. However, this was not observed in Figs. 8 and 9, whereby these increases all resulted in a similar increase in $k$.

The results of $\phi$ and $k$ after heat treatment are plotted in Fig. 6; the only measured change is the increase in $k$ with $\phi$ remaining approximately unchanged. The average uncertainty values for $k$ and $\phi$ are 3.1% and 13.7% respectively. This suggests that the SLM processing parameters control the level of porosity found in AM parts, in line with the results reported by Maskery et al. [9].

As the different heat treatment parameters seemed to have a uniform effect on $k$, it can be seen that Eqs. (6)–(8) fit these data points very well, with the $R^2$ values equal to 0.94, 0.95 and 0.95 respectively. As before, it is possible to determine the value of $k_0$ for the heat treated samples from Eqs. (6)–(8) and was found to be equal to an average value of 191.5 W/(m·K).

All of the measured values of $k$ and $\phi$ are plotted together in Fig. 10, whereby the values of $k$ are normalised by $k_0$, allowing all the data to collapse onto the relations from literature. Eqs. (6)–(9) are also included in Fig. 10, as well as for completeness, a best fit of the experimental data from this work:

$$k = k_0(1 - \phi)^{-1.466} \quad (10)$$

This equation is based on the work described by Torquato [63] for the derivation of effective-medium approximations, where the exponent can vary between 1.5 and 4. Here the exponent is found to be 1.466 ($R^2 = 0.93$), which makes it almost identical to Eq. (7). As experimental measurements of porosity are generally more straightforward compared to those for thermal conductivity, Eq. (10) can be useful to designers or engineers of thermal management equipment who wish to incorporate AM parts in their designs, or as a parameter within numerical simulations to study the influence of part densification on heat transfer performance.
As there was no observed change in porosity but an increase in thermal conductivity after heat treatment, the microstructures and compositions of the AM samples were examined based on the procedure outlined in Section 2.6.

Fig. 11 presents some SEM images of radial cross-sections for configurations A, B and D, where the heat treatment temperature was varied from 300 °C to 500 °C and the dwell time was kept constant at 5 h (additional SEM images in radial and axial cross-sections are provided in the supplementary information of this paper). The microstructures were observed to become coarser with increasing heat treatment temperature. This can be seen most clearly at 500 °C, where lighter precipitation regions are observed. This process takes place to a lesser degree for other temperatures. For the case of samples produced with the “points” configuration, Fig. 11 shows that there is no apparent difference in the microstructures as a result of the two different scanning strategies employed during this study. For example, when comparing the images for configurations A and B. In all cases, the heat treatment temperature leads to a similar change, i.e. increased precipitation with increasing heat treatment temperature.

EDX analysis establishes these regions to be Si, as shown for configuration C in Fig. 12. The as-built sample (Fig. 12 (a)-(c)) shows a structure where the three main elements of the alloy, Al, Si and Mg, are well dispersed throughout the matrix. The average % compositions for all configurations are shown in Table 6 and were found to be in good agreement with the quoted values for the alloy which are listed in Table 1. The presence of oxygen was also detected and in all cases was consistently found to be approximately 5%; this can be due to residual oxygen in the argon environment during the SLM process or thin oxide layers on the powder surface. The EDX detector limit is 0.5% but the values for Mg are included in Table 6 for completeness.

As the percentage compositions remain approximately constant before and after heat treatment, the increasing Si particle size may be attributed to particle coalescence. This is consistent with previous studies in the literature in which it was suggested by Fiocchi et al. [18] that during the SLM process, the Al matrix becomes supersaturated with Si due to rapid cooling, and is then deposited at the cell boundaries. During heat treatment, this excess Si precipitates out.

Fig. 13 shows the effect of heat treatment dwell time for a fixed temperature of 400 °C on the microstructure of 3 different configurations. It can be seen that as the dwell time increases, the precipitation of Si remains similar in all cases, with a slight increase observed at 5 h. Compared to the results in Figs. 11, 13 suggests that the heat treatment temperature has a more significant effect on the microstructure than the heat treatment dwell time. As mentioned earlier in Section 3.4, this verifies the observation reported by Cingi et al. [68] whereby an increase in dwell time during anneal heat treatment from 3 to 6 h had no effect on thermal conductivity of aluminium alloy die-castings.

Concerning thermal conductivity, while there was a difference between the as-built and heat treated samples, there was no significant change due to the different heat treatment parameters, as discussed in Section 3.4, even though there is a corresponding change in microstructure. In the case of aluminium alloy castings, Chen et al. [69] stated that the arrangement of different precipitates within the matrix causes a significant change in thermal conductivity, whereby the lengthening of the mean free path for conductive electron transport through the more conductive Al phase saw a greater increase in the overall material thermal conductivity. Silbernagel et al. [70] reported a similar result for the electrical resistivity of AM AlSi10Mg. However, a heat treatment cycle of 300 °C and 2 h appears to be sufficient to create the minimum level of precipitation necessary to increase this mean free path through the material, and greater precipitation leads to the diminishing returns observed for the thermal conductivity. This minimum heat treatment cycle effectively acts to relieve internal stresses within the material caused by the complex thermal cycling which takes place in the melt pool during the SLM process, as outlined by Vora et al. [71]. A stress relieving cycle is therefore not only very important for obtaining a balanced set of mechanical properties of AM parts as seen by Fiocchi et al. [18], but also for the thermal conductivity, a pointed also noted by Selo et al. [25]. The dwell time of 2 h is chosen here as the values of $k$ were still seen to be increasing for the shorter 1 h dwell time (Fig. 8(a)) and because it also corresponds to the recommended stress relieving
cycle for mechanical properties described by Tang and Pistorius [35].

3.6. Energy density considerations for thermal conductivity

In Fig. 14 the values of energy density for the different SLM configurations, calculated from Eq. (1) and listed in Table 2, are plotted against the measured values of thermal conductivity for the as-built parts. The configurations with the greatest energy densities (A, B, and C) result in the greatest values of $k$ and therefore the lowest values of $\phi$.

While studies such as Olakanmi [27] have already reported the relationships between energy density and porosity, the results in this work show that it is possible to produce AM parts with energy density values almost one order of magnitude lower while suffering a 26% reduction in $k$. If these parts then undertake the recommended stress relieving heat treatment cycle, $k$ can be increased up to and above the as-built configurations which have the highest values of $E_d$. The additional energy consumption associated with the heat treatment step can be estimated from the energy required to raise the material up to the heat treatment temperature. The specific heat capacity $c_p$ for AlSi10Mg is given by Li and Gu [72] as $739 \text{ J/(kg} \cdot \text{K)}$ at $20^\circ \text{C}$ and $922 \text{ J/(kg} \cdot \text{K)}$ at $400^\circ \text{C}$. The energy density associated with the heat treatment can be written as

$$E_{d,HT} = Dc_p\Delta T$$

Using the largest value of $c_p$ and a heat treatment temperature rise from ambient up to $500^\circ \text{C}$, a liberal estimate of the energy density for the heat treatment is found be $1.173 \text{ J/mm}^3$, which is significantly lower than the values of $E_d$ for the SLM process presented in Table 2 and Fig. 14. Using this approach, i.e. printing with a lower $E_d$ and then increasing $k$ with an additional heat treatment step could be very beneficial from an energy savings point-of-view. Furthermore, considering the statement by Thompson et al. [73] that parts fabricated for thermal management purposes are normally subjected to less strenuous load conditions, the negative effects of additional porosity on the mechanical properties, as outlined in Section 1, could potentially be tolerated.

4. Conclusions

In this study, the thermal conductivity of parts additively manufactured by SLM using AlSi10Mg has been experimentally characterised. The understanding of this property is of significant importance to designers and engineers who wish to make use of the benefits that AM can
offer in the areas of thermal management and heat transfer. Samples were fabricated using different processing parameters and scanning pattern configurations. It was found that those produced with the standard “contour-hatch” strategy had the highest values of thermal conductivity compared to the “points” strategy since the latter caused higher levels of porosity. The close relationship between thermal conductivity and porosity was examined; it was found that AM parts are well represented by previous correlations from literature for convex porous materials.

The effects of an anneal heat treatment on the thermal conductivity of the AM samples was investigated. This additional post-processing step resulted in a significant increase in thermal conductivity of 18–41%. The parameters of temperature and dwell time were varied, but within the range studied here, there was found to be no optimal parameter set, with all parameters having a similar effect on thermal conductivity. There was no measurable change in porosity before and after heat treatment, meaning it is the AM processing parameters that control the level of porosity in the parts.

In order to better understand the change in thermal conductivity, the microstructures were imaged using SEM and EDX. This revealed that the alloying elements are well dispersed in the matrix for the as-built configuration, but with increasing heat treatment temperature, the microstructure coarsens and Si particles coalesce into larger groupings. Given the apparent importance of internal residual stress on the thermal conductivity, further studies are planned which intend to quantify these stresses by X-Ray Diffraction analysis.

Even though the parts produced with the “points” strategy have higher levels of porosity, it has been found that after the heat treatment step, the thermal conductivity increases up to that for parts produced with the “contour-hatch” strategy. Furthermore, the energy density for the “points” strategy is approximately one order of magnitude less than that for the “contour-hatch” strategy which can lead to significantly reduced energy usage. Coupled with this is the fact that thermal management components, such as heat sinks, are normally not subjected to strenuous loading conditions and the negative effects of additional

**Fig. 12.** EDX maps for configuration C for the as-built (a)-(c) and heat treated samples at 300 °C (d)-(f), 400 °C (g)-(i) and 500 °C (j)-(l) all with a dwell time of 5 h. The % atomic composition are given in the top right hand corner.

<table>
<thead>
<tr>
<th>Table 6</th>
<th>Average chemical composition (mass %) of configurations from EDX.</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Element</strong></td>
<td><strong>A</strong></td>
</tr>
<tr>
<td>Al</td>
<td>84.49</td>
</tr>
<tr>
<td>Si</td>
<td>11.71</td>
</tr>
<tr>
<td>Mg</td>
<td>0.17</td>
</tr>
<tr>
<td>O</td>
<td>2.51</td>
</tr>
</tbody>
</table>
porosity on the mechanical properties could potentially be tolerated in these cases. As such, it could be possible to manufacture thermal management solutions using low energy density strategies and use heat treatment to overcome the thermal issues.

Data availability

The raw/processed data required to reproduce these findings cannot be shared at this time as the data also forms part of an ongoing study.

Declaration of Competing Interest

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

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Appendix A. Supplementary data

Supplementary data to this article can be found online at https://doi.org/10.1016/j.matchar.2021.110945.


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