Selective Laser Melting of Aluminium and Aluminium Metal Matrix Composites: A Review

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Abstract

Selective Laser Melting (SLM) is gaining in importance as companies begin to exploit its advantages to produce parts that will enable them enter the market sooner, at a lower cost and/or with parts having an increased geometric complexity. Since aluminium is the second most popular engineering material after steel, its use in SLM was inevitable. In this review, we look at the Selective Laser Melting of aluminium and aluminium matrix composites. We explore some of the inherent difficulties in working with aluminium including the presence of a stable oxide layer, high reflectivity and thermal conductivity and poor flowability of the powder. We also review the unique microstructures that are produced during the SLM process and its effect on the mechanical properties. Included in this is the effect heat treatment on the structure and properties. Finally we look at the benefits and problems of producing aluminium metal matrix composites using SLM

1. Selective Laser Melting of Aluminium Alloys

Selective Laser Melting (SLM) is one of the important additive manufacturing processes as it has the ability to produce near full density components, directly from computer models, in a range of important engineering alloys. SLM of titanium and its alloys are ideal target materials as the material is expensive and problematic to process using conventional manufacturing technologies. Titanium is difficult to machine, which coupled to the high cost of the removed material, adds to the component cost. In addition, during the casting of titanium, protection from oxidation is required, increasing manufacturing costs. Therefore, the net shape ability, high material utilisation and minimal machining that characterise Selective Laser Melting (SLM) make it an attractive alternative. In contrast, SLM of aluminium does not have the same drivers. It is relatively low cost, easy to machine and can be die cast into complex shapes. Nonetheless, SLM of aluminium is gaining increasing importance and has been driven largely by the geometric flexibility that the process offers, as well as the enhanced properties resulting from the fine grain size formed during the rapid cooling.

Selective Laser Melting is a complex metallurgical process that requires an understanding of the role of a large amount of interrelated process parameters. Broadly, these parameters can be divided into four key areas that are related to the laser, the scanning strategy, properties of the powder and the temperature of the part bed. These have been well summarised by Aboulkhair [1], which is shown in Figure 1. In terms of optimising the performance of SLM manufactured components, a predominant goal is to achieve near 100% density. Hence, many studies have focused on the optimisation of parameters in order to achieve this trait, in particular the effect of scan speed, hatch spacing and laser power. However, aluminium has several power-related characteristics that have a big impact on the successful processing.

1.1. Powder-related characteristics.

Selective laser melting of aluminium poses several unique challenges in producing high density components. These are mostly characteristics of the powder and include stability of oxide layer, poor flowability, high reflectivity and high thermal conductivity.

1.1.1 Oxide

The thermodynamic stability and passivating nature of the ubiquitous sesquioxide, which covers all aluminium. The oxidation of a metal, M, may be represented by:

$$M + O_2 \rightarrow MO_2 \qquad (1)$$

The free energy of formation, ΔG, of the oxide is given by:

$$\Delta G = -RT \ln K_3 \quad (2)$$

where R is the gas constant, T is the temperature in Kelvin and K_3 is the equilibrium constant given by:

$$K_3 = (PO_2)^{-1} \qquad (3)$$

where PO_2 is the partial pressure of oxygen when reaction (1) is at equilibrium. For aluminium at 600°C, a $PO_2 < 10^{-50}$ atmospheres is required to reduce the oxide[2]. This is not attainable. Hence, it is neither possible to reduce the oxide nor prevent oxide from forming on freshly exposed metal.

It has been shown that significant amounts of oxide are present within the SLM'ed aluminium and tended to form between scan tracks rather than between layers, as shown in Figure 2 [3]. It is also expected that the oxide would form on top of the melt pool, which may retard its spreading and alter the wetting characteristics and result in a porous structure with weak mechanical properties[4]. However provided sufficient energy is imparted, this surface oxide will break up and become encapsulated within the melt pool [3]. However, it may then act as a micro crack in the system and therefore be detrimental to the final mechanical properties of the product [4].

1.1.2 Flow

Aluminium generally has poor flowability, which can impede the deposition of a thin layer of powder that is critical to the SLM process. The poor flow of aluminium powder is related to two characteristics: non-spherical shape and low density. During the atomisation process, the formation of oxide islands on the powder surface acting as pinning points, and constrain the surface tension forces that are attempting to spheroidise the molten particle [5]. The result is that Al tends to have a non-spherical morphology, even when atomised under an inert atmosphere. A comparison between the shape of Al-12Si and Ti-6Al-4V powder is given in Figure 3. This Figure clearly shows that the Al powder is less spherical than the Ti-6Al-4V, which impacts on the powder flowability. The low density of the aluminium particles are also a factor in flowability as the very strong inter-particle cohesion, mainly ascribed to van der Waals forces, dominate the particle weight [6].

1.1.3 Reflectivity and conductivity

Aluminium is highly reflective of the laser energies in the infra-red range. At the typical 1μm wavelength of modern Selective Laser Melting lasers, aluminium will only absorb 7% of the incident laser energy [7]. Although the actual absorption of a powder bed will be higher due to multiple absorption and reflection events, higher than expected laser energies are required in order to overcome the reflectivity. In addition, there will be a significant difference in absorptivity between previously melted (i.e. solid) aluminium and the neighbouring powder. Hence, the temperature gradients are likely to form during the use of overlapping scan tracks, which may lead to balling as a result of temperature gradient driven convective flow (i.e. the Marangoni effect). Interestingly, Si has a very high absorptivity (~70% [7]) and this may be a factor in the success of Al-Si based alloys (which are detailed below). Since Si has very low solubility in solid Al [8], prealloyed Al-Si powder will contain particles of essentially pure Si. Thus it is possible that these particles, especially those near the surface of the powder particle, will readily absorb the laser energy and heat rapidly. Heat transfer in to the surrounding aluminium (within a given prealloyed particle) will then occur, causing melting. For this approach to be successful, rapid heat transfer needs to occur within a given particle (from the Si to the aluminium). The time, τ, required for the diffusion of heat across a particle of diameter, D, is given by.

$$\tau \approx \frac{D^2}{\alpha} \qquad (4)$$

where α is the thermal diffusivity (~1cm^2/s for Al at room temperature and 0.7cm^2/s at 600°C [9]). Typically, the particle size of the powder used in SLM in less than 50μm, and therefore the time required to diffuse across on powder particle is a few tens of microseconds. This is two orders of magnitude less than the interaction time (few milliseconds). In reality, the fine distribution of the Si within the atomised prealloyed powder results in an effective diffusion distance of much less than the particle diameter and therefore the heat transfer may be considered to be instantaneous.

In addition to aluminium's high reflectivity, it has very high thermal conductivity (κ). So, not only does a significant amount of the laser energy get reflected, but there is also rapid heat conduction away from the melt pool into the already formed solid and/or substrate. The net result of this is threefold. Firstly, more laser energy is required than for materials with lower conductivity. Secondly there is a greater difference between the thermal conductivities of the solid and powder possibly causing density differences depending on whether or not the area being built is on solid or powder. Finally, the width of the melt trace will be much larger than for low conductivity metals as heat is conducted outwards, melting the surrounding powder. For example, our own research has shown that under optimum conditions for Al-12Si (κ~100 W/m-K) [10] the melt trace width is ~200μm, while for titanium (κ=6.7 W/m-K) [11] and stainless steel (κ=21.4 W/m-K) [10] it is 100 and 130μm, respectively. Although a large melt trace width can be compensated for via a beam offset, it will affect the size of the smallest feature that can be produced.

1.1.4 Composition

Despite the above difficulties, the Selective Laser Melting of aluminium can be successfully performed. The majority of this work has been done on Al-Si based alloys (eg Al-12Si and AlSi10Mg) [1, 3, 4, 6, 12-27], with a small amount on 6061 [3], as well as Al-Cu and Al-Zn alloys [12]. However, in the latter two systems, only single line scans were performed. The Al-Si system is generally thought to be most suitable for SLM due to its narrow freezing range. Another factor may be the amount of material that undergoes isothermal solidification. During solidification the nucleated particles will grow until they impinge and join onto their neighbours. It is at this point when the strength of the solidifying structure starts to form and the application of stress can results in cracking and distortion. The Al-12Si and AlSi10Mg alloys will undergo isothermal solidification during this critical final solidification stage and therefore should be less prone to cracking and distortion.

By far the most widely studied alloys are the Al-10Si-0.5Mg, which is equivalent to the casting alloy A360 and Al-12Si (casting alloy 413). For these two high Si content alloys, there appears to be very little difference between their processing. The as processed microstructure is extremely fine, consisting of a cellular dendritic α Al with very fine, dendritic Si particles [13, 16, 19, 27]. Also typical of the microstructure are areas of coarser grains, which are located at the boundary of the molten pool. In this region, the heat from subsequent laser scan lines and layers raises the temperature of the material causing a heat affected zone to form. The microstructure of these materials is discussed further below.

1.2. Laser and Scan related parameters

By far the most commonly studied parameters for the SLM of Al is the effect of the laser energy density on the porosity and properties of the parts. The energy density, E is given by

$$E = \frac{P}{v.s.t} \qquad (5)$$

Where *P* is the laser power (W), *v* is the scan speed (mm/s), *s* is the scan spacing (mm), and *t* is the layer thickness (mm). Thus increasing laser power, decreasing scan speed or scan spacing or less commonly layer thickness all cause an increase in energy density. Generally, it is reported that a minimum energy density (or speed below a critical value) is required to produce maximum density [1, 3, 14, 23, 27], as shown in Figure 4. At lower energies (higher speeds), there is significant porosity caused by incomplete melting [1, 23, 26]. Buchbinder [14] used a high power (1000W) SLM machine to process AlSi10Mg powder. They concluded that a faster build speeds, along with high density could be attained using such high power lasers, as shown in Figure 4. Wang [27] similarly found that high density could be attained provided that the energy density was >40J/mm^3 and this was true when processing in Ar, N_2 of He, Figure 4b. In almost all cases, SLM has been performed using a continuous laser at a wave length of 1.06μm. There has been one report [16] where both Al-12Si and AlSi10Mg were processed using pulsed YAG laser. Although parts with a fine microstructure could be produced, the maximum density that could be achieved for Al-12Si and AlSi10Mg was ~94 and 95%, respectively.

In contrast to the effect of laser energy, little work has been performed on investigating the scanning strategy [1, 23, 26]. Read et al [23] used a statistical approach to optimise not only the scan speed and scan spacing, but also the size of the islands in a checkerboard style strategy. A more detailed study on the role the scanning strategy plays in the formation of the porosity was undertaken by Aboulkhair [1]. They compared 6 different scanning strategies and reported that by optimising the scanning speed and hatch spacing only, the maximum density that could be reached was 97.7%. However, by employing more sophisticated strategies (such as pre-sintering), densities >99.5% could be attained. A summary of their results are shown in Figure 5. Thijis [26] used the scanning strategy not to improve the density, but rather to affect the texture. They showed that if the scanning of the laser is performed only in one direction, a strong partial <100> fibre along the build direction forms. If however, the vectors are rotate 90° between layers, a weak <100> cubic texture forms along the build direction. Thus isotropic or anisotropic parts can be produced through appropriate choice of scanning strategy.

1.3. Microstructure, properties and heat treatment.

1.3.1 Microstructure

The Selective Laser Melting process is characterised by rapid heating and cooling. For Al-12Si, the cooling rate has been predicated to be greater than 10^3 K/s [19]. Therefore, the microstructure of parts is extremely fine grained. At a macro scale, the structure consists of overlapping half-moon melt pools, as shown in Figure 6a. Due to the overlap of scanning vectors, the size of the microstructural features are usually slightly smaller than the size of the molten pool that is formed during processing [26]. At a higher magnification, Figure 6b, the microstructure consists very fine 0.5-1μm cellular aluminium surrounded by a network of nano-sized Si particles [1, 16, 19, 20, 22, 24, 27]. Similar microstructures have been reported in melt spun Al-Si ribbons [28] and form due to the very high cooling rate.

The as processed microstructure contains at least two distinct regions: one significantly finer than the other, as shown in Figure 7. Inside the melt pool, fine cellular-dendrites tend to grow towards the centre of the melt pool. At the edges of the melt pool, the material is affected by the heat generated by both the overlapping scan lines and creation of subsequent layers which causes localised coarsening. Even in this coarsen region, the microstructure is extremely fine in comparison to those resulting from castings. As discussed below, this fine microstructure results in enhanced properties.

1.3.2 Properties

Due to the rapid solidification, the cellular aluminium is supersaturated in Si [19, 22]. Li et al [19] used TEM energy-dispersive X-ray spectroscopy (EDX) mapping to determined that the Si concentration in the as processed was ~7 wt%. This far exceeds the maximum equilibrium solubility of ~1.6 wt% [8] and is even greater than what has been reported in melt spinning [29, 30]. Since the equilibrium solubility of Si in solid Al is low, during solidification the advancing solidification front must reject the solute into the liquid. Since solidification occurs very rapidly in SLM, rejection of the Si cannot occur fast enough and therefore it becomes trapped in the Al matrix. The Si atom, being a smaller than Al (Si atomic radius is 111pm against Al's 118pm), induces solid solution hardening when dissolved in the Al [31]. Hence the extended solubility, along with the very fine grain structure, will cause an increase in strength over conventionally processed (eg cast) material. A summary of reported properties are given in Table 1. In general, the strength of the Al-12Si and AlSi10Mg alloys are approaching 400MPa, with the ductility around 3-5%. This is greater than that of cast material, which typically has an as processed strength and ductility of 300MPa and 2-3%, respectively [32]. As a consequence of these enhanced properties, SLM could well be an attractive alternative to casting.

Due to the layer-by-layer manufacturing approach of additive manufacturing, properties are often dependant the testing direction, with those parallel to the build direction usually lower. For aluminium, the reported data in the literature tends to suggest that the only orientation-sensitive property is ductility, which is normally reported to be lowest in the build direction [17, 20]. Although the reasons for this have not been well investigated, Kempen [17] suggested that the low ductility was due to the presence of border pores which form at the start and end of each scan vector. Since when a sample is built along the build direction there are more layers, it follows that there will be more border pores, which act as critical defects and initiate failure at lower strains.

The fatigue properties have received much less attention than static properties. In the most comprehensive study, Brandl et al [13] investigated the effect of platform temperature, build direction and heat treatment on the fatigue strength of AlSi01Mg. This work concluded that the peak hardening process (ie T6) has the biggest effect on the fatigue strength. Although heating the substrate slightly decreased the fatigue strength, it removed any orientation dependence of the properties. By choosing the best combination of parameters, a fatigue limit of ~200MPa (no substrate heating, 0° orientation and T6 heat treatment) could be attained. At the other extreme, a material that had been processed with a substrate temperature of 300°C, 90° orientation and in the as built condition had a fatigue strength of ~100MPa. This work also identified the crack initiation site, which were pores or unmelted particles, and concluded that improved processing was required to increase the density and avoid imperfections in the material. Siddique [24] investigated the effect of the processing condition of both the tensile and fatigue strength of Al-12Si. Using continuously increasing load tests, they also found that heating the base plate decreased the fatigue, while a low temperature (240°C) post heat treatment had very little effect. They reported a 10^7 run out at ~80MPa. This work also found that the crack initiation site were pores, especially those greater than 50μm and located within 250μm of the surface.

1.3.3 Heat Treatment

The extremely fine microstructure produced in the SLM process also responds well to heat treatment, with a range of different properties able to be generated. During elevated temperature annealing, two main events occur. Firstly, there is a rapid decrease in the amount of Si that is trapped in the cellular Al, which causes an increase in the volume fraction of Si [19, 22]. This will decrease the amount of solid solution hardening and therefore the strength. Secondly, the nano-sized Si particles coarsen into spherical particles that are up to ~2μm in size [13, 19, 22]. The heat treatment process has also created a homogenous microstructure, with the melt pools and two-scale structure of the as-proceeded material no longer evident.

It has been shown that in the as processed material there is an orientation relationship between the Al and Si, which can be expresses as $(111)_{Si}||(200)_{Al}$, and is different to that in cast material [33]. This enables the Si phase to grow along the most stable plane with the lowest free energy: the most dense-packed plane $\{111\}_{Si}$ and is the reason that the Si phase grows into a spherical morphology. As a consequence of the heat treatment, the ductility of the alloy can be increased substantially, albeit at the cost of strength. For example, Prashanth et al [22] reported a tensile ductility in Al-12Si of ~14% in samples heat treated for 6h at 450°C, while Li et al [19] heat treated the same alloy at 500°C for 4h and measured 25% ductility.

The author's own unpublished work on the properties of various Al-Si alloys produced using Selective Laser Melting is summarised in and for SLM from author's own unpublished work. and compared to the handbook data for Al-Si casting alloys in Figure 8. There are several key trends that this data revels. Firstly, it is apparent that the properties of SLM material occupy a different (better) area of property space. Secondly, the as processed (F) yield strength decreases with Si content. This is not surprising as lower Si contents will likely mean less excess Si becomes trapped in solution which in turn causes a drop in the solid solution hardening. The drop becomes more pronounced once the Si content is less than 7wt%, which agrees well with the amount of Si measured in solution in Al-12Si [19]. Solution treatment (T4) decreases the strength and increases the ductility. However, the biggest change occurs in the Mg-free alloys where the ductility is ~25%. In the alloys containing Mg, the ductility is lower (~12-17%), while the strengths tend to be higher. It is likely that, due to the presence of Mg these alloys are undergoing a natural ageing process (formation of Mg_2Si precipitates [31]). Thirdly, the as processed ductility is generally low (although higher than equivalent casting alloys) except in the cast of Al-5Si, where it is ~14%. This again ties in well with the measured extended solubility. It is possible that in the Al-5Si alloy, there is only a small amount excess Si in solution, as well as lower amounts of crack-initiating free Si particles. These would combine to enhance the ductility. Finally, the Al-Si-Mg alloys in the peak aged (T6) condition, the material has a very good balance of strength and ductility. Significant strengthening is occurring which indicates that the Mg content is still reasonably high, despite loss of Mg being reported during SLM [1]. On balance, the best as processed properties are achieved with the Al-5Si alloy (good strength combined with high ductility), while the Al-12Si alloy produces the best properties in the T4 condition and the AlSi7Mg has the best peak aged properties. So selection of the alloy should be made based on whether or not heat treatment will occur.

1.4 Summary

Despite the obstacles posed by the high reflectivity and conductivity, oxide layer and poor flowability of the powder, aluminium components with high density, excellent mechanical properties can be produced via Selective Laser Melting. Work in this area has almost exclusively focussed on Al-Si alloys which have been based on conventional casting alloys. As a result of the high cooling rate, the microstructure produced is extremely fine and consists of a supersaturated cellular Al matrix surrounded by nano-size Si particles and therefore mechanical properties exceeding cast material. In addition, simple heat treatments can be used to create unique microstructures, some which possess very high ductility.

2. Selective Laser Melting of Aluminium Matrix Composites

Many modern technologies require materials with combinations of properties that cannot be met by conventional metal alloys, ceramics, and polymeric materials. By combining two or more physically distinct phases, composites can be produced with aggregate properties that are different, and often far superior, from the constituents individually. Aluminium alloys are often used in applications for their low density and corrosion resistance, however are often let down by their low stiffness and relatively poor wear resistance. The automotive, marine and aerospace industries are particularly interested in production of aluminium metal matrix composites (AMMCs), because introduction of a reinforcement phase gives the potential to produce components with a high stiffness-to-weight ratio. Future use of aluminium MMCs within these fields is heavily dependent on the ease of fabrication and material properties obtained, and hence are underpinned by the chemical compatibility between the matrix and reinforcement.

Casting is a commonly used method to produce AMMCs, however preventing segregation [34, 35] of the heavier composite particles and reactions between them and the melt [36] are two of the main issues this process faces. In addition, subsequent machining of the cast part (eg to remove risers and runners) is difficult due to the presence of the hard reinforcing phase and therefore increases costs [37]. Powder metallurgy routes, and in particular press-and-sinter processing, overcome both the segregation issues and the need for significant machining [38]. However, the parts usually suffer from low density and lower mechanical properties. Despite these difficulties, AMMCs have successfully been used as components in automotive, aerospace, opto-mechanical assemblies and thermal management applications [39].

Selective Laser Melting offers the potential to combine the benefits of both casting and powder metallurgy approaches. Since SLM uses powder as the feedstock, the segregation problems are minimised (compared to melt processing). Subsequent melting by the laser facilitates the possibility of high density parts, similar to casting. However, since the goal of the SLM process is to produce net-shaped parts, little, if any, machining is required. Finally, SLM is characterised by rapid heating and cooling and the material remains in the molten state for only a very short time. This should minimise any reactions between the reinforcing phase and the aluminium.

Introducing a secondary particle into the Al matrix by SLM can result in the enhancement of mechanical properties only if the final AMCs microstructure can be optimised by controlling the processing parameters. For example, Gu et al. report the fabrication of AlSi10Mg/TiC nanocomposite parts with an improved microhardness of 188.3 $HV_{0.1}$, a tensile strength of 486 MPa and elongation about 10.9% by selective laser melting (SLM) [40]. This improvement in mechanical properties was attributed to the combined effects from two strengthening mechanisms: grain refinement strengthening and grain boundary strengthening. Both these originated from the novel microstructure that was produced, which consisted of a ring-structured nanoscale TiC reinforcement in AlSi10Mg matrix, as shown in Figure 9 [40]. Further, it was shown that the structure could be tailored through changing the laser energy density. They also found that the TiC reinforcement in the SLM fabricated parts experienced a microstructural change from the standard nanoscale particle morphology (the average size 77-93 nm) to the relatively coarsened submicron structure (the mean particle size 154 nm) with increasing laser energy density [41]. The distribution of the TiC particle also becomes homogenized as the laser energy density increased. These together gave rise to a considerably low coefficient of friction of 0.36, and a reduced wear rate of of the SLM fabricated AlSi10Mg/TiC AMCs [41]. Ghosh et al. successfully fabricated Al4.5Cu3Mg/SiC AMCs with improved wear properties via SLM [42]. They found that the wear resistance decreased with increasing SiC particle size, Figure 10a, which was attributed to the limited plastic deformation of the surface material when large SiC particles tend to be embedded with the matrix alloy. The wear resistance was also shown to improve with increasing content of the SiC as shown in Figure 10b, primarily as a result of the increased hardness. However, this trend stopped for additions ≥ 20% SiC due to the effect of abrasive wear and cracking [42]. In their study, cracks were generated at the interface between the SiC and Al alloy due to the difference in coefficients of thermal expansion and two types of residual stresses formed during SLM, namely thermal stress and contraction stress.

As such, it can be seen that the interfacial microstructure including phase formation between the Al matrix and the reinforcements hold the key to achieving desired properties. Therefore, more work has been focused on investigating and controlling the interfacial microstructure between the Al matrix and the reinforcements. For example, an advanced in-situ AMCs with different additions of Fe_2O_3 into pure Al was successfully fabricated using SLM by Dadbakhsh et al. [43]. The influence of the Fe_2O_3 content and SLM processing parameters on the microstructure and phase formation in the AMCs and the resultant hardness of the components were systematically investigated. Through varying the SLM parameters and the Fe_2O_3 content, the microstructure and the in-situ formed hard phases in the AMCs could be tailored. At higher Fe_2O_3 content an enhanced hardness was obtained, originating from a very fine, well-bonded, homogeneous distribution of hard particles including Al_2Fe, AlFe, Fe_3Al as well as equilibrium $Al_{13}Fe_4$ (Al_3Fe) in the Al matrix [43]. It is also interesting to note that the porosity induced by entrapped gas can also be mitigated by increasing the Fe_2O_3 content [43].

Ocelik et al. investigated the reaction zone at the interface between Al and SiC in the Al/SiC AMCs fabricated using laser melt injection (LMI) , which is similar to SLM regarding the melting process [44]. Randomly-oriented large Al_4C_3 plates (~30 µm in length) were observed embedded in the Al matrix, probably formed from free carbon. In addition, a large amount of coherent small Al_4C_3 plates were also found, which were formed by reaction between the solid SiC and the molten Al during LMI (see Figure 11a). These small Al_4C_3 plates on the surface of the SiC caused particle cracking[44]. In our recent study of SLM of Al12Si with SiC particles, it was found that although high (97+%) density could be attained, both needle shape and spherical Al_4C_3 were observed in the microstructure, Figure 11b. This was accompanied by the loss of SiC. It is also found that the extent of SiC breakdown was closely related to the laser energy density with higher laser density resulting in more break-down [45]. Two possible theories were proposed for the breakdown: reaction theory and melt theory, although no one theory could completely explain the results. Hence it was concluded that both mechanisms were in operation.

3. Summary

This paper has reviewed the Selective Laser Melting of both aluminium and aluminium metal matrix composites. Aluminium suffers from poor flowability, high reflectivity and thermal conductivity as well as the presence of a dense and stable oxide layers. These are all barriers to their successful processing. None the less, SLM of aluminium can be performed with good results. However, the work has largely been limited to the Al-Si base alloys which are copies of conventional casting alloys. Hence there may be an opportunity to undertake alloy design to investigate whether or not alloys with improved processing can be designed. As a result of the high cooling rate, the mechanical properties of the material is generally superior to that from casting, and the alloy responds well to simple heat treatments, with high ductility possible.

Both in-situ and ex-situ aluminium matrix composites have produced using SLM and it appears to be a promising technique to overcome problems with conventional processing routes. Additions of TiC resulted in a material with high hardness and reduced wear rates. However, despite the high cooling rate and short interaction times, break down of SiC appears to be problematic, with the formation of Al_4C_3 needles in the microstructure. In situ reaction of Fe_2O_3 with the aluminium matrix resulted in a material with very high hardness.

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Tables

Table 1. Summary of reported properties for Al-12Si and AlSi10Mg alloys.

Alloy	Condition/ direction	Yield Strength (MPa)	Tensile Strength (MPa)	Strain to failure (%)	Ref
Al-12Si	Optimised	202.2 ± 4.3	369.3 ± 3.4	4.38 ± 0.16	24
	Optimised	260	380	3	22
	Heat treated 450°C for 6h	95	140	15	22
	Ar	223 ± 11	355 ± 8	4.2 ± 0.6	27
	N_2	224 ± 7	368 ± 11	4.8 ± 0.6	27
	He	221 ± 11	242 ± 43	1.5 ± 0.4	27
	As processed	240	360	4	19
	2h @ 500°C	110	190	25	19
AlSi10Mg	X	-	420	-	14
	Z	-	360	-	14
	X	~250	~330	1.2	23
	Z	~240	~320	~1	23
	X	-	391 ± 6	5.5 ± 0.4	17
	Z		396 ± 8	3.47 ± 0.6	17
	X	243 ± 7	330 ± 3	6.2 ± 0.3	20
	Z	231 ± 3	329 ± 2	4.1 ± 0.2	20

Table 2. Selection of properties from various Selective Laser Melted Aluminium alloys (author's unpublished work)

Alloy	Heat treatment	Yield Strength (MPa)	Tensile Strength (MPa)	Strain to failure (%)
Al-12Si	F	223.5	355.1	4.2
	T4	100.6	175.4	25.0
Al-Si10Mg	F	207.8	367.7	4
	T4	119.4	212.3	12
	T6	209.6	269.1	10
AlSi7Mg	F	192.8	320.1	5
	T4	108.7	204.4	16.8
	T6	227.4	273.1	9.7
Al-5Si	F	145.8	255.8	14.2
	T4	66.7	132.9	27.3

F=as processed, T4=4h@500°C, T6=T4+18h@160°C

Figure Captions

Figure 1. Important parameters in Selective Laser Melting (adapted from [1])

Figure 2. SEM micrograph of Selective Laser Melted 6061 after deep etching with NaOH to reveal the oxide (from [3]).

Figure 3. SEM image images of (a) Al-12Si and (b) Ti-6Al-4V powder. The Al powder is clearly less spherical than the Ti-6Al-4V. Author's own images.

Figure 4. The effect of scan speed on the density of (a) AlSi10Mg 12Si at different laser powers (scan spacing 0.15mm and layer thickness 50μm) [14] and (b) Al-12Si under different atmospheres [27]. Both results show that there is a decrease in density once a critical speed has been exceeded.

Figure 5. The influence of scanning strategy on the relative density of AlSi10Mg parts (from [1])

Figure 6. Microstructure of (a) AlSi10Mg [20] and (b) Al-12Si [27] after Selective Laser Melting. In (a) the half-moon shaped melt pools are clearly visible while (b) shows the fine cellular aluminium (light grey phase) surrounded by the Si (dark grey). The insert in (b) a SEM image showing the very fine scale of the Si particles.

Figure 7. SEM image of an AlSi10Mg alloy after etching with Weck's reagent. Between the he two melt pools (mp1 and mp2) is an area of coarser microstructure (mpc) (from [20]).

Figure 8. Strength and ductility plot for Al-12Si produced via casting and SLM. Parts produced from SLM clear have a better strength-ductility relationship. Data for casting from [32] and for SLM from author's own unpublished work.

Figure 9. Continuous ring of nanoscale TiC in an AlSi10Mg matrix (from [40]).

Figure 10. The effect of (a) SiC size and (b) volume percent of SiC on the wear rate of AlSi10Mg/SiC AMC (from [42]).

Figure 11.Dark reaction products visible around SiC particles in an (a) Al-SiC AMC produced using laser melt injection. [44] and (b) Selective Laser Melted Al-12Si [45]