Microstructural evolution and hardness of rapidly solidified hypereutectic Al-Si surface layers by laser remelting

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1. Introduction

The use of high silicon Al-Si alloys in the automobile, aerospace, and transport industries has been constantly increasing due to their excellent mechanical, tribological, and casting properties [1, 2]. However, when the silicon content exceeds 20 wt%, the alloys suffer from a deterioration of other mechanical properties due to the formation of massive and angular silicon particles with sharp edges, which act as crack initiation sites and significantly reduce the fatigue life. Therefore, to meet the growing demands in automotive, transport, aerospace, marine, aeronautical, and missile technology; the size and morphology of the silicon phase must be refined and modified as the finer primary silicon crystals generally result in improved mechanical properties, such as toughness and ductility [3].
Refinement of the eutectic Si can be achieved by controlling the nucleation and growth of the silicon phase mainly by chemical methods and rapid solidification. The chemical modification involves the addition of some modifier elements, in trace levels, such as sodium [4,5], strontium [6,7], and rare earth metals [8,9]. Na and Sr are well accepted as an eutectic Si modifier, which effectively modifies the morphology of eutectic Si from flake-like to fibrous at a level of a few hundred parts per million. Shina et al. [7] have shown that the addition of Sr, as a master alloy, to Al-10.5%Si-2.0%Cu alloy, transformed the morphology of the eutectic Si from coarse acicular shape to a lamellar and a fibrous one, as the amount of Sr increased. Knuutinen et al. [8] have studied the effect of Ba, Ca, Y, and Yb on 356.0 alloys, indicating that all additions promote alteration in the eutectic and growth modes. Ba and Ca led to a good modification of Si into a fine fibrous silicon structure, while Yb and Ca addition and eutectic Si were refined to a smaller size, but still maintained a plate-like morphology. Nogeta et al. [9] have studied the effects of different concentrations of individual additions of rare earth metals (La, Ce, Pr, Nd, Sm, Eu, Gd, Tb, Dy, Ho, Er, Tm, Yb, and Lu) on eutectic modification in Al-10 wt.%Si and showed that all of the rare earth elements caused a depression of the eutectic growth temperature, but only Eu modified the eutectic silicon to a fibrous morphology with different responses of each element.

For the hyper-eutectic Al-Si alloys, where the microstructure consists of coarse, angular and irregular shapes of the primary Si immersed in Al-Si eutectic, the phosphorus (P) proved to be an effective modifier and refiner of primary Si particles but not for the eutectic Si [10]. Wu et al. [11] have reported that the addition of phosphorus to the Al-20 wt.%Si alloy with a controlled temperature and holding time has led to a reduction of the size of the primary silicon to 20 μm due to the formation of AlP particles, which acted as heterogeneous nucleation agent on the solid AlP particles, which promotes the modification and refinement of the primary Si without affecting the eutectic Si morphology. A combination of P with other elements was also investigated to refine the primary and the eutectic Si. Faraji et al. [12] have demonstrated that (P + Sr) was effective to some extent in improving the strength and ductility but the size of the primary silicon crystals did not reach the nanoscale size. Li et al. [13] have demonstrated that the morphology of primary Si was refined from coarse irregular star-like and plate-like shape to fine block-like when the addition contents of Y increased to 0.8%. The average size of primary Si reduced from 89 μm to 33 μm. Xia et al. [14] have reported that the addition of erbium (0.5%Er) refined the primary Si from 94 μm to 33 μm and the morphology transfers from coarse star-like and polygonal shape to a fine blocky shape. A similar result was obtained by the addition of Er to 05% [15]. However, when the level of rare earth Er was up to 0.8%, the primary and eutectic Si phases became coarser. It is clear from the above-mentioned research that modification of hypereutectic Al-Si alloys by chemical additions is not efficient in reducing the size of the primary Si to a micron-scale or nano size, nor does it produce a desirable shape. Besides, there are a lot of negatives resulting from the chemical agents, such as evaporation, and oxidation during services. Qian et al. [16] and Liao et al. [17] have reported that Sr modifier increases the porosity level and deteriorates the performance of the castings.

Rapid solidification methods, such as melt spinning [18], powder atomisation [19], and laser melting [20–25] proved to be effective approaches for the refinement of the Si phase. The high-cooling rates associated with these processes, which may reach $10^5$ to $10^7$ °C/s play a large role in the size of the critical nuclei, and subsequently, the effective number of
nuclei that will ultimately produce fine-grained structures. Chaus et al. [26] have recently shown that rapid quenching technology resulted in significant grain refinement of the eutectic constituents (α (Al + Si), as well as primary silicon, which gave rise to better tensile and fracture properties compared to that of the conventionally cast Al-18%Si-2%Cu alloy. Xu et al. [18] obtained a 40 µm thick melt-spun ribbon by using a single roller melt-spinning technique of the Al-20 wt.%Si alloy and reported a drastic change in the morphology of the primary Si to fine block-shape with a significant reduction in the size to 2 µm. Kalay et al. [19] have studied gas-atomised Al-Si powders of different compositions (15, 18, 25, and 50 wt.%Si), and showed that as the droplet size decreased, the structure became much finer. The primary silicon in Al-50 wt.%Si alloy was continued to form even at the smallest droplet sizes (25 µm) with sizes ranging from 2 to 5 µm, the Al-25 wt.%Si showed a predominantly eutectic state with a few primary Si while Al-18 wt.%Si, and Al-15 wt.%Si showed dendritic and eutectic structures. Laser surface treatment of hypereutectic Al-Si alloys has been studied by several investigators, who showed a reduction of the primary Si to 10 µm, and the formation of fibrous eutectic Si. Bhowmik et al. [25] have recently developed Al-50 wt.% Si laser-clad on Al 7075 substrate and found local heterogeneity in the composition and the structure with the single clad layer due to different cooling rates. Three phases, namely primary Si of size, Al-Si eutectic, and α-Al dendrites were obtained with an increase in the solubility of Si in Al. Recently, Abboud and Mazumder [21] have produced a series of Al-Si layers on a commercially pure Al substrate with compositions 35, 50, and 60 wt.% and showed a great refinement in all the micro-structural constituents and increased hardness and the highest cooling rates did not suppress the formation of the primary Si but reduced the sizes to 2 µm and increased the amount of the nano-fibrous eutectic. Lein et al. [22] reported a high density of nanotwinned ultrafine Si in hypereutectic Al-Si alloy by laser surface remelting of Al-20 wt.%Si alloy.

It appears from the above-mentioned review that there is an increased interest in the high-silicon content of Al-Si alloys due to their outstanding mechanical properties, especially after the modification. Therefore, to increase the engineering and industrial applications of these alloys, the size, and the morphology of silicon crystals should be refined to a minimum value. The present investigation is an attempt to use the laser melting and remelting technique, at controlled laser processing parameters, for two purposes. First, to fabricate high silicon Al-Si alloyed layers by laser melting at a fixed laser power density and scan speed. The second is to remelt the produced layers at a fast scan speed with relatively low power levels to refine the silicon phase to a nano-sized. The other objective of the study is to measure the nano-indentation hardness in different areas of the eutectic region and correlate the hardness value with the resulting eutectic spacing. The long-term objective of this research is to create a surface layer rich in nanosized silicon crystals on the aluminium surface, which improves its wear resistance [27].

2. Materials and sample preparation

2.1. Preparation of the Al-Si layer

A commercial purity aluminium (CP Al) plate of 50 mm long x 25 mm wide x 6 mm thick, and silicon powder, of 325 mesh, and 99.9% purity have been used to prepare the alloyed layers using conventional laser alloying technique [21]. Four slots are made, two
of which are 0.3 mm deep, and the other two of 0.35 mm deep were prepared along the length of the aluminium plate by a machining process. The slot’s width was kept constant at 1 mm. The reason for preparing slots of different depths is to obtain different silicon ratios after the laser melting. The slot dimensions together with laser processing parameters were chosen based on experience and previous experimental research by Abboud and Mazumder [21]. Silicon powder was inserted and compacted in the grooved aluminium plate until the slot gap is filled and the extra powder was removed. Great care was taken to ensure the uniformity of the silicon powder distribution in the designated spaces before the laser treatment. This method was chosen due to the difficulty of feeding the silicon powder into the feeder, as well as powder jam, which hinders its regular flow.

2.2. Fabrication and remelting of the Al-Si tracklayer

A solid-state Nd-YAG (HLD 4002) disk laser operated at 600–2000 W with a 2 mm beam diameter was used to produce the Al-Si alloyed track layers and for overlapping remelting treatments. The Al-Si alloyed layers were prepared by irradiating the laser beam onto the surface of a commercial purity Al substrate, which contains silicon powder to a specific depth. In this experiment, the laser head was stationary while the Al substrate was moving. The relative movement between the Al substrate and the stationary laser head was controlled by a computerised table moving in three perpendicular directions (X-Y-Z). The movement in X-direction was used to control the scanning speed while in Z direction is used to change the laser beam diameter. The first experiment was designed to fabricate the Al-Si tracklayers of 40 mm length using laser power of 2000 W and at a scanning speed of 10 mm/s, followed by a remelting process at a constant traverse speed of 180 mm/s and different laser powers (1000, 800, and 600 W). Two tracklayers, 1A (two remelts) and 2A (four remelts) were selected for study and analysis. Table 1 illustrates the various processing parameters used in the fabrication and remelting treatments of the tracklayers 1A and 2A. Before performing any remelting experiments, the top surface of the alloyed track was ground, polished, cleaned, and slightly etched. The processes of melting and remelting were conducted within an inert controlled atmosphere to avoid oxidation during melting and solidification.

| Tracklayer no. | Laser treatment | Power W | Speed mm/s | Si-powder thickness, mm |
|---------------|----------------|---------|------------|------------------------|
| 1A            | Fabrication    | 2000    | 10         | 0.3                    |
|               | Remelting      | 2000    | 180        | --                     |
|               | Remelting      | 1000    | 180        | --                     |
| 2A            | Fabrication    | 2000    | 10         | 0.35                   |
|               | Remelting      | 2000    | 180        | --                     |
|               | Remelting      | 1000    | 180        | --                     |
|               | Remelting      | 800     | 180        | --                     |
|               | Remelting      | 600     | 180        | --                     |
2.3. Microstructure characterisation

After the first laser melting and remelting experiments were done, transverse sections were cut, cold mounted, ground on 400, 600, 1200, and 2000 grit SiC paper then polished with a 3 µm and fine polished with 0.3 µm. The samples were cleaned and etched with Keller’s reagent (containing 95 ml of H2O, 2.5 ml of HNO3, 1.5 ml of HCl, and 1.0 ml of HF) for 3 s, rinsed with water, and then dried. The composition and the microstructure of the alloyed zones were examined by scanning electron microscopy (TESCAN MIRA3 FEG SEM) equipped with an energy-dispersive spectrometry (EDS) detector and transmission electron microscopy (JEOL 2010 F AEM). Thin foils were prepared by cutting a slice of 0.3 mm thick parallel to the laser scan and further ground to 100 µm. A punching tool was used to obtain a 3 mm diameter foil. The sample was mechanically polished and finally thinned by a Gatan Precision Ion Polishing System where low-angle and low-current polishing conditions were used in conjunction with a liquid nitrogen cold stage. The electron microscopy and nanoindentation experiments were conducted at the Michigan Center for Materials Characterisation at the University of Michigan.

2.4. Nanoindentation hardness

The nanoindentation hardness and elastic modulus of the eutectic region of tracklayer A-1, which exhibited different spacings, were measured at different locations. The indentation hardness experiment was carried out at room temperature employing a Bruker TI-950 Triboindenter equipped with a diamond tip of a three-sided Berkovich probe. The indenter was employed to measure with a peak load of 1mN and a quasistatic trapezoidal loading function (5 s load and 2 s unload hold time) and an indentation depth of 200 nm. The reported data is an average of at least three indents in each zone. Measurements were carried out on the transverse section, which was mechanically polished to 0.3 µm finished, and cleaned with ethanol.

3. Results

3.1. Tracklayer –1A

3.1.1. General shape and dimension of tracklayer 1A

Figure 1(a) shows a typical transverse section of the laser fabricated Al-Si alloyed layer (tracklayer-1A), which is processed at a power (P) 2000 W, a beam diameter (bd) 2 mm, and scanning speed (v) 10 mm/s. After the completion of the first melt to synthesise the alloy, it was remelted twice at a traverse speed of 180 mm/s and laser powers 2000 W and 1000 W, respectively. The as-fabricated alloyed zone, before its remelted, had a conduction limited shape of a total depth of 0.55 mm and a width of 1.4 mm (Figure 1(a)). It is apparent from the figure that all the silicon powder has melted and dissolved in the molten aluminium forming a heterogenous Al-Si alloy but with a fine structure (Figure 1(a), zone 1). The primary Si particles tend to segregate and cluster at the centre and the edge of the melted zone. However, when remelting was performed at a faster speed and less power, the structure became much finer with less heterogeneity due to the rapid remelting process. The depths of the remelted zones (2 and 3) were 250 and 120 µm at powers of 2000 W and 1000 W, respectively. Figure 1
3.1.2. Composition analysis

EDS analysis was performed at different locations in different zones in the Al-Si alloyed layer, and it was found that the silicon content in the lower part, where the microstructure exhibited a dendritic structure, was slightly lower than that in the other part of the alloyed layer (see Figure 1(a)). The silicon content near the interface was 17.6 wt.%Si, while it was 20 wt%Si in the eutectic region excluding the Si particles and 24 wt.%Si in the region, which showed clustering of Si crystal and the matrix (Figure 2(a)). Similar values were obtained in zones 2 and 3 (Figure 2(b,c)). The average composition of the alloyed layer was found to be 25 wt.%Si. The results of the EDS analysis of the eutectic region showed an increase in the silicon content of the eutectic concentration to 20 wt.%Si in zone 1 and 22 wt.%Si in zone 2 and 24 wt.% in zone 3 (Figure 2(c)). These values are much higher than those reported in the Al-Si diagram (12.6 wt.%Si). This indicates that

(a) also showed a narrow dendritic region at the interface between the base metal and the melted zone 1, which are interpreted as α-Al dendrites. Also, the interfaces between the melted zone 1 and the melted zones 2 and 3 showed an agglomeration of fine silicon particles that coexisted with near eutectic and hypoeutectic structures. The upper part of the third melted zone showed the influence of the convectional fluid flow, which led to the formation of Si-eutectic and Si-poor zones and pushed the silicon particles towards the top and the edges of the melted zone. It is important to mention here that segregation and clustering of the silicon crystals are very common in a conventional cast and even the rapidly cooled hypereutectic Al-Si alloys due to the lower density of the Si particles and the strong convectional fluid flow, which arises from the temperature gradient and other solidification parameters. Figure 1(b) illustrates the decrease of the remelted depth with reducing laser power (or decreasing the heat input), which is equal to the (P/v, J/mm). As the heat input reduces, the depth decreases, and thus the cooling rate increases. Figure 1(b) indicated that the melted depth decreased linearly with the decrease in the heat input and when the heat input reaches a minimum value of 5.55 J/mm, the depth decreases rapidly reaching 120 µm. Due to the smallest melted depth of the third zone, it is expected that the cooling rate will be the highest compared to the first and second melted zones and will display the finest structure.

Figure 1. (a) Cross-section of the laser fabricated Al-Si alloyed layer (tracklayer 1A) and laser remelted at different heat inputs. (b) Variation of the remelted depth with the heat input.
as the melted depth became smaller, the cooling rate increased, pushing the eutectic point towards a higher silicon content as reported by Zhao et al. [24] in an earlier study on Al-20 wt.%Si and by Abboud and Mazumder [20] on Al-17 and 20 wt.%Si alloys.

3.1.3. Microstructural analysis

The microstructure of zone 1 in Figure 1(a), which is processed at 2000 W and 10 mm/s, exhibits a hypoeutectic structure in the lower part, becoming fully eutectic and hypereutectic above the interface. The refined structure, consisting of α-Al dendrites and Al-Si eutectic occupies a volume of approximately 80% of the remelted zone, while the rest is a dispersion of the isolated primary silicon phase (Figure 3(a)). The primary Si crystals are distinguished by their complex and irregular shape with many side branches and with sizes ranging between 3 and 5 µm, while the eutectic structure shows a fibrous morphology with an average spacing of 75 nm (Figure 3(b)). The microstructure of zone 2, which was processed at the scanning speed (180 mm/s) is much finer than that shown in zone 1 and consists of relatively finer primary Si crystals of size about 2 µm with columnar dendrites radiating around and near the primary Si, and a significant increase in the eutectic colonies of a spacing ranging between 30 and 50 nm (Figure 3(c,d)). The major difference between the microstructures of zone 1 and zone 2 apart from the refinement, is the decrease in the volume fraction of the α-Al dendrites in the matrix with an increasing amount of the eutectic.
colonies (Figure 3(a,c)). Furthermore, the primary Si crystals are found to be partially surrounded by the α-Al phase with many eutectics’ colonies grown directly from it (Figure 3(c)).

To achieve further refinement of the microstructural constituents of the Al-25 wt.% layer, a second remelting treatment was performed, but at 1000 W and 180 mm/s. The resulting solidified zone was 125 µm deep. The average size of the primary Si flakes was reduced to ≤ 1.5 µm (Figure 4(a-c)). As the primary silicon became smaller, its morphology changes from star-like to blocky and granular with a slight tendency to branch laterally. From the other side, the Al-Si eutectic was modified and showed an interwoven
fibrous structure of spacing ranging between 20 and 30 nm. Most of the eutectic silicon fibres grew from the vicinity of the primary Si crystals and spreads randomly in different directions (Figure 4(d)).

3.1.4. Nanoindentation hardness
Nanoindentation hardness (NIH) was measured at different locations in the remelted zone of the Al-25 wt.%Si alloyed layer including the fully eutectic and the hypereutectic regions, and the resulting indentation hardness was correlated with the eutectic spacing ($\lambda$). The result showed that the indentation hardness of the transverse section of the remelted Al-25 wt.%Si alloyed layer were 1.75, 2.55, and 3.15 GPa, which corresponds to the eutectic spacing, $\lambda$, of 50, 35, and 20 nm, respectively. It is clear that as the heat input decreases, the melt depth became smaller and cooled rapidly so as a result, the eutectic spacing $\lambda$ decreased, and the nanoindentation hardness increases. The maximum value

![Figure 4](https://example.com/figure4.png)

Figure 4. (a-c) SEM micrographs were taken at different locations in the laser remelted zone of Al-25 wt.%Si layer processed at (1000 W and 180 mm/s) showing sub-micron primary-Si crystals, few $\alpha$-Al dendrites, and a high proportion of the Al-Si eutectic. (d) SEM micrograph showed nanofibrous Al-Si eutectic.
obtained was 3.15 GPa at the top of zone 3, which displayed the finest structure with \( \lambda \), as small as 20 nm. However, some regions in the upper part of zone 3 showed high values ranging between 5.4 and 6.4 GPa. It is expected that the presence of nano-sized primary Si contributed to the high hardness value considering that the indentation of the primary silicon alone is approximately 10 GPa. The effect of the increased nanoindentation hardness (H) with a decrease in the eutectic spacing (\( \lambda \)) is presented in Figure 5. An empirical relationship followed this formula, \( H = K (\lambda)^{-n} \) where H is the hardness in GPa and \( \lambda \) is the average spacing in nm, K and n are constants dependent on the composition of the alloy. In the present work, the value of the constant n is approximately 0.6, while the constant K is 20 \( \pm \) 1 for Al-25 wt.%Si. The general trend of increased strength in ultrafine laser melted Al-Si fibrous eutectic is consistent with earlier studies by Lien et al. [22].

3.2. Track-layer 2-A

3.2.1. Microstructural characterisation

A cross-section of the tracklayer 2A, which was fabricated at a power 2000 W and a speed of 10 mm/s and remelted at 180 mm/s, is presented in Figure 6(a). It is apparent from the figure that both zones 1 and 2 showed a high percentage and conglomeration of the primary Si phase in the centre of the melted zone while at the edges, the concentration was lower. EDS analysis at different locations in zone 1 and zone 2, showed a silicon content in the Si-poor region ranging between 22 and 25 wt.%Si while in the Si-rich region ranged between 29 and 33 wt.% (Figure 6(b)). Based on the optical micrograph, SEM images, and the EDS results, the average content of silicon was estimated to be approximately 30 wt.% Si, which is higher than the 1A tracklayer. Microstructure analysis of the two zones showed a typical hypereutectic structure consisting of a star-like primary Si phase, a high proportion of aluminium dendrites, and Al-Si eutectic. The average size of the primary Si in the
melted zones 1 and 2, were 7 μm and 4 μm, respectively. The resulting structure and the very small size of the primary Si were due to the high cooling rate associated with laser melting. Furthermore, the obtained sizes are much finer than those obtained by the conventional casting process, which were in the range of 50–100 μm and much closer to those obtained using the rapid solidification technique (2–10 μm) that reported by [18,19].

To achieve further refinement of the primary Si crystals to the nanoscale level, the surface of the Al-30 wt.%Si alloyed layer (tracklayer 2A) was subjected to several overlapping laser remelting at a fast speed of 180 mm/s and powers of 1000, 800, and 600 W, respectively. A typical cross-section and a top view of the remelted surface layer are shown in Figure 7(a, b), respectively. Figure 7(b) showed segregation of very fine primary Si near the top surface in all the remelted zones. More details about the microstructure of each zone are presented in Figures 8 to 11. Figure 8 showed the microstructure of zone 2, which was remelted at 2000 W and showed a fine hypoeutectic structure consisting of primary Si crystals of sizes ranging from 2 to 3 μm, embedded in a hypoeutectic matrix containing α-Al dendrites and Al-Si eutectic. Figures 9(a,b) present the microstructure of zone 3, which was remelted at 1000 W and showed a further refinement of the microstructural constituent as the size of the PSi crystals was reduced to 1–2 μm and increased in the proportion of the eutectic structure. Zone 3 marks the beginning of the formation of a modified eutectic resembling a fur-like shape (Figure 9(b)) emerging from the pre-existing primary Si crystals and branched out. Figure 10(a-d) show the microstructure of zone 4, which is remelted at the lowest power 800 W and experienced the highest cooling rate. There was a significant increase in the quantity of the modified eutectic and a decrease in the amount of α-Al dendrites around the primary Si and within the matrix. Furthermore, the sizes of the primary Si crystals were refined to 1 μm, while the eutectic spacing was reduced to less than 20 nm.

3.2.2. TEM analysis
TEM study at different regions in the remelted Al-30 wt.%Si layer showed that the primary silicon crystals and the eutectic silicon contained a high density of parallel twins, while the α-Al phase contains dislocations and ultrafine twinned silicon
The twins in the primary Si crystals were parallel and have different directions with spacings ranging between 5 and 20 nm (Figure 11(b)). The eutectic silicon exhibits different morphologies. Figure 12(a,b) show worm-like Si particles of different lengths nucleated and growing in the α-Al cells/dendrites in addition to the nano-sized Si crystals, which precipitated during cooling to room temperature. Figure 13(a,b) show Al-Si eutectic silicon of nanosized fibrous morphology. It is apparent from the figure that as the laser remelt power is reduced, the eutectic size and spacing became much finer. The fibrous eutectic in zone 4, which
experience the highest cooling rate showed the smallest spacing of approximately 10 to 15 nm. TEM examination revealed a high density of parallel twins extended along the length of the silicon fibre (Figure 11(b)). This observation supported the twin plane re-entrant edge (TPRE) theory that the formation of twins accelerates the growth of the eutectic silicon since silicon growth occurs more readily at the re-entrant edge [28].

### 3.3. Cooling rate

It is known that in laser melting at different heat inputs, as the melted depth decreases, the cooling rate increases, which, in turn, leads to a large undercooling in the smallest depth and produces a fast growth rate. In this study, the cooling rate for Al-Si alloyed layers was calculated using what is known as the Eagar-Tsai approach, which is based on the Rosenthal model [29,30]. The model is based on conduction and considers the laser heat input as a Gaussian area, which is more realistic than the point-input approach. Thermo-physical properties of the Al-Si alloys were kept constant and calculated through the mixing of mass per cent for different alloys. Solution for Rosenthal equation is only possible with numerical integration. Additionally, discretisation of an area of interest (in other words, laser material interaction region) is a must. These requirements increase the computation significantly. Therefore, it was submitted to a workstation after defining the boundary conditions, and input values are thermophysical properties of Al and Si [31] and laser processing parameters. To reduce the computation time, the discretised matrix was solved 10 times, each of which with 0.01 ms time increment. Upon completion, heat history and time steps were used for cooling rate calculations. Results are shown in Table 2.
Discussion

The equilibrium-phase diagram of Al-Si alloy (12.6 wt%Si) shows a great reduction in the sizes of the primary Si and the eutectic spacing to a nanoscale. Although the two curves have the same trend, the decrease in the eutectic spacing is much steeper, especially at the highest cooling rate. This can be seen more clearly in the TEM micrographs in Figure 13(b), which showed a significant decrease in the eutectic spacing to approximately 10–15 nm at the highest cooling rate.

4. Discussion

Results of the laser rapid remelting of the Al-25 wt.%Si and 30 wt.%Si alloyed layers, which were processed at relatively lower laser powers and a higher scan speed, showed the potency of the cooling rate not only in the refining of the silicon size but also in reducing the quantity of the primary Si crystals, decreasing the amount of the aluminium dendrites, and also triggering the eutectic Si to nucleate heterogeneously from the primary silicon phase and branched out of it. Furthermore, the increase of the cooling rate caused a shift in the eutectic composition towards a high silicon content reaching 20 wt.%Si or slightly more, which is far more than the value reported in the Al-Si equilibrium-phase diagram (12.6 wt%Si). In the present results, the microstructure of the fabricated and the remelted layers are hypereutectic and consistent with the Al-Si phase diagram, but with great refinement in the scale of all the phases (primary Si, α-Al, and eutectic Si). Solidification of Al-25 and 30 wt.%Si alloyed layers at cooling rates commence with the nucleation of the primary Si nucleus. Once the Si nuclei form, it grows by rejecting aluminium atoms and accumulating them around the vicinity of the silicon crystals, causing undercooling in the surrounding area leading to the nucleation and growth of the aluminium dendrites around the primary Si. Upon further cooling, the aluminium dendrite continues to nucleate and grow rapidly by rejecting Si atoms on both sides of the dendrite. When the temperature reached the eutectic, the remaining liquid solidified forming Al-Si fibrous eutectic. The final solidified microstructure contains a high amount of aluminium dendrites, which upon further cooling to room temperature, begins with the precipitation of nano-sized silicon particles. The solidification path will follow this sequence, L → primary Si. + L, L → primary Si + (Al around the primary Si) + L, L → primary Si + Alden. + L, L → primary Si + Al + Eutectic. (Al+Si). Due to the high thermal conductivity of the Al phase as compared to the Si and the slow growth of
the silicon phase, the eutectic is formed by nucleation of the Al phase first, which forces the silicon atom to occupy the space between the Al dendrites. However, under a very rapid cooling condition, as in the case of the remelting of the Al-25 and 30 wt.%Si layers at the fastest speed and low heat input, the highest cooling rate led to a reduction in the amount of aluminium that formed around the primary Si crystals, which acted as a nucleus for the growth of the eutectic aluminium phase and this will trigger the eutectic silicon to grow quickly upon the underlying solidified zone of primary Si + eutectic and spread massively in different directions forming a radiant rose.

On the other side, observations by the TEM micrographs at higher magnification demonstrate that most of the primary silicon contains parallel and multiple twins, which have a major contribution to the rapid growth of the eutectic silicon. These results are consistent with the TPRE mechanism [28], which indicate that twins influence the growth and cause branching and twisting of the Si fibres. The formation of heavily
twinned Si fibres is reported by many researchers. Nogita et al. [32] have reported higher twin densities in the modified than the unmodified alloy and twins catalyse crystal growth leading to the formation of complex and more faceted morphology. Another factor that made the silicon phase twin easily is the low stacking fault formation energy of the silicon [33].

5. Conclusions

(1) Hypereutectic Al-%Si alloyed layers with nanoscale structures were fabricated on the surface of a commercial pure aluminium substrate by laser melting technique at controlled laser processing parameters.
The average silicon contents of the produced Al-Si layers were 25 wt.% and 30 wt. %Si and the microstructure was heterogeneous and composed of clustering of primary Si with a size ranging from 5 µm to 7 µm, supersaturated α-Al cells/dendrites, and fibrous Al-Si eutectic.

(2) By laser rapid remelting technique utilising a fast scan speed (180 mm/s) and low powers (800 W), a greater structural refinement was achieved as the size of the primary silicon particles were reduced to less than 1 µm as well as the fibrous eutectic spacings were decreased to approximately 10–15 nm.

(3) The average silicon contents of the produced Al-Si layers were 25 wt.% and 30 wt. %Si and the microstructure was heterogeneous and composed of clustering of primary Si with a size ranging from 5 µm to 7 µm, supersaturated α-Al cells/dendrites, and fibrous Al-Si eutectic.

Figure 12. (a-b): TEM micrographs were taken from the Al-30 wt.%Si alloyed layer, laser remelted at a scan speed of 180 mm/s, and power 2000 W (zone 2 in Figure 7) showing worm-like silicon particles of different lengths within the α-Al cells or dendrites.

Figure 13. TEM micrographs show the nanosized fibrous Al-Si eutectic in Al-30 wt.%Si alloyed layer laser remelted at a scan speed of 180 mm/s and powers: (a)1000 W (zone 3 in Figure 7). (b) 800 W (zone 4 in Figure 7). As the laser power is reduced, the Al-Si eutectic becomes much finer.
(4) The great impact of the cooling rate is not limited to the refinement of the silicon phase and a reduction in the eutectic spacing but rather in the great reduction in the quantity of the α-Al dendrites and increased the proportion of the nanofibrous eutectic substantially. It also triggered the Si eutectic to grow rapidly adjacent to the primary Si templates and spread massively throughout the melted zone.
(5) The nanoindentation hardness was inversely proportional to the Al-Si fibrous eutectic spacing, \( \lambda \), reaching 3.15 GPa at the spacing of 10–15 nm.

(6) Transmission electron microscopy revealed different Si morphologies, such as block-like primary Si, worm-like Si at the boundaries of the \( \alpha \)-Al cells, and nanoscale fibrous eutectic Si, which were internally nano-twinned.

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