

Article

Micro-Alloying with Samarium to Enhance the Thermal Conductivity and Mechanical Properties of Al-5Ni Cast Alloys

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Abstract

The research focuses on how the samarium (Sm) addition amount affects the microstructural development, mechanical characteristics, and thermal conductivity of Al-5wt%Ni alloys. Microstructural characterization revealed that minor Sm additions concurrently refined α -Al grains and eutectic structures (α -Al + Al₃Ni), attributed to the reduced initial nucleation temperature of α -Al and an adsorption inhibition effect on Al₃Ni phase formation. The experimental findings revealed that the integration of Sm improved the yield strength, maximum tensile strength, and heat conduction. The enhancements were chiefly driven by three processes: the refinement of grains, the improvement of the eutectic structure's morphology, and diminishing the dissolvability of Fe/Si impurities within the α -Al matrix. The combined improvement of mechanical and thermal characteristics via regulated Sm addition offers a hopeful approach for the microstructural design of Al-Ni alloys. This study provides a crucial basis for the production of Al-Ni alloys that possess high thermal conductivity and satisfactory mechanical properties.

Keywords: Al-5Ni alloy; micro-alloying element Sm; microstructure; mechanical properties; thermal conductivity



Academic Editor: Shouxun Ji

Received: 19 May 2025

Revised: 26 June 2025

Accepted: 27 June 2025

Published: 30 June 2025

Citation: Huang, Y.; Wang, K.; Zhu, Z.; Cheng, W.; Li, W. Micro-Alloying with Samarium to Enhance the Thermal Conductivity and Mechanical Properties of Al-5Ni Cast Alloys. *Crystals* **2025**, *15*, 609. <https://doi.org/10.3390/cryst15070609>

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1. Introduction

The rapid progress in power semiconductor technology and electric vehicle battery development has led to a growing demand for lightweight structural materials with superior thermal conductivity and strength [1,2]. Aluminum alloys are widely used in the field of thermal management due to their excellent thermal conductivity, lightweight, and good machinability [3,4]. However, traditional cast aluminum alloys such as Al-Si, Al-Cu, and Al-Mg systems exhibit relatively low thermal conductivity (well below 200 W/(m·K)), which makes it increasingly difficult to meet the thermal management requirements of high-power electronic devices [5]. Traditional cast aluminum alloys, owing to their high fluidity, are well-suited for manufacturing complex structural thermal management components. However, the high solid solubility of elements such as Si, Mg, and Cu in the aluminum lattice typically limits their thermal conductivity [6]. Therefore, it is of great significance to develop a novel casting aluminum alloy that possesses both high strength and good thermal conductivity.

The solid solubilities of Ni, Co, and Fe in aluminum alloys are extremely low [7–9], making them promising to be major alloying elements. Luo et al. [9] found that the excellent

thermal conductivity of the Al-2Fe-0.3Co hypereutectic alloy reaches 235 W/(m·K), which is attributed to the low solid solubility of iron and cobalt in the aluminum lattice. However, the Al-2Fe alloy exhibits insufficient fluidity during casting. Al-Ni-based alloys possess a relatively high eutectic composition and low solid solubility of Ni, endowing them with excellent formability and thermal conductivity. These alloys are well-suited for manufacturing structural components requiring specific thermophysical properties [7,10,11]. Despite this, the Al-Ni eutectic alloy's microstructure consists of a rough eutectic form (α -Al + Al₃Ni), leading to a decline in its mechanical characteristics [12]. Modification represents an economical and highly efficient approach for regulating the morphologies and distributions of precipitates that form during the solidification process [13–16]. Therefore, through appropriate modification, it is feasible to enhance both the thermal conductivity (TC) and mechanical properties of Al-Ni-based alloys concurrently.

Aluminum alloys are usually altered by adding a minor amount of rare earth elements like Er, Ce, Sm, among others. Pozdniakov et al. [17] found that Ni and rare earth elements (Y, Er, Yb, Gd) can reduce the volume fraction of the Al₂Cu phase in Al-Cu alloys at 180–250 °C, achieving a good combination of tensile strength and electrical conductivity. For Al-Ni alloys, Suwanpreecha et al. [18] introduced scandium (Sc) into near-eutectic Al-Ni alloys. The study found that the precipitated Al₃Sc phase contributes to enhancing the material's strength. However, the research also pointed out that the solubility of Sc in the aluminum matrix may have an adverse effect on electrical and thermal conductivity (ETC) performance. There were some reports indicating that adding Sm element would not make the TC of aluminum alloys worse as its solubility in α -Al matrix was low [19]. Rao et al. [16] mentioned that introducing Sm might alter both the rough acicular eutectic Si and the needle-shaped β -Al₅FeSi phase in ADC12 alloy by lowering the eutectic melting point. Qiu et al. [20] investigated that Sm had a remarkable effect on the microstructures of eutectic Si, which was transformed from coarse plates to fine and fibrous. The alloy of Al-7Si-0.7Mg, altered with Sm, exhibited an excellent mix of maximum tensile strength (UTS) and stretching.

Present studies are motivated by the scarcity of data in the existing literature on how Sm addition affects the microstructures, mechanical characteristics, and thermal conductivity of almost eutectic Al-5Ni alloys. In this research, the intention was to figure out the inner mechanism of Sm modification and explore the relationship between thermal conductivity and Sm content. Through comprehending how Sm modification influences the microstructure and properties and their mutual relationship, alloy design and processing procedures can be refined for better performance. The study assists in making educated decisions regarding the alteration of Al-Ni-based alloys to produce heat-dissipating materials with enhanced mechanical characteristics and heat transfer efficiency.

2. Materials and Methods

2.1. Preparation of Samples

The permanent mold casting was conducted with the Al-5Ni and the Al-5Ni-*x*Sm (*x* = 0.1, 0.2, 0.3, 0.5) alloys. Initial melting was performed in a resistance furnace by heating 99.7% purity aluminum ingots (containing 0.08 wt.% Si and 0.05 wt.% Fe impurities) to 1023 K. The Al-10Ni master alloy was then added to reach the desired 5 wt.% Ni composition, followed by 5 min of mechanical stirring to ensure melt homogenization. Controlled additions of Al-20Sm master alloy were then incorporated to produce the designed Sm-containing samples. The ultimate molten material was moved into a steel mold that had been preheated to 473 K to solidify. The casting is processed into a tensile specimen.

2.2. Measurements

The metallographic specimens were first ground in stages with abrasive papers and then polished mechanically to obtain smooth, mirror-like surfaces. Etching was carried out using a 0.5 vol.% solution of hydrofluoric acid. Microstructural analysis involved three different techniques: optical microscopy (OM, Leica DMI3000, Wetzlar, Germany), field-emission scanning electron microscopy (FE-SEM, Thermo Fisher Verios G4, Waltham, MA, USA), and energy-dispersive X-ray spectroscopy (EDS, Oxford x-Max, Oxfordshire, UK). Phase composition was analyzed using X-ray diffraction (XRD, Bruker D8 Advance, Karlsruhe, Germany) with Cu-K radiation, at a scan rate of $5^\circ/\text{min}$. For tensile tests, the specimens were tested at room temperature with an Instron 5982 testing machine, maintaining a constant strain speed of 0.5 mm/min. The as-cast Al-5Ni-xSm alloys were machined into disk-shaped specimens ($\phi 12.7 \times 3 \text{ mm}^2$) for thermal property characterization. The Netzsch LFA457 laser flash analyzer was used to measure thermal diffusivity at room temperature. The densities of the samples were ascertained using the Archimedes method, employing deionized water as the immersion medium. The Neumann–Kopp rule was employed to determine specific heat capacities [21,22]. Consequently, we calculated the thermal conductivity (λ) by employing this specific formula:

$$\lambda = \alpha \cdot \rho \cdot C_p \quad (1)$$

Here, α represents the thermal diffusivity (mm^2/s), ρ denotes the density (g/cm^3), and C_p signifies the specific heat capacity ($\text{J}/(\text{g} \cdot \text{K})$). The total measurement uncertainty for thermal conductivity was maintained below 5%. To ensure experimental reproducibility, triplicate measurements were conducted for each specimen, with results reported as the statistical mean. Furthermore, the samples' electrical conductivity was gauged through the vortex technique utilizing the PZ-60A apparatus.

Compared with DSC, the cooling curve thermal analysis (CCTA) technique [23,24], with dynamic temperature change as its core, is more suitable for industrial process simulation and macroscopic characterization of non-equilibrium solidification behavior. Therefore, in this study, a CCTA device was used to collect the cooling curves of molten Al-5Ni-xSm alloy samples. Prior to the experiment, K-type thermocouples were positioned at the center and edge of the melt for temperature collection. During solidification, the temperature evolution was recorded using K-type thermocouples (measurement accuracy: $\pm 0.01^\circ\text{C}$) and an NI 9212 high-speed data acquisition system.

3. Results and Discussion

3.1. Microstructure

Figure 1 illustrates the evolution of the microstructure in Al-5Ni alloys with different levels of samarium addition. The base alloy (Figure 1a) exhibits characteristic dendritic α -Al grains, while incremental Sm incorporation induces progressive microstructural refinement. As can be seen from Figure 1, a distinct trend of secondary dendrite arm spacing (SDAS) reduction can be observed. The secondary dendrite arm spacing (SDAS) of Al-5Ni-xSm alloys was measured by using a digital image analysis software (Nano measurement) [13], is presented in Figure 2. As depicted in this figure, the secondary dendrite arm spacing (SDAS) measures $45.7 \pm 8.2 \mu\text{m}$ in the absence of Sm. When the additional amount of samarium is 0.1 wt.%, the SDAS decreases markedly to $23.4 \pm 4.4 \mu\text{m}$. Upon increasing the Sm content to 0.2 wt.%, the SDAS shows a slight increment to $25.4 \pm 3.8 \mu\text{m}$, followed by no discernible change at 0.3 wt.% Sm. An increase in SDAS to $27.5 \pm 3.9 \mu\text{m}$ is observed when the Sm content reaches 0.5 wt.%. Results indicate that the SDAS reaches the minimum value when the addition amount of Sm reaches 0.1 wt.%. However, this effect gradually diminishes

as the Sm content continues to increase. This microstructural refinement, accompanied by homogenization of α -Al and eutectic structure (α -Al + Al_3Ni) morphology, provides favorable conditions for mechanical property enhancement. The observed microstructural modification indicates that Sm plays a distinct role as a phase morphology modifier in the Al-Ni alloy system.

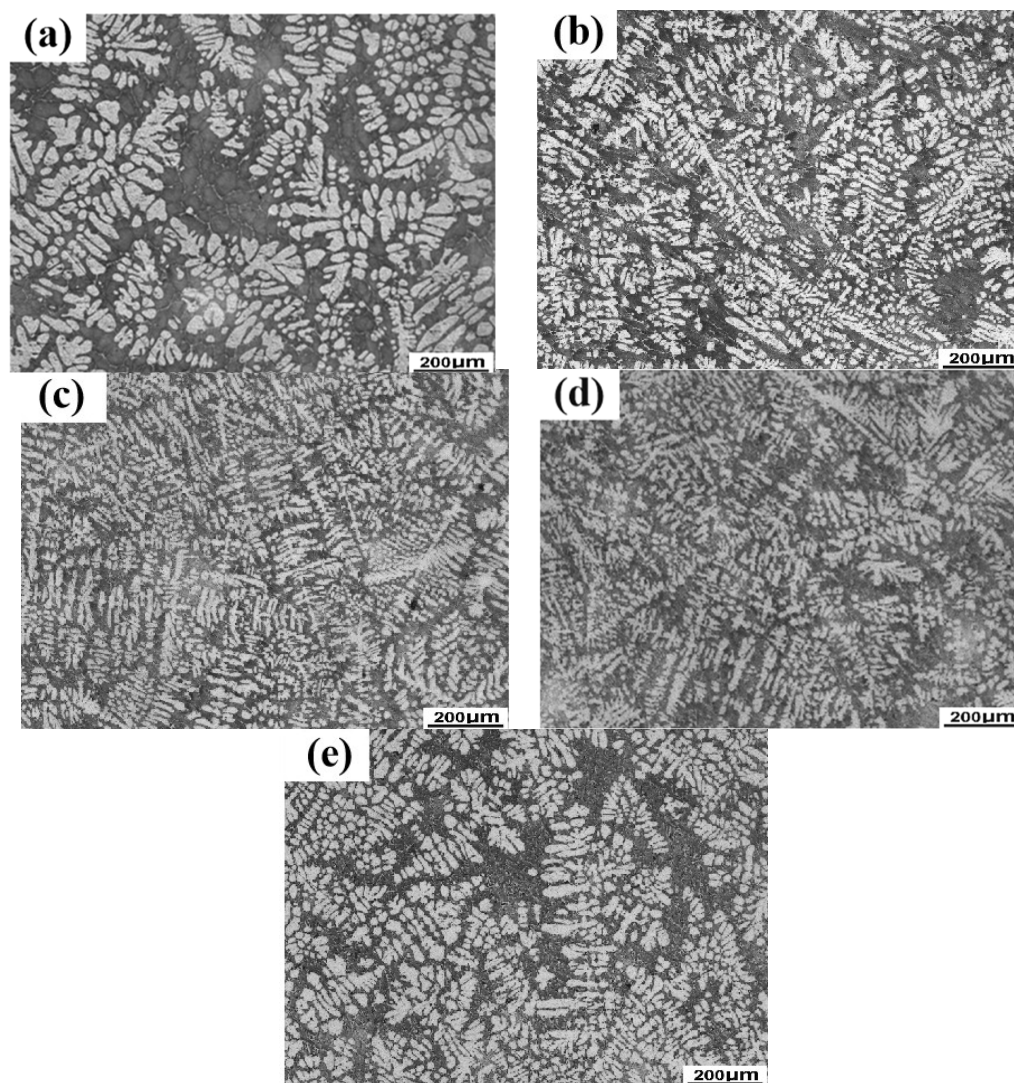


Figure 1. Microstructures of Al-5Ni alloys modified with different Sm additions: (a) 0 wt.%; (b) 0.1 wt.%; (c) 0.2 wt.%; (d) 0.3 wt.%; (e) 0.5 wt.%.

X-ray diffraction analysis of Al-5Ni-xSm alloys (Figure 3) revealed consistent α -Al and Al_3Ni phase formation across all compositions, with Sm-containing variants exhibiting additional Al_3Sm intermetallic phases (Figure 3a). High-resolution examination of Al_3Ni 's strongest diffraction peaks (Figure 3b) showed invariant peak positions regardless of Sm content, indicating preserved lattice parameters. This crystallographic stability suggests Sm primarily interacts through surface segregation at Al_3Ni interfaces rather than bulk solid solution formation. The combined XRD evidence supports a microstructure evolution mechanism involving independent Al_3Sm precipitation alongside unmodified Al_3Ni phase formation.

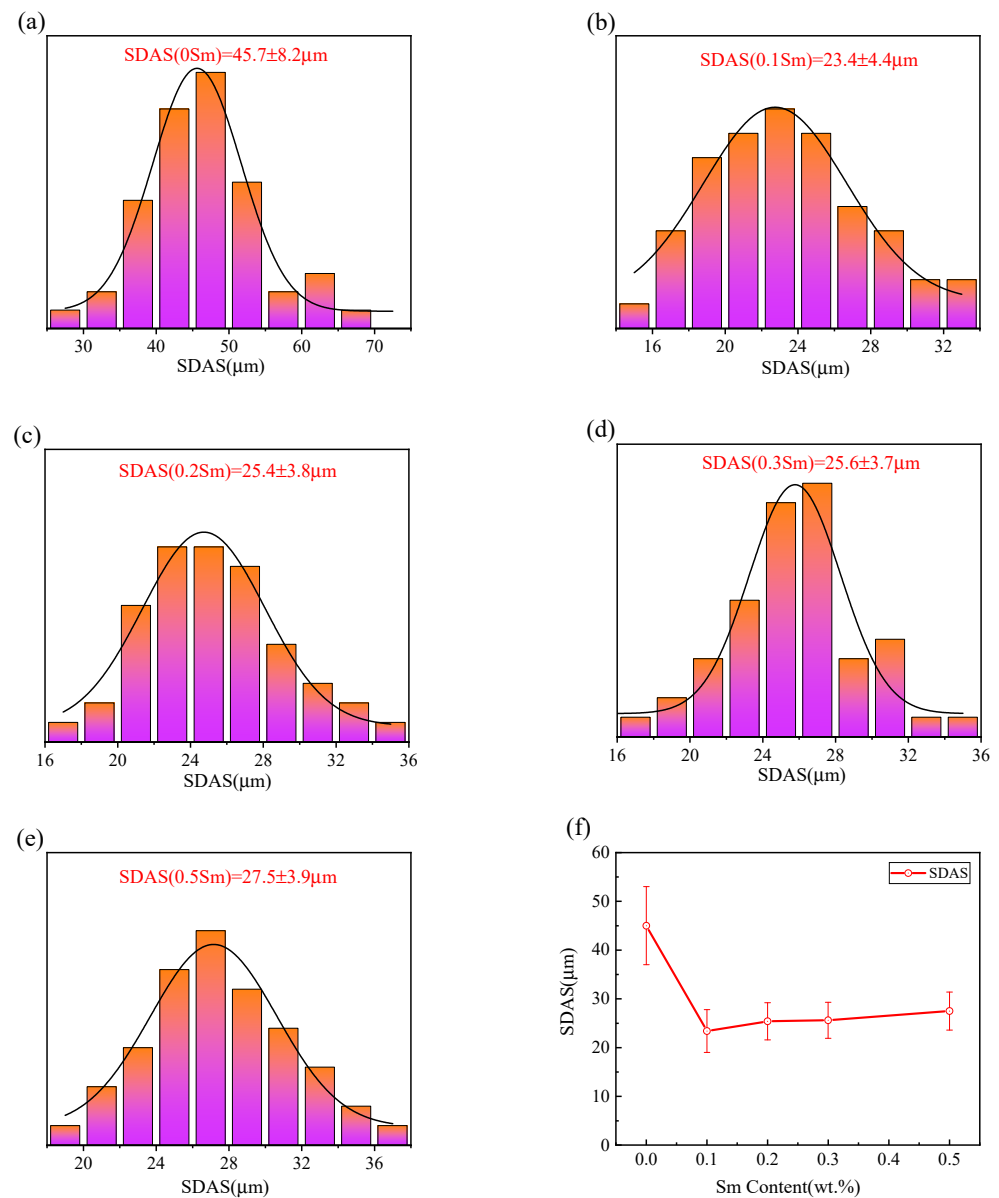


Figure 2. The secondary dendrite arm spacing (SDAS) of Al-5Ni-xSm alloys: (a) 0 wt.%; (b) 0.1 wt.%; (c) 0.2 wt.%; (d) 0.3 wt.%; (e) 0.5 wt.%. (f) The variation trend of SDAS with different Sm contents.

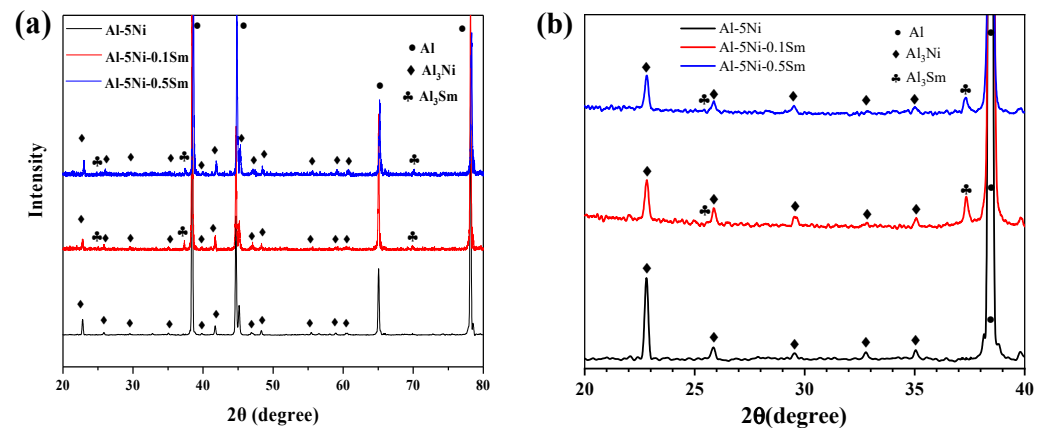


Figure 3. (a) X-ray diffraction patterns of as-cast Al-5Ni-xSm alloys and (b) the zoomed-in XRD spectra in the angular range 20~40°.

3.2. Crystallization

Figure 4 presents the cooling curves of Al-5Ni alloys with varying Sm concentrations, accompanied by their corresponding first and second derivative curves. The solidification parameters are tabulated in Table 1. The base Al-5Ni alloy exhibited α -Al nucleation temperature (T_N) at 656.8 °C, eutectic growth temperature (T_{EG}) at 640.4 °C, and solidus temperature at 621.6 °C. Sm-modified alloys displayed significant solidification characteristic parameter modifications, including a secondary exothermic peak following the (α -Al + Al_3Ni) eutectic reaction (Figure 4b,c). This appearing peak, attributed to the formation of the Al_3Sm phase as confirmed by XRD analysis and the Al-Sm phase diagram [19], shows that the nucleation temperatures of the Al_3Sm phase are 615.2 °C and 606.3 °C for the 0.1 wt.% and 0.5 wt.% Sm-containing alloys, respectively. The progressive intensity enhancement of this peak with increasing Sm content reflects elevated Al_3Sm phase fraction.

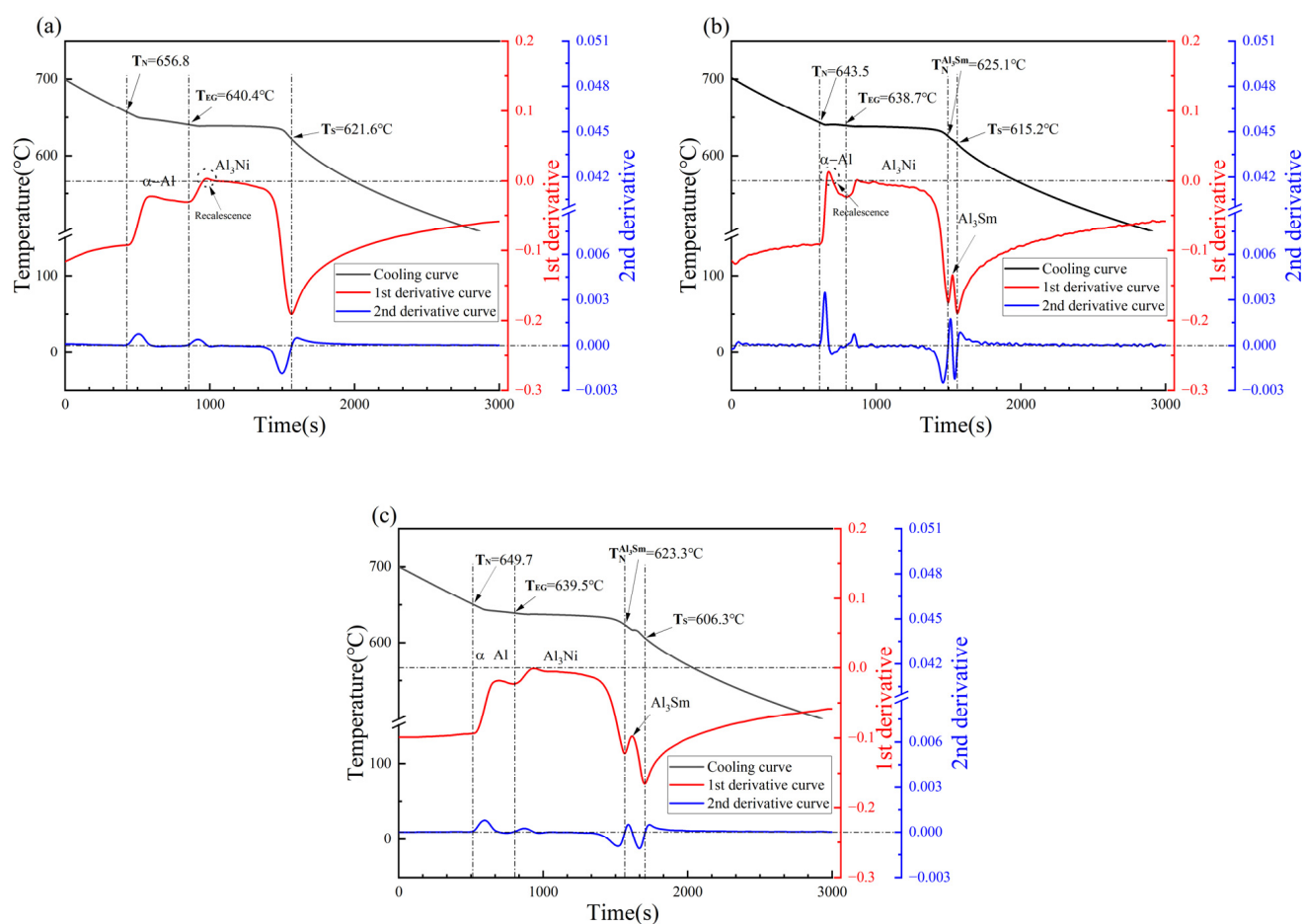


Figure 4. The cooling curves with corresponding first and second derivative curves of (a) Al-5Ni alloy, (b) Al-5Ni-0.1Sm alloy, (c) Al-5Ni-0.5Sm alloy.

Table 1. The solidification characteristic parameters of Al-5Ni alloys containing different contents of Sm.

Sm Content (wt.%)	T_N (°C)	T_{EG} (°C)	$T_N^{Al_3Sm}$ (°C)	T_S (°C)	ΔT (°C)
0	656.8	640.4	-	621.6	35.2
0.1	643.5	638.7	625.1	615.2	28.3
0.5	649.7	639.5	623.3	606.3	43.4

Analysis of solidification parameters (Table 1) reveals a non-monotonic dependence of α -Al nucleation temperature (T_N) on Sm concentration in Al-5Ni-xSm alloys. T_N decreases from 656.8 °C (0 wt.% Sm) to 643.5 °C (0.1 wt.% Sm), followed by a marginal increase to 649.7 °C at higher Sm levels. And the nucleation temperature of the eutectic structure (T_{EG}) exhibits the same trend. The decrease in the initial nucleation temperature implies a greater degree of undercooling to promote the nucleation, thereby refining the grain size.

The most interesting results obtained from the first-derivative curves show that recalescence occurs in both the Al-5Ni alloy and Al-5Ni-0.1Sm alloy (see Figure 4a,b for details). Recalescence is a specific thermal effect observed during the solidification of metals or alloys, manifesting as a temporary temperature rise on the cooling curve [24]. Fundamentally, it results from the latent heat released during solidification, which causes a transient local temperature increase. The presence of recalescence is indicated by the region where the first-order derivative of the cooling curve is greater than zero. In the Al-5Ni alloy, recalescence occurs at the onset of the (α -Al + Al₃Ni) eutectic reaction. This indicates that during the initial stage of the eutectic reaction, nuclei of the eutectic structure form and grow rapidly, releasing substantial latent heat of crystallization. When the latent heat release rate exceeds the system's heat dissipation rate, a localized temporary temperature rise creates a "recalescence peak". During recalescence, the undercooling decreases as temperature rebounds, reducing the nucleation driving force and slowing the crystallization rate, thereby forming a large amount of coarse and brittle eutectic structures.

The addition of Sm elements to the Al-5Ni melt can eliminate the recalescence phenomenon during the eutectic reaction. This elimination helps hinder the development of (α -Al + Al₃Ni) eutectic structures, thereby leading to the refinement of eutectic morphology. During solidification, Sm undergoes solute redistribution, accumulating preferentially at solid-liquid interfaces. This interfacial segregation enhances liquid-phase compositional gradients, increasing composition undercooling that promotes eutectic nucleation while simultaneously restricting eutectic growth through solute redistribute effects. The coupled mechanisms of nucleation site multiplication and diffusion-limited growth kinetics facilitate the development of refined, uniformly distributed eutectic microstructures.

Meanwhile, a pronounced recalescence phenomenon of α -Al was also observed in the Al-5Ni-0.1Sm alloy. Due to the addition of 0.1% Sm, the initial nucleation temperature (T_N) decreases from 656.8 °C to 643.5 °C, generating a higher degree of undercooling. When the undercooling reaches the critical value, massive nucleation and growth of α -Al occur, releasing substantial latent heat of crystallization. However, compared with the Al-5Ni alloy, the growth duration of α -Al is significantly shortened, leading to finer and more uniform α -Al in the Al-5Ni-0.1Sm alloy. As shown in Figure 4c, the recalescence phenomenon of both α -Al and eutectic structures disappears in the Al-5Ni-0.5Sm alloy. Compared with the Al-5Ni-0.1Sm alloy, the growth duration of α -Al increases and its nucleation temperature rises, which leads to the coarsening of α -Al.

The difference (ΔT) between the liquidus temperature (T_N) and solidus temperature (T_S) is the non-equilibrium solidification range of the alloy melt. As shown in Table 1, the addition of 0.1wt% Sm to the Al-5Ni alloy reduces the non-equilibrium solidification range of the melt from 35.2 °C to 28.3 °C, indicating improved fluidity of the alloy. However, when the Sm content increases to 0.5 wt.%, the non-equilibrium solidification range expands to 43.4 °C, leading to deteriorated fluidity of the melt.

3.3. Effect of Sm Addition on Mechanical Properties

Figure 5 displays the tensile stress-strain profiles and mechanical property statistics for Al-5Ni alloys with varying Sm concentrations. The base alloy (0 wt.% Sm) demonstrates yield strength (YS) of 93.2 ± 1.5 MPa, ultimate tensile strength (UTS) of 130.5 ± 2.2 MPa,

and elongation (EL) of $19.5 \pm 0.5\%$. Sm-modified alloys exhibit significant mechanical enhancement, with optimal performance achieved at 0.1 wt.% Sm: $YS = 125.5 \pm 1.4$ MPa (+34.7%), $UTS = 172.4 \pm 2.5$ (+32.1%), and $EL = 20.0 \pm 0.5\%$ (+2.5%). Further Sm addition to 0.5 wt.% induces progressive property reduction ($YS = 112.5 \pm 1.3$ MPa, $UTS = 158.4 \pm 1.9$ MPa, $EL = 18.5 \pm 0.5\%$), demonstrating a concentration-dependent strengthening mechanism. This non-monotonic response suggests an optimal Sm threshold for mechanical performance enhancement in the Al-Ni system. Combining the microstructural observations in Figure 1 and the cooling curve thermal analysis, the addition of Sm reduces the nucleation temperatures of α -Al and eutectic structures, increases the degree of undercooling, and promotes nucleation. Meanwhile, the solute redistribution effect hinders the growth of eutectic structures, which is corroborated by the variation trend of SDAS in Figure 2. Sm addition facilitates the refinement of dendritic α -Al grains and optimization of the size, morphology, and distribution of eutectic structures, thereby significantly enhancing the material strength. Excessive incorporation of Sm induces a poisoning effect, suppressing the effect of lowering the nucleation temperature. This leads to the coarsening of eutectic structures (as illustrated in Figure 1e), thereby causing a gradual decline in mechanical properties.

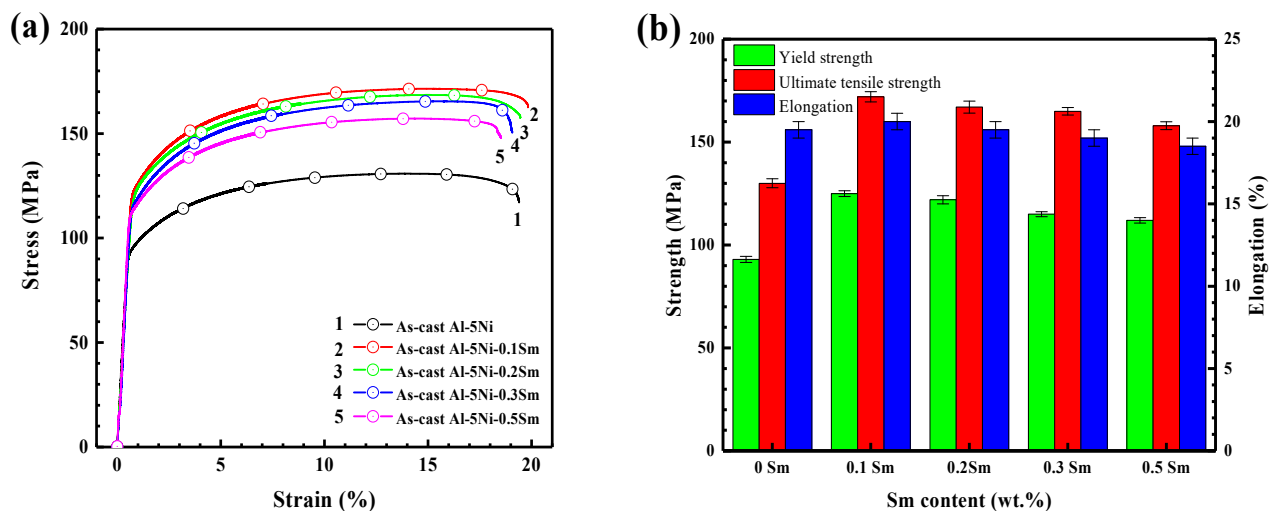


Figure 5. (a) Engineering stress–strain curves and (b) mechanical performance statistics of the Al-5Ni alloys with different Sm contents.

This research determines the Q value, a holistic quality metric, to evaluate the strength and elongation of Al-5Ni-xSm alloys. The Q value is denoted in the form of flow:

$$Q = UTS + a \times \log(\text{Elongation}\%) \quad (2)$$

For aluminum alloy, the value of a is 150 MPa [25,26]. For Al-5Ni alloys containing Sm at concentrations of 0%, 0.1%, 0.2%, 0.3%, and 0.5%, the respective Q values are 323.5 MPa, 367.2 MPa, 360.5 MPa, 356.8 MPa, and 348.1 MPa. Clearly, the Q value for the Al-5Ni-0.1Sm alloy stands at 367.2 MPa, marking it as the highest in the sample set. The figure is approximately 13.5% higher compared to the Sm-free alloy. Therefore, incorporating 0.1% Sm may lead to an impressive combination of strength and elongation performance.

Figure 6 illustrates the tensile fracture morphologies of Al-5Ni-xSm alloys. Examining the structure of tensile fractures allows for a clearer understanding of how these alloys break. In the sample without Sm (as clearly shown in Figure 6a), distinct quasi-cleavages and tear ridges are observable. These features are characteristic of brittle fracture. Moreover, the rod-like eutectic Al_3Ni phases are readily identifiable on the fracture surface. The

experimental findings align well with the relatively inferior tensile performance of the Sm-free alloy. For the Al-5Ni-0.1Sm alloy, Figure 6b vividly illustrates how its fracture surface displays multiple small, uniformly distributed dimples. This morphological feature indicates a transition in the fracture mechanism from brittle to ductile. As the Sm concentration escalates, the dimples grow rougher. Additionally, the fracture surfaces display visible tear ridges, as illustrated in Figure 6c,d, which suggests a mixed fracture morphology characterized by quasi-cleavage and micropore aggregation. The findings from the analysis of the tensile fracture morphology align well with the results of the tensile experiments. Of all the alloys examined, the Al-5Ni-0.1Sm alloy exhibits the most advantageous tensile characteristics.

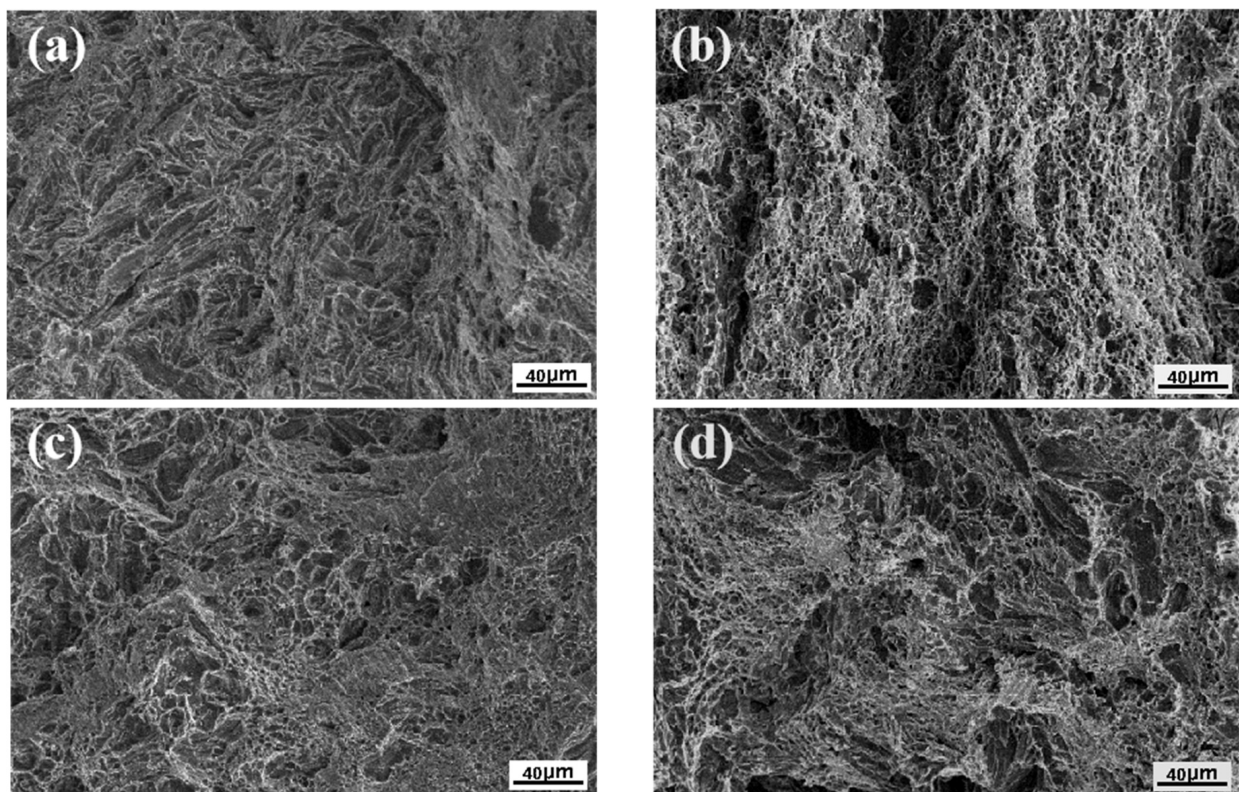


Figure 6. Tensile fracture patterns of Al-5Ni alloys with different Sm contents: (a) 0 wt.%Sm, (b) 0.1 wt.%Sm, (c) 0.3 wt.%Sm, (d) 0.5 wt.%Sm.

The mechanical performance of near-eutectic Al-5Ni-xSm alloys is governed by the morphological characteristics and spatial distribution of the (α -Al + Al₃Ni) eutectic architecture. In Sm-free alloys (Figure 1a), the eutectic structure exhibits a coarse, irregular morphology with uneven dispersion, creating stress concentration sites that detrimentally impact mechanical integrity. This interconnected network of brittle Al₃Ni phases facilitates preferential crack propagation pathways under applied loads, as evidenced by the progressive elongation reduction at elevated Sm concentrations. Sm addition modifies the eutectic structure through morphological refinement and homogenized dispersion (Figure 1b–d), enhancing load transfer efficiency while mitigating stress concentrations. The resultant microstructure optimization, achieved through Sm-induced eutectic modification, demonstrates a strengthening–toughening synergy that significantly improves tensile performance despite minor ductility trade-offs at higher Sm levels. This microstructural engineering approach effectively decouples conventional strength–ductility conflicts in near-eutectic Al–Ni systems through controlled interface design.

3.4. Effect of Sm Addition on the Thermal Conductivity

Figure 7a presents the thermophysical properties of Al-5Ni-xSm alloys, illustrating a linear decrease in specific heat capacity from 0.878 J/(g·K) (0 wt.% Sm) to 0.874 J/(g·K) (0.5 wt.% Sm), attributed to the comparatively low heat capacity of Sm J/(g·K). Concurrently, density increased from 3.009 g/cm³ to 3.033 g/cm³ due to Sm's higher atomic mass. Figure 7b depicts the properties of thermal diffusivity, electrical conductivity, and thermal conductivity. The thermal conductivities are calculated based on thermal diffusivity, heat capacity, and density, in accordance with Equation (1). It is observable that the changes in electrical and thermal conductivity of Al-5Ni-xSm alloys, when varying in Sm content, are similar to the fluctuations in thermal diffusivities. Initially, there is an increase in the electrical and thermal conductivities, alongside a rise in thermal diffusivity, followed by a gradual decline. The alloy without Sm exhibits the minimal levels of thermal diffusivity and electrical and thermal conductivity, recorded at 82.55 ± 0.38 mm²/s, 31.33 ± 0.15 MS/m, and 218.1 ± 1.0 W/(m·K), respectively. The alloy of Al-5Ni-0.1Sm exhibits maximum thermal diffusivity, electrical and thermal conductivity, achieving 82.93 ± 0.29 mm²/s, 31.50 ± 0.11 MS/m, and 219.3 ± 0.8 W/(m·K). The data indicates a rough rise of 0.5%, 0.5%, and 0.6% in comparison to the original Al-5Ni alloy. With an increase in Sm content to 0.5%, there is a gradual reduction in thermal diffusivity, electrical conductivity, and thermal conductivity, dropping to 81.56 ± 0.39 mm²/s, 31.08 ± 0.15 MS/m, and 216.3 ± 1.0 W/(m·K).

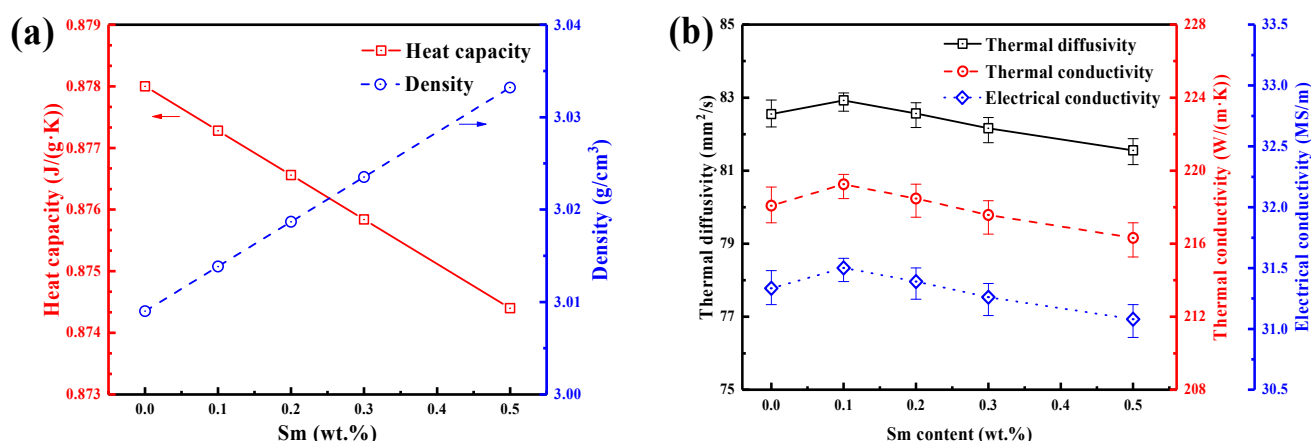


Figure 7. (a) The heat capacity and density of Al-5Ni-xSm alloys, (b) thermal diffusivities and thermal/electrical conductivities of Al-5Ni-xSm alloys.

Research on Al-5Ni-xSm alloys indicates that their increased heat transfer rate is due to the poor solubility of Ni and Sm within the α -Al matrix. Sm exhibits moderate capability for microstructural modification, and the variation in thermal conductivity is intricately correlated with the evolution of the alloy's microstructure. The trace additions of Sm refine (α -Al + Al₃Ni) eutectic structures, concomitantly increasing electron conduction pathways. However, excessive Sm incorporation (0.5 wt.%) gives rise to eutectic coarsening (as shown in Figure 1e), thereby deteriorating thermal conductivity. Previous studies [27] confirm that rare earth elements purify the α -Al matrix through impurity segregation effects. However, detailed interfacial chemistry analysis (to be presented subsequently) is required to fully elucidate these conductivity enhancement mechanisms.

3.5. Promotion Mechanism of Sm Addition on Al-5Ni Alloys

The SEM results, as provided in Figure 8, disclose the effect of samarium addition on the compositions and morphologies of secondary phases in the Al-5Ni alloy. The

morphologies of secondary phases in the Al-5Ni alloy are displayed in Figure 8a. As seen, two kinds of morphologies of secondary phases can be found in the alloy. The secondary phase near the boundary of the α -Al grain, marked as phase 'A' and termed as 'Eutectic boundary phase', is flake-like. The needle-like phases, labeled as phase 'B' and termed as 'Eutectic internal phase', are far from the boundary of the α -Al grain. These secondary phases with different morphologies are extensively generated in the Al-Ni hypoeutectic alloy with different preparation methods [28,29].

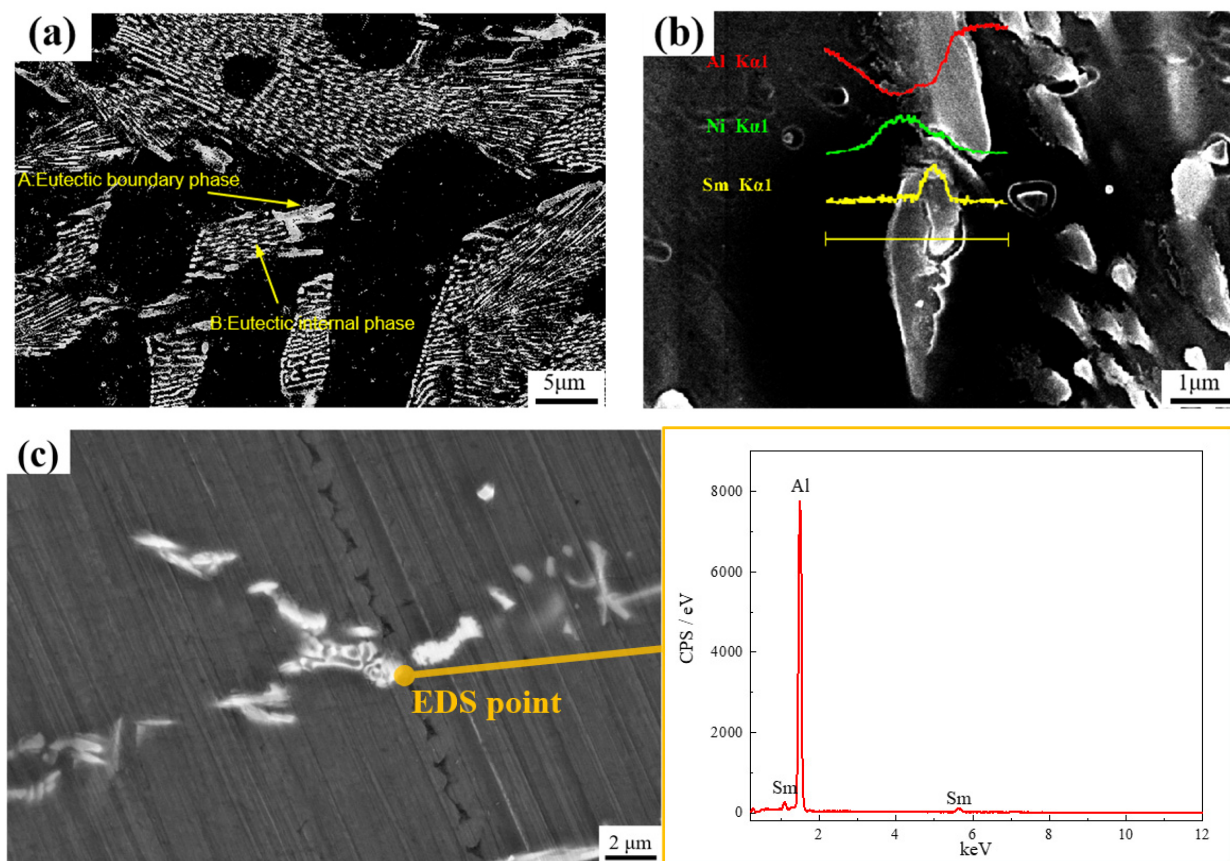


Figure 8. The SEM observation (a) Al-5Ni, (b) Al-5Ni-0.1Sm alloys with corresponding results of line scanning, (c) the precipitation of Al-Sm phase in the Al-5Ni-0.5Sm alloy.

The morphological difference between the eutectic internal phase and eutectic boundary phase is closely connected to the solidification process. Previous studies [30,31] establish 27.6 vol.% as the critical secondary phase fraction for morphological transitions. For this study, the theoretical volume fraction of the Al_3Ni phase, about 8.3 vol.%, is computed by the thermodynamic software JmatPro 7.0. During the early phase of solidification, the Ni element's concentration remains fairly consistent, indicating that the Al_3Ni phase's phase volume fraction is less than 27.6 vol.%. Observing the rod or needle-shaped Al_3Ni phase within the eutectic grain is straightforward. During the last phase of solidification, the residual liquid exhibits a high Ni element concentration due to the redistribution of solutes and non-equilibrium solidification. This occurrence reveals that the Al_3Ni phase's related phase volume fraction exceeds 27.6 vol.%. Hence, the eutectic boundary phase manifests a morphology resembling flakes.

Introducing the Sm element might lead to the creation of a flake-like eutectic boundary phase due to adsorption poisoning. Highly active Sm element will be enriched at the boundary of the liquid/solid interface, inhibiting the migration and diffusion of Ni atoms. Therefore, there are more remaining liquids with high Ni concentration, which facilitates

the formation of flake-like eutectic boundary phase. The relative line scanning results for the secondary phase in the Al-5Ni-0.1Sm alloy are provided in Figure 8b. The experimental results demonstrate that the Sm element mainly distributes alongside the flake-like eutectic boundary phase. Combined with the XRD analysis, the Sm-rich layer may be the Al_3Sm phase. In other words, the flake-like Al_3Ni phases are coated with Al_3Sm phases. Figure 8c displays the morphology of the Al-Sm phase. By referring to the XRD patterns, the Al-Sm phase can be determined as Al_3Sm , indicating that the excessive addition of samarium will lead to the precipitation of Al_3Sm in the Al-5Ni matrix.

Figure 9 presents EPMA elemental mapping of the Al-5Ni alloy. Illustrated in Figure 9c,d, the minimal presence of Fe and Si impurities renders these elements almost undetectable in the initial alloy, a fact supported by the experimental findings. Figure 10 depicts the EPMA mapping outcomes for the Al-5Ni-0.1Sm alloy. Importantly, Figure 10c,d distinctly show clusters of Fe and Si located at the edges of grains. The finding suggests that introducing a minor quantity of the Sm element markedly reduces the solubility of Fe and Si impurities in the α -Al matrix. Furthermore, an examination of Figure 10e makes it evident that Sm predominantly accumulates in the region surrounding the eutectic Al_3Ni phase.

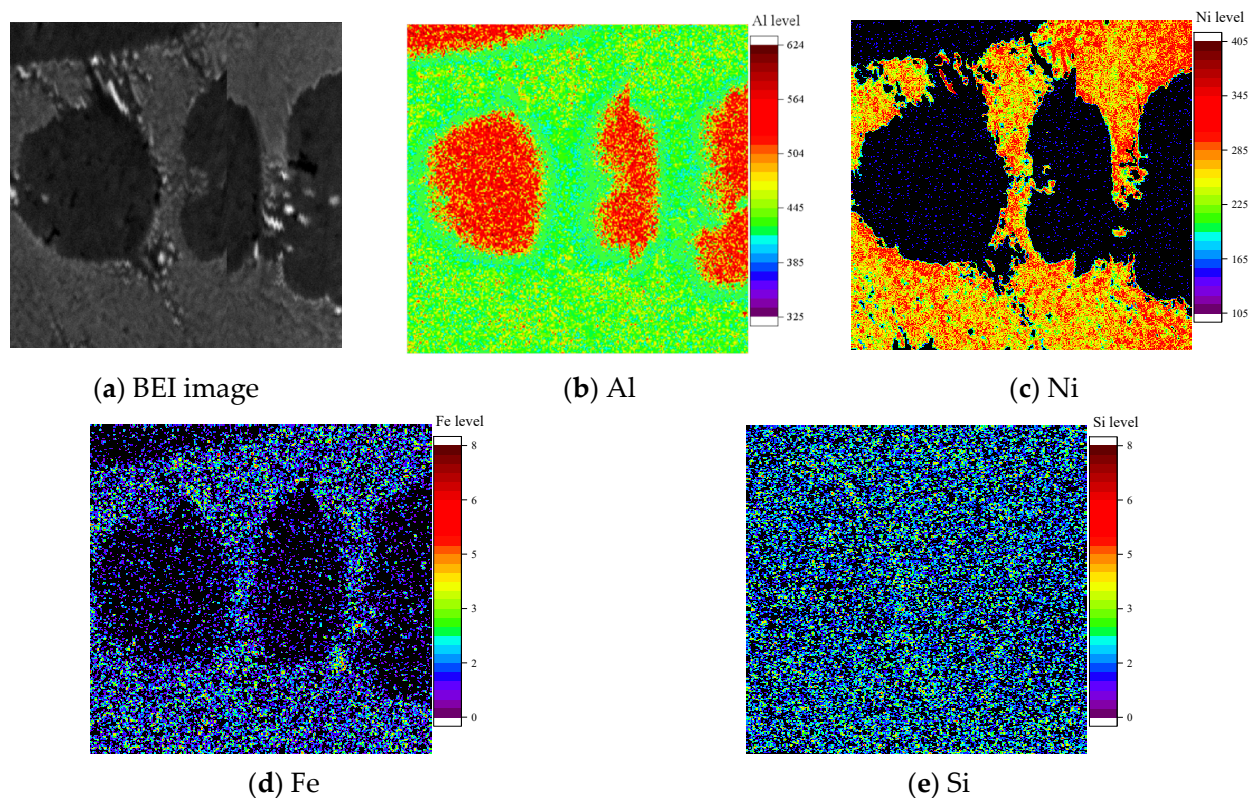


Figure 9. EPMA mapping of Al-5.0Ni alloy. (a) BEI image, (b) Al, (c) Ni, (d) Fe, and (e) Si.

In metallic systems, thermal transport arises from coupled electron and phonon contributions, with total thermal conductivity (λ) comprising electronic (λ_e) and lattice (λ_g) components as per established models [22]. Among these, the electronic thermal conductivity plays a more dominant role compared to the lattice component. According to the Wiedemann–Franz rule [32,33], a proportional correlation is observed between electronic thermal and electrical conductivity. Concurrently, the Mattissen–Flemming rule [34,35] suggests that the main contributors to electrical resistance are solid solute atoms, contaminants, and imperfections. The associated expression is shown as follows:

$$\rho = \rho^{Al_{pure}} + \rho^{precipitates} + S_{GB}\Delta\rho^{GB} + L_{dislo}\Delta\rho^{dislo} + C_v\Delta\rho^{vac} + \sum_i C_{solu}^i \Delta\rho_{solu}^i. \quad (3)$$

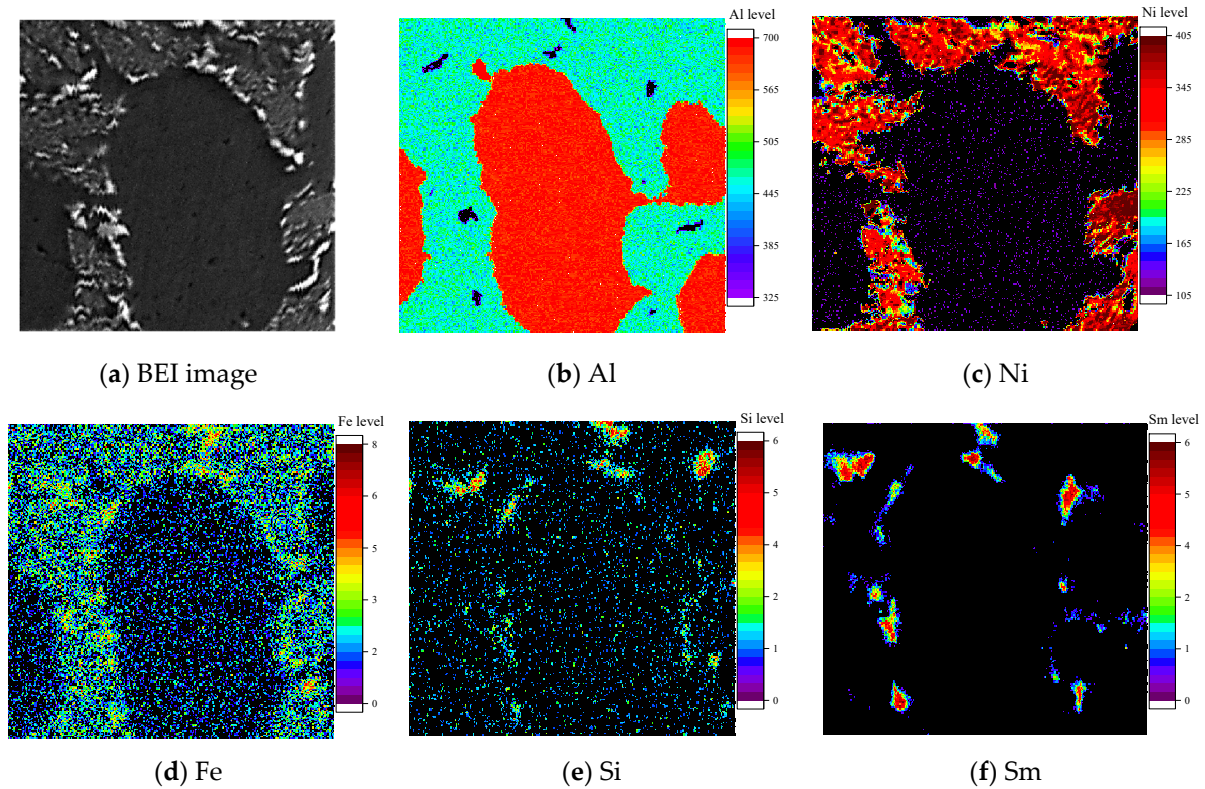


Figure 10. EPMA mapping of Al-5.0Ni-0.1Sm alloy. (a) BEI image, (b) Al, (c) Ni, (d) Fe, (e) Si, and (f) Sm.

The value of $\rho^{Al_{pure}}$ is $2.665 \times 10^{-8} \Omega \cdot m$, as reported in reference [34]. $\rho^{precipitates}$ represents the electrical resistivity caused by precipitates. The negligible effect of minor Sm additions on precipitate volume fraction permits the exclusion of precipitation-mediated resistivity contributions. The coefficients ($\Delta\rho^{GB}$, $\Delta\rho^{dislo}$, $\Delta\rho^{vac}$, and $\Delta\rho_{solu}^i$), represent the individual contributions of grain boundaries, dislocations, vacancies, and solid solute atoms to the electrical resistance. S_{GB} is the grain boundary fraction per unit volume, L_{dislo} is the dislocation density, C_v is the vacancy concentration and C_{solu}^i is the concentration of solid solute atom i . According to Miyajama's study [36], for aluminum, $\Delta\rho^{GB}$ is approximately $2.6 \times 10^{-15} \Omega \cdot m^2$, $\Delta\rho^{dislo}$ is around $2.7 \times 10^{-25} \Omega \cdot m^3$, and $\Delta\rho^{vac}$ is about $2.6 \times 10^{-8} \Omega \cdot m/at. \%$. Considering the minimal magnitudes of $\Delta\rho^{GB}$, $\Delta\rho^{dislo}$, and $\Delta\rho^{vac}$, the impact of grain boundaries, dislocations, and vacancies on electrical resistance is trivial. As a result, the electrical resistance is greatly influenced by the levels of atoms in solid solutions. Mulazimoglu [37] and Cooper [38] quantified solute-specific resistivity coefficients through systematic alloy characterization, deriving empirical relationships between aluminum's electrical resistivity and solute concentrations. These correlations are mathematically expressed as

$$\Delta\rho_{solu}^{Si} = 7 \times 10^{-9} \Omega \cdot m \cdot (wt. \%)^{-1} \quad (4)$$

$$\Delta\rho_{solu}^{Fe} = 1.682 \times 10^{-8} \Omega \cdot m \cdot (wt. \%)^{-1} \quad (5)$$

According to the aforementioned formulas, each 1 wt.% rise in Si or Fe content results in a roughly 26.4% or 63.4% reduction in the electrical resistance of pure Al, respectively. In reality, trace amounts of Fe and Si elements are invariably present in commercially

pure aluminum. Moreover, as Jiang et al. [8] and Zhang et al. [27] emphasized that rare earth elements present in molten aluminum possess pronounced interfacial adsorption capabilities. Such a trait grants them exceptional purification capabilities for the α -Al matrix. In this study, the same observation (as illustrated in Figure 9c,d) reveals that Sm can also extract Fe and Si elements from the Al-5Ni melt. This extraction process effectively reduces the solid solubility of these impurities within the alloy. Moreover, as illustrated in Figure 1, the incorporation of Sm results in notable refinement of both α -Al grains and the eutectic (α -Al + Al_3Ni) structure. This microstructural refinement contributes significantly to improved thermal conductivity. The mechanism driving this phenomenon has been thoroughly detailed in our earlier work [13]. The reduction in grain size and optimization of the eutectic arrangement increases intergranular contact surface, thereby expanding pathways for electron flow. As a result, Sm-induced refinement directly promotes higher thermal conductivity in the near-eutectic Al-5Ni alloy. However, the excessive addition of samarium in the Al-5Ni alloy results in the precipitation of Al_3Sm within the Al matrix (as illustrated in Figure 8c). As an intermetallic phase in the Al-5Ni alloy, Al_3Sm acts as a scattering source that hinders the transmission of free electrons. Consequently, as illustrated in Figure 7b, the TC property of the Al-5Ni alloy decreases with increasing samarium content.

4. Conclusions

This study systematically investigates the influence of samarium incorporation on the microstructure, solidification behavior, mechanical characteristics, and thermophysical behavior of Al-5Ni alloys, offering insight into the underlying rare earth modification mechanisms. The principal findings are outlined as follows:

(1) The introduction of Sm leads to simultaneous refinement of α -Al grains and eutectic (α -Al + Al_3Ni) phases, driven by two synergistic mechanisms—enhanced nucleation undercooling of the α -Al matrix and suppressed Al_3Ni phase growth via interfacial adsorption effects.

(2) The controlled Sm alloying significantly enhances the tensile properties and thermal conductivity (TC). This is attributed to the addition of moderate Sm (0.1 wt.%), which lowers the nucleation temperature of dendritic particles and promotes solute redistribution, thereby refining the size of eutectic structures and optimizing their morphology and distribution to improve mechanical properties. Meanwhile, the thermal conductivity is enhanced by inhibiting the solid solubility of Fe/Si elements in the α -Al matrix.

Author Contributions: Conceptualization, K.W.; data curation, Y.H.; formal analysis, Y.H.; funding acquisition, W.L.; investigation, Y.H., Z.Z. and W.C.; methodology, K.W.; project administration, W.L.; resources, W.L.; supervision, W.L.; validation, Z.Z.; writing—original draft, Y.H.; writing—review and editing, Y.H. All authors have read and agreed to the published version of the manuscript.

Funding: The Guangdong Major Project of Basic and Applied Basic Research (2021B1515130010), Guangdong Major Project of Basic and Applied Basic Research (2020B0301030001), Science and Technology Service Network Initiative of Chinese Academy of Sciences (Dongguan Special Project: 20211600200082), Guangdong University's Innovative Team Project (2021KCXTD022), Dongguan Social Development Science and Technology Project (NO.20231800938572), and The Fund Project of Laboratory Management Professional Committee of Guangdong Higher Education Association (GDJ20240022) have all provided funding for this study.

Data Availability Statement: The original contributions presented in the study are included in the article; further inquiries can be directed at the corresponding author.

Acknowledgments: The writers express their gratitude for the support provided by Dongguan University of Technology's Analytical and Testing Center in conducting the SEM observation. In

the meantime, we recognize the assistance provided by the Guangdong Research Center for High Performance Light Alloys Forming Technology, along with the Novel Light Alloy and its Process Technology Key Laboratory in Dongguan City.

Conflicts of Interest: The authors declare no conflicts of interest.

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