

Ni-based superalloy synergistic strengthened by in-situ nano-carbides and residual graphene fabricated via laser powder bed fusion

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Abstract: With the increase in power of the industrial gas turbine and thrust-weight ratio of aeroengine, the conventional strengthening method of adding refractory elements into superalloys has become difficult to meet the demands for the higher mechanical properties. A novel Ni-based superalloy was designed with enhanced strength and hardness based on the graphene nanosheets (GNs) synergistic in-situ nano-carbides strengthening in the present work. Nano-carbides were induced by in-situ reaction of the GNs with alloy powders during additive manufacturing. The microstructure and thermophysical properties of different alloys with 0.1wt.% GNs and without GNs were investigated by SEM, EBSD, TEM, differential scanning calorimetry (DSC), and small angle neutron scattering (SANS). Residual GNs were also detected by SANS and DSC. The nano-carbides are uniformly distributed in the matrix and combine with residual GNs to refine the cellular structure. Compared with the original alloy (ASE100), the hardness of the alloy with 0.1wt.% GNs (ASE100-0.1GN) is increased by 31 HV (from 315 HV to 346 HV), and the yield tensile strength is increased by 86 MPa (from 756 MPa to 842 MPa). The GNs react with alloy melt in the molten pools to generate nano-carbides under the Marangoni effect during manufacturing process. The dispersion nano-carbides are distributed at both grain boundaries and within grains, effectively hindering the movement of dislocation and enhancing the strength of alloy.

Keywords: synergistic strengthening; in-situ reaction; additive manufacturing; superalloy; nano-carbides strengthening

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1 Introduction

As the power of industry gas turbines and thrust-weight ratio of aeroengines continue to increase, superalloys

with higher mechanical properties and microstructure stability are required^[1,2]. The typical strengthening methods in superalloys contain solid solution strengthening and precipitation strengthening, which rely on the addition of refractory elements and the introduction of nano-precipitates. Recently, more attention has been paid to the strengthening method by introducing nanoparticles in superalloys, instead of the traditional strategy of adding more refractory elements^[3,4].

As classical nanoparticles, TiC carbides have a wide range of applications in the field of superalloys in recent years^[5-10]. Jiang et al.^[11] introduced 0.02wt.% TiC nanoparticles into the K214 superalloy melt and fabricated the strengthened alloy by vacuum induction

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melting. Finer grains, coarse secondary dendrite arms, and lower micro-segregation were obtained. The ultimate tensile strength and elongation of the strengthened alloy at room temperature increased from 966 MPa and 1.5% to 1,060 MPa and 2.8%, respectively. Furthermore, Huang et al.^[9] placed the Ni-Cr-W Ni-based superalloy ingots and the TiC nanoparticles package in the melting crucible of the vacuum induction melting furnace to prepare the strengthened alloy. The TiC nanoparticles mainly acted as nucleation sites for MC-type carbides and accelerated the precipitation of MC carbides, which refines the carbides and enhances the high temperature strength^[9]. However, it remains challenging to control the size and distribution of TiC nanoparticles when directly introduced into the melt. Recent studies are focused on addressing this critical issue by utilizing more advanced fabrication processes or designing an effective introduction method.

Laser powder bed fusion (LPBF) is an additive manufacturing (AM) technique used for fabricating alloy components with complex geometry and internal structure^[12-19]. High melt cooling rates and great temperature gradient achieved by fast laser scanning in LPBF provide a solution to the coarsening and inhomogeneous distribution of the nanoparticles^[10, 20-23]. Taheri et al.^[24] demonstrated that TiC nanoparticles can act as heterogeneous nucleation sites during solidification of superalloy. This refined the grains, thus reducing the strain concentration and inhibiting the cracks in GTD-111 superalloy fabricated by LPBF. Furthermore, the TiC nanoparticles were also reported to refine the grains and reduce local strain concentration in the LPBF CM247LC superalloy^[25]. The uniform distribution of TiC nanoparticles can enhance the strength of grain boundaries, which decreases the crack tendency of CM247LC superalloy via LPBF^[25]. The LPBF creates possibility for TiC nanoparticles strengthened superalloy, and it has potential for further enhancement. The formation of TiC nanoparticles through the in-situ reaction of graphene nanosheets (GNs) with superalloys may suggest a novel method.

GNs exhibit a great potential due to the excellent thermal conductivity, high aspect ratio, and intrinsic strength and modulus^[26, 27]. Gao et al.^[28] demonstrated that carbides induced by in-situ reaction improve the interfacial strength between graphene and the matrix, thereby enhancing the mechanical transfer efficiency of the superalloys fabricated by hot isostatic pressure. Furthermore, the yield strength and ultimate tensile strength were significantly increased by grain refinement, load transfer, and Orowan looping strengthening provided by GNs in K418 superalloy manufactured by LPBF^[20]. Meanwhile, the author pointed out that a few of GNs with defects reacting with Nb and Mo to form (Mo,Nb)C carbides, which precipitated along the grain boundaries and may serve as a secondary

strengthening phase. However, further investigation on these reaction-formed carbides is lacking.

In the present work, the concept of in-situ alloying reaction was employed to design an ASE100-0.1GN superalloy (ASE100 with 0.1wt.% GNs), which was reinforced synergistically by GNs and in-situ formed nano-carbides. The strengthened alloy with high strength and hardness was manufactured by LPBF. The microstructure was observed and the formation mechanisms were analyzed. This work expands the application scope of superalloys and provides reference for the design and fabrication of nanoparticles strengthening superalloys.

2 Materials and experimental procedure

2.1 Powder characterization

The gas-atomized ASE100 alloy powders with particle sizes in a range of 15–53 μm were used for preparing the composite powders. GNs (0.1wt.%, with a diameter of approximately 30 μm and a thickness of approximately 4 nm) were uniformly mixed with the ASE100 alloy powder in the deionized water and then dried at 90 °C. The specific chemical compositions of the ASE100 alloy powder and composite powder are listed in Table 1. The introduction of GNs increased the corresponding C content in composite powder without influencing the content of other elements. Powder morphology was further characterized by a Zeiss Sigma 500 scanning electron microscope (SEM) to verify the homogeneity of GNs, as shown in Figs. 1(a–d). Size distributions of the two powders reveal a high degree of similarity, whereas the uniformly distributed GNs are adsorbed on the surface of the composite powder, as seen in Fig. 1(d). The EDS mapping analysis of the local region in composite powder also confirms the presence of GNs [Fig. 1(e)].

2.2 LPBF process and characterization

The LPBF was performed using the HBD-150 SLM system. The selection of process parameters was based on the optimal parameters of ASE100 alloy, and the process parameters were as follows: laser power of 180 W, scanning speed of 800 $\text{mm}\cdot\text{s}^{-1}$, thickness of 30 μm , and hatch distance of 0.1 mm ^[7]. Cubic samples in size of 8 mm×8 mm×8 mm were fabricated to investigate the microstructure, and multiple tensile samples with gauge dimensions of 60 mm×14 mm×2.5 mm were prepared for mechanical tests, as shown in Fig. 2. The samples manufactured using ASE100 alloy powder and composite powder (0.1wt.% GNs) were named as ASE100 and ASE100-0.1GN.

The different alloy samples used for microstructure

Table 1: Chemical compositions of ASE100 alloy powder and composite alloy powder (wt.%)

	Cr+Mo+W	Co	Ti+Nb	Al	C	Ni
ASE100 powder	17.00	25.00	5.50	2.30	0.02	Bal.
Composite powder	17.00	25.00	5.50	2.30	0.12	Bal.

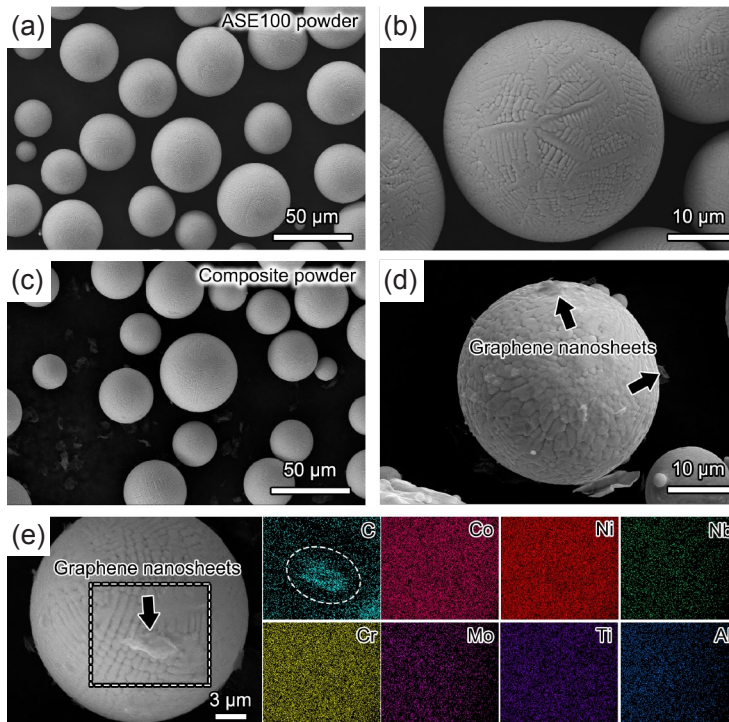


Fig. 1: Microscopic morphology of different powders: (a, b) ASE100 alloy powder; (c, d) composite powder; (e) EDS mapping analysis of composite powder

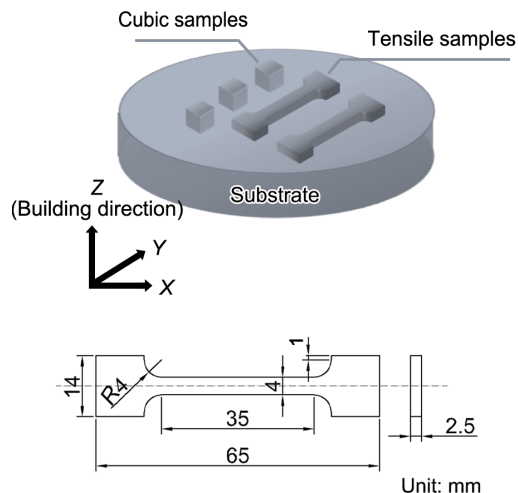


Fig. 2: Schematic of as-built samples fabricated by different powders and dimension of tensile samples (unit: mm)

investigation were ground with SiC emery paper from 200 to 2000 grits and polished with oxide polishing suspension. The as-built samples were electro-chemically etched in a solution composed of 15 g CrO_3 +10 mL H_2SO_4 +150 mL H_3PO_4 at 5 V for 5 s to reveal the molten pool boundaries and cellular structure. The optical topography and microstructure observation were conducted on a Zeiss Axiovert 200MAT optical microscope and a Zeiss Sigma 500 scanning electron microscope equipped with an Oxford EBSD detector system and a FEI Talos F200X scanning transmission electron microscope (STEM). The thin foils for TEM characterization were mechanically ground and twin-jet electro-polished in an alcoholic solution containing 10vol.% perchloric acid at 21 V and -25°C . Measurement of the cell size was performed by Image-Pro Plus 6.0 software.

The small-angle neutron scattering (SANS) investigations were performed at the China Institute of Atomic Energy (CIAE). The samples in the as-built state were cut into 10 mm×7 mm plates with a thickness between 3 mm and 4 mm for SANS measurements. The SANS experiments were performed using the very small angle neutron scattering (VSANS) instrument BL-14 at the China spallation neutron source (CSNS). Neutron data were collected with a wavelength range from 2.2 Å to 6.7 Å, employing a collimation length of 12.75 m. Samples were encased in quartz cells with an optical path length of 1 mm (or 2 mm or in sapphire cells with an optical path length of 1 mm or 2 mm). Data reduction was performed using the direct beam method which included measurement of the direct beam, solid angle and transmission correction as well as solvent scattering background (or quartz empty cell scattering background) subtraction.

2.3 Mechanical property tests

Ambient temperature tensile tests (MTS E45.305) were carried out using a strain rate of $10^{-3}\cdot\text{s}^{-1}$. The microhardness tests were performed at the horizontal and vertical sections using a Vickers microhardness tester (YMVS-1) under the load of 0.3 kg with 10 s holding time.

3 Results and discussion

3.1 Characterization of matrix and nanoparticles

Figure 3 illustrates the metallurgical micrographs of the ASE100 alloy and ASE100-0.1GN alloy in as-built state in the XOZ plane. A typical fish-scale molten pool is observed on the vertical plane parallel to the building direction both in the ASE100 and ASE100-0.1GN, which contains a substantial number of small micro-dendrites [Figs. 3(c, f)], while the depth and width of the molten pool exhibit a significant difference. The molten pools become narrower and deeper after the addition of GNs. The depth of molten pools in ASE100 is approximately 56.5 μm. In comparison, the molten pool depth of the ASE100-0.1GN increases by 140%, reaching approximately 135.6 μm. The deeper and narrower melt pools of ASE100-0.1GN are mainly attributed to the excellent thermal conductivity and laser absorption of GNs^[7]. The energy transfer efficiency of the laser is enhanced by the addition of GNs, which leads to the acceleration of energy density and temperature gradient during the manufacturing process. Therefore, the depth of molten pools is increased in ASE100-0.1GN.

Further observation is conducted inside the molten

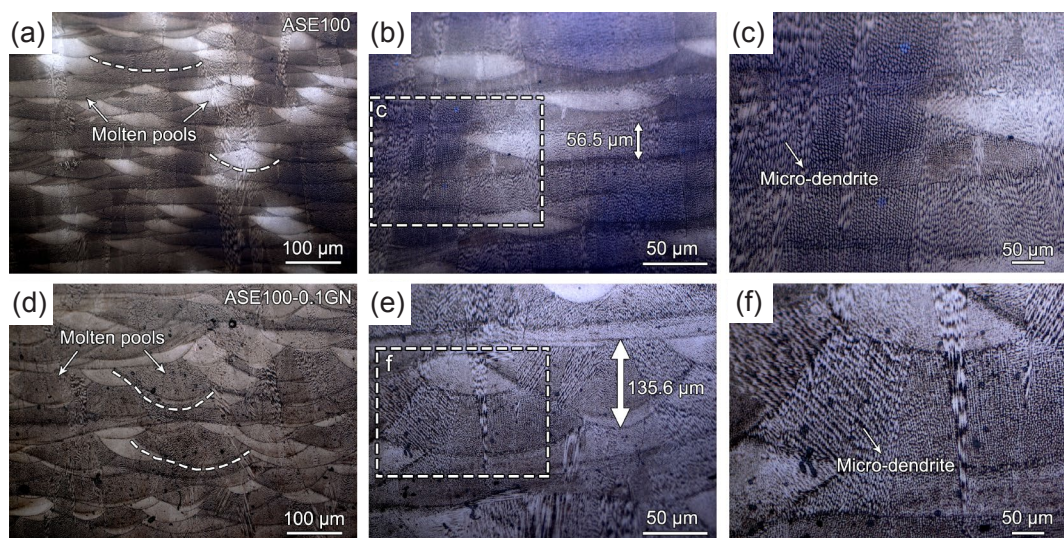


Fig. 3: Optical images of as-built ASE100 (a-c) and as-built ASE100-0.1GN (d-f) in XOZ plane

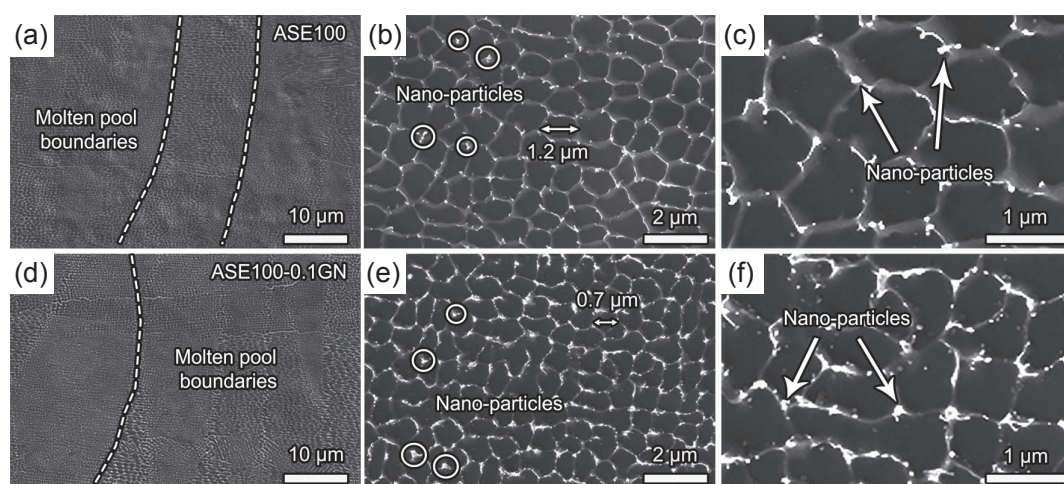


Fig. 4: SEM images of as-built ASE100 (a-c) and ASE100-0.1GN (d-f) in XOZ plane

pools of ASE100 and ASE100-0.1GN (Fig. 4). The molten pool boundary and cellular structure are presented in both ASE100 and ASE100-0.1GN, while the size of cells exhibits a significant difference. Compared with the ASE100 (around 1.2 μm), the cells in ASE100-0.1GN are finer and the average diameter is around 0.7 μm . During the fabrication, GNs are homogenized and dispersed within the matrix owing to the efficient action of Marangoni flow within the molten pool^[7]. The presence of dispersed GNs in the molten pool effectively refines the cells during subsequent solidification. Additionally, some GNs react directly with alloy powder to form a substantial number of nano-scaled carbides during solidification, as indicated by the white circle in Fig. 4(e). To better identify the difference in the morphology, distribution, and size of nano-scaled carbides in ASE100 and ASE100-0.1GN alloys, the higher-magnification observation was conducted to reveal the microstructure of nano-carbides, as seen in Figs. 4(c, f). Compared with the nanoparticles precipitated during the solidification process in the ASE100, the ASE100-0.1GN reveals a greater number of nanoparticles with uniform distribution induced by the in-situ reaction.

Figure 5 shows the EBSD analysis of the ASE100 and

ASE100-0.1GN in the as-built state. Herein, Figs. 5(a) and (c) correspond to the inverse pole figure with grain boundaries along the building direction of the ASE100 and ASE100-0.1GN. In the plane perpendicular to the building direction, the grains in both the ASE100 and ASE100-0.1GN exhibit a classical columnar morphology. The introduction of GNs has little impact on the orientation of column grains. However, a large number of fine grains appear in the ASE100-0.1GN, indicating that the addition of GNs can effectively promote the columnar-to-equiaxed transition (CET) of grains^[29]. The statistical results of grain size distribution in ASE100 and ASE100-0.1GN are shown in Figs. 5(b) and (d), respectively. Compared to the ASE100, ASE100-0.1GN contains more grains with equivalent circular diameters less than 50 μm , which corresponds to the results of the CET transition in Fig. 5(c). The average grain size of ASE100 and ASE100-0.1GN is 25.29 μm and 17.76 μm , respectively.

STEM and EDS mapping were performed to investigate the difference in microstructure between ASE100 and ASE100-0.1GN. As seen in Figs. 6(a, b), Ti is segregated around the boundary of the cellular structure in both ASE100 and ASE100-0.1GN and Co is segregated within the cellular

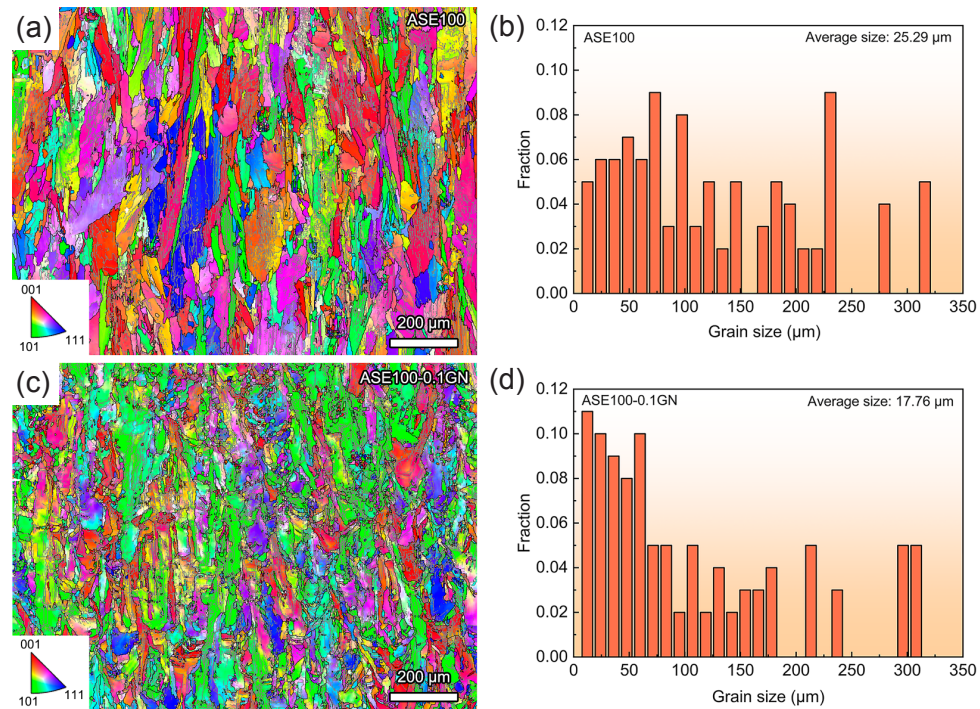


Fig. 5: Inverse pole figures with grain boundary of as-built ASE100 (a) and ASE100-0.1GN (c), and grain size distribution of as-built ASE100 (b) and ASE100-0.1GN (d)

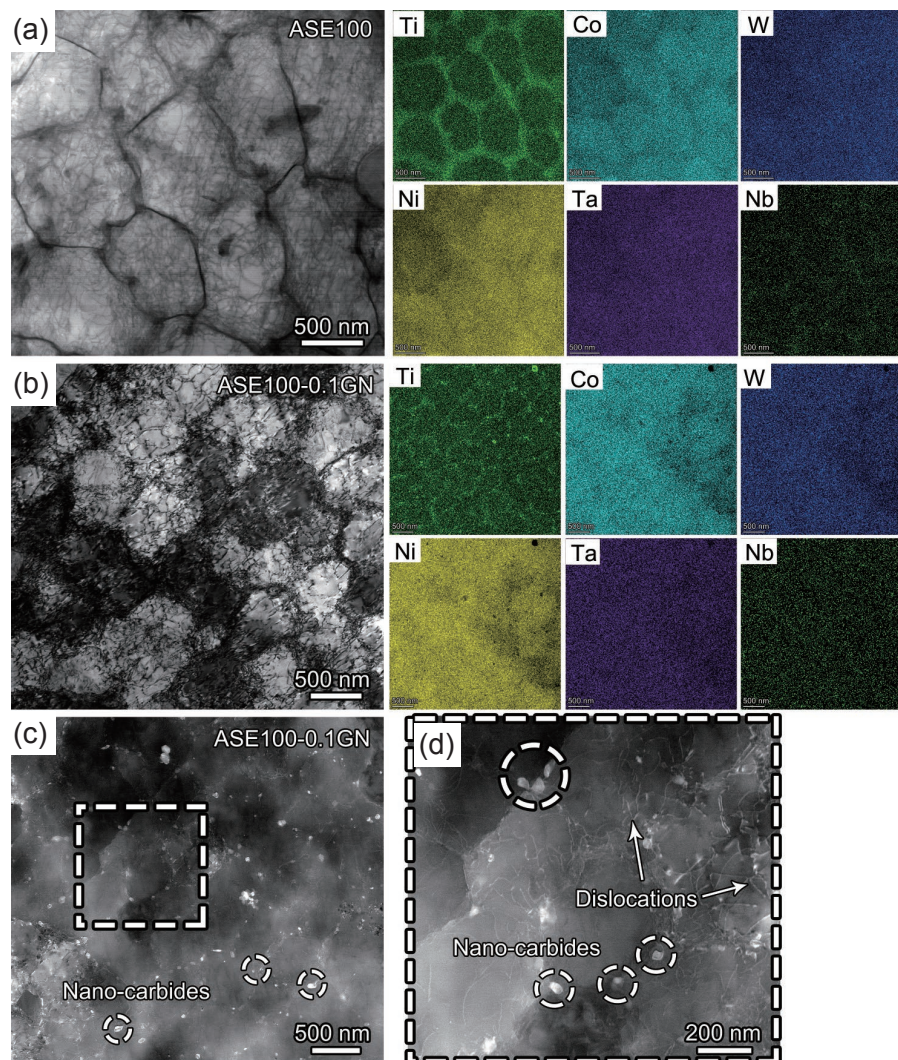


Fig. 6: STEM and EDS mapping results of as-built ASE100 (a) and ASE100-0.1GN (b-d)

structure. However, the degree of element segregation in ASE100-0.1GN is significantly lower than ASE100. Evidently, the in-situ reaction of the manufacturing process in ASE100-0.1GN consumes the Ti and refines the cells, thus reducing the element segregation of the matrix. These reaction-induced nanoparticles are identified as Ti-rich carbides and mainly distributed along the cell boundaries, which may inhibit the movement of dislocations, as seen in the HAADF images in Figs. 6(c, d). Meanwhile, a high density of dislocation tangles is observed along the cell boundaries in both the ASE100 and ASE100-0.1GN. It can be attributed to the ultrafast cooling rates and ultrahigh temperature gradients^[30]. The ASE100 powder and composite powder melt rapidly and subsequently solidify under the radiation of a high-energy laser, which introduces a high degree of residual stresses and leads to the entanglement of dislocations^[31]. The presence of finer cells with more nano-carbides enables the ASE100-0.1GN to accommodate more dislocations than the ASE100.

Small-angle neutron scattering (SANS) was conducted to quantitatively analyze the volume fraction and distribution of in-situ reaction-induced nano-carbides and residual GNs in the ASE100-0.1GN. Figure 7 exhibits the SANS scattering pattern with fitting curves and size polydispersity of nanoparticles in ASE100-0.1GN. The fitting of size and distribution is based on the structural characteristics of nano-carbides (sphere-like) and

GNs (lamellar-like), the equations are as follows:

Sphere-like:

$$I(q) = \frac{\phi}{V} \left[3V(\Delta\rho) \cdot \frac{\sin(qr) - qr \cos(qr)}{(qr)^3} \right]^2 + background \quad (1)$$

where ϕ is the volume fraction, V is the volume of the scatterer, r is the radius of the nano-carbides, $background$ is the background level, and $\Delta\rho$ is the difference of scattering length density (SLDs) of solute and solvent scatterers.

Lamellar-like:

$$I(q) = scale \frac{2\pi P(q)}{q^2 \delta} + background \quad (2)$$

$$P(q) = \frac{2\Delta\rho^2}{q^2} (1 - \cos(q\delta)) = \frac{4\Delta\rho^2}{q^2} \sin^2\left(\frac{q\delta}{2}\right) \quad (3)$$

where $P(q)$ is the form factor, the δ is total layer thickness, and $\Delta\rho$ is the scattering length density difference between matrix and GNs.

Based on the fitting results by SANS, the average thickness of the residual GNs is 31.73 ± 1.32 nm, while the volume fraction is $(0.71 \pm 0.03)\%$. As for nano-carbides, the average radius is 24.16 ± 1.43 nm and the volume fraction is $(0.10 \pm 0.02)\%$. Both residual GNs and nano-carbides exhibit uniform distribution in size according to the SANS pattern in Figs. 7(b, c), thereby enabling the nanoparticles with this dispersion to fully exert the synergistic strengthening effect.

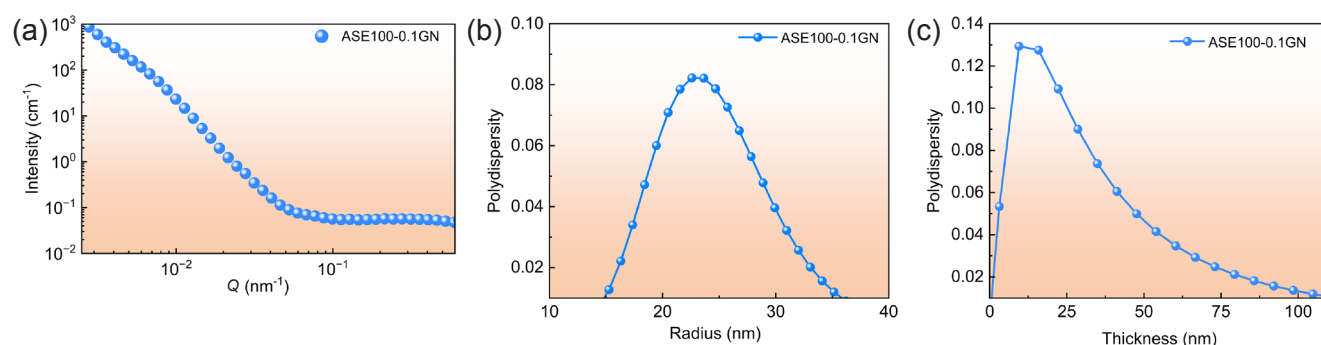


Fig. 7: SANS results of as-built ASE100-0.1GN: (a) SANS patterns of samples; (b) size distribution of nano-carbides; (c) size distribution of residual GNs

3.2 Thermophysical properties and in-situ reaction mechanism

To further enhance the comprehension of the in-situ reaction mechanism and thermophysical properties of different alloys, DSC was performed on the ASE100 and ASE100-0.1GN. Figures 8(a) and (b) show the heating curves from 800 °C to 1,450 °C of the ASE100 and ASE100-0.1GN, respectively. When the temperature is increased to 1,000 °C, a heat absorption peak appears in both the ASE100 ($\sim 1,005$ °C) and ASE100-0.1GN ($\sim 1,001$ °C), which is identified as the dissolution temperature of the γ' phase. With the further increase in temperature, the alloys gradually approach their melting point and begin to melt. It is worth noting that the liquidus of ASE100-0.1GN is approximately 4 °C higher than that of ASE100. Moreover, there is a strange

exothermic peak ($\sim 1,339$ °C) near the liquid phase line of the ASE100-0.1GN, which is not detected in the ASE100. As the temperature exceeds the liquidus, ASE100-0.1GN exhibits a heat absorption peak (1,379 °C) that is not present in ASE100. Combined with the SANS results in Fig. 7 and DSC results, the heat absorption peak at 1,379 °C in ASE100-0.1GN can be identified as the solution of residual GNs. When the temperature gradually reaches the liquidus temperature of the alloy, the matrix begins to melt. The remaining GNs have a higher melting point, therefore suspend in the molten metal^[7]. At this point, the remaining GNs react again with the molten metal to form nano-carbides due to the relatively slow heating rate (10 °C $\cdot$ min⁻¹). This not only reflects the formation mechanism of in-situ reaction-induced nano-carbides during the manufacturing process, but also explains the additional exothermic peak of the ASE100-0.1GN near the liquidus.

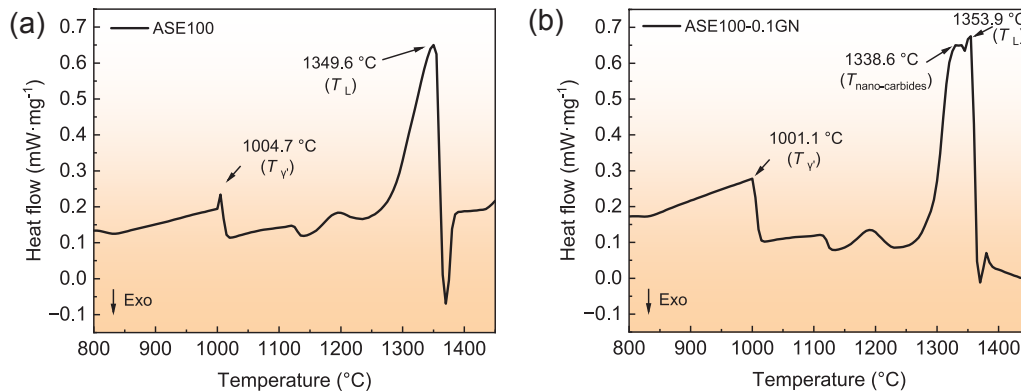


Fig. 8: DSC results of as-built ASE100 (a) and as-built ASE100-0.1GN (b)

3.3 Mechanical properties and strengthening mechanism

To examine the synergistic strengthening effect of GNs and in-situ reaction-induced nano-carbides, the hardness and tensile tests were performed on the ASE100 and ASE100-0.1GN, and the results are shown in Fig. 9. Apparently, the introduction of GNs enhances the hardness both in the *XOY* plane and the *XOZ* plane, as seen in Fig. 9(a). The hardness of the ASE100 in the *XOY* plane and the *XOZ* plane is 317 ± 4 HV and 315 ± 10 HV, respectively. The hardness of ASE100-0.1GN in the *XOY* plane and the *XOZ* plane is 329 ± 11 HV and 346 ± 10 HV, respectively. During the fabrication, the in-situ reaction-induced nano-carbides exist at cell boundaries, while the residual GNs distribute in the matrix^[20]. Thus, the hardness of the ASE100-0.1GN in the *XOY* plane and the *XOZ* plane increases by 4% and 10%, respectively, under the synergistic effect of multiple nanoparticles. Notably, the hardness of the ASE100 in different planes is close, whereas the hardness of ASE100-0.1GN in the *XOZ* plane is higher than in the *XOY* plane. This can be attributed to the CET in the *XOZ* plane after the introduction of GNs, which corresponds to the EBSD results in Figs. 5(a) and (c). In addition, the yield tensile strength (YTS) of ASE100-0.1GN is significantly increased compared with that of ASE100. The introduction of GNs provides an increase in YTS of about 86 MPa (from 756 ± 7 MPa to 842 ± 40 MPa). The ultimate tensile strength (UTS) of ASE100-0.1GN is maintained at almost the same level as that of ASE100 (around $1,073 \pm 5$ MPa of ASE100-0.1GN

and around $1,068 \pm 3$ MPa of ASE100), whereas the elongation slightly decreases. On the one hand, GNs refine the grains of the alloy, thus boosting the YTS of ASE100-0.1GN based on the Hall-patch relationship^[32]; on the other hand, the in-situ reaction-induced nano-carbides exert a pinning effect on the movement of dislocations. Therefore, the strength and hardness of ASE100-0.1GN alloy are significantly enhanced with a slight sacrifice in plasticity.

Figures 10(a-d) exhibit the fracture surface of the ASE100 and ASE100-0.1GN. Voids and cracks are observed in both the ASE100 and ASE100-0.1GN according to Figs. 10(a, c). Cracks are generally initiated from voids and gradually propagate until rupture. However, although the ASE100-0.1GN exhibits a higher YTS than ASE100, it contains more coarse voids [Figs. 10(c, d)]. This may explain the lack of a significant increase in UTS and a slight decrease in plasticity. Further characterization was conducted on the longitudinal section of the fracture under STEM, as shown in Figs. 10(e, f). Nano-carbides are uniformly distributed along the cell boundaries and within the sub-grains, which may refine the cells and inhibit the movement of dislocations. The aggregation of dislocations around the nano-carbides in Fig. 10(f) further confirms the contribution of nano-carbides to the enhanced strength and hardness of ASE100-0.1GN. In addition, Chen et al.^[20] and Gao et al.^[28] confirmed that the addition of GNs can enhance the strength of superalloys without sacrificing the plasticity. In this work, the high strength residual GNs along the crack propagation path can inhibit the further expansion of cracks. However, the

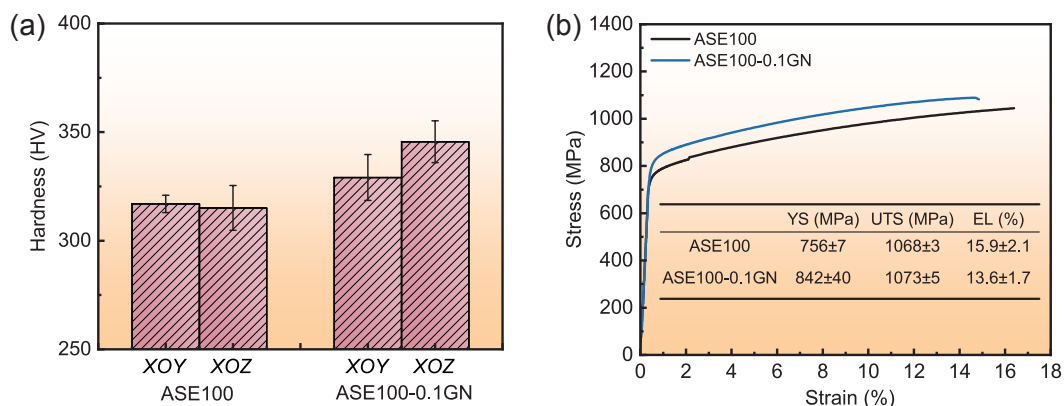


Fig. 9: Vickers microhardness (a) and room temperature mechanical properties (b) of ASE100 and ASE100-0.1GN

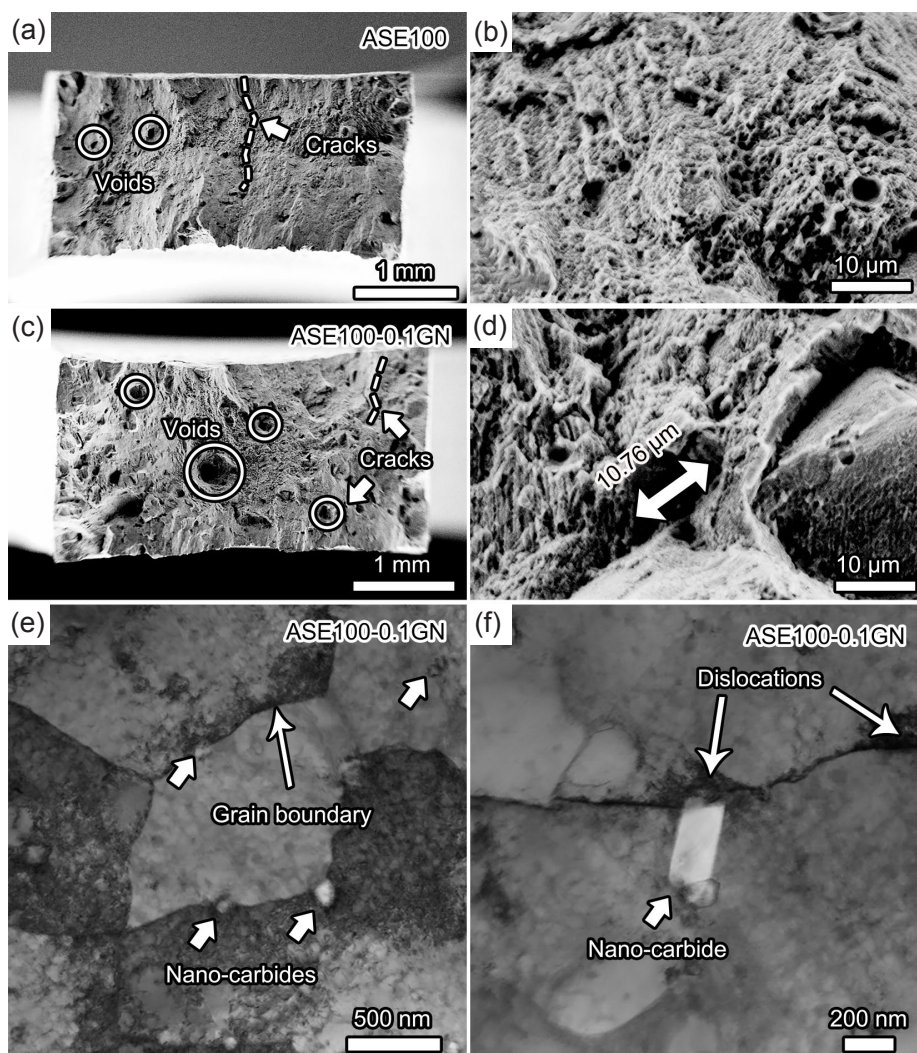


Fig. 10: Fracture surfaces of ASE100 (a, b) and ASE100-0.1GN (c, d), and nano-carbides in ASE100-0.1GN (e, f) by STEM analysis after room temperature tensile tests

large-sized void defects present in the ASE100-0.1GN alloy are more significant than those observed in the ASE100 alloy under the influence of process parameters. These large voids may act as the sites for crack initiation and deteriorate the performance (especially ductility). Further enhancement may be achieved by optimization of the process parameters and the microstructure of the ASE100-0.1GN.

As illustrated in Fig. 11, the EBSD analysis results of the longitudinal section after fracture of both ASE100 and ASE100-0.1GN provide a quantitative and qualitative analysis of the mechanical behavior under strain. The evolution of grain orientation after severe plastic deformation varies insignificantly among different alloys [Figs. 11(a, d)]. The Kernel average misorientation (KAM) map, a map containing the average misorientation between every point and its surrounding points, is typically used to reflect the strain concentration distribution inside the grains. Figures 11(b, e) display the KAM results of ASE100 and ASE100-0.1GN. The maximum values of KAM in the two alloys are similar, whereas the accumulation of strain in ASE100-0.1GN is more uniform in the plane compared to ASE100. Generally speaking, the strain accumulation distribution can reflect the

dislocation density within the alloys. As mentioned above, the refined cellular structure with more nano-carbides enables the ASE100-0.1GN to accommodate more dislocations than the ASE100 [corresponding to Figs. 10(e, f)]. The statistical results of grain size distribution in the fracture surface of ASE100 and ASE100-0.1GN show that the distribution of grain size of ASE100-0.1GN is more uniform than ASE100. The average grain size of ASE100 and ASE100-0.1GN is 17.72 μm and 14.47 μm , respectively. Under the combined effect of fine grain strengthening and dispersion strengthening provided by GNs and nano-carbides, the yield strength of ASE100-0.1GN is significantly enhanced.

4 Conclusions

In summary, a novel Ni-based superalloy named ASE100-0.1GN was designed based on the in-situ reaction between GNs and alloy powder during LPBF. Compared with the original alloy (ASE100), the novel alloy possesses an enhanced strength and hardness due to the refined microstructure and the dispersed nanoparticles. The main conclusions are as follows:

(1) The grain morphology in the as-built ASE100-0.1GN

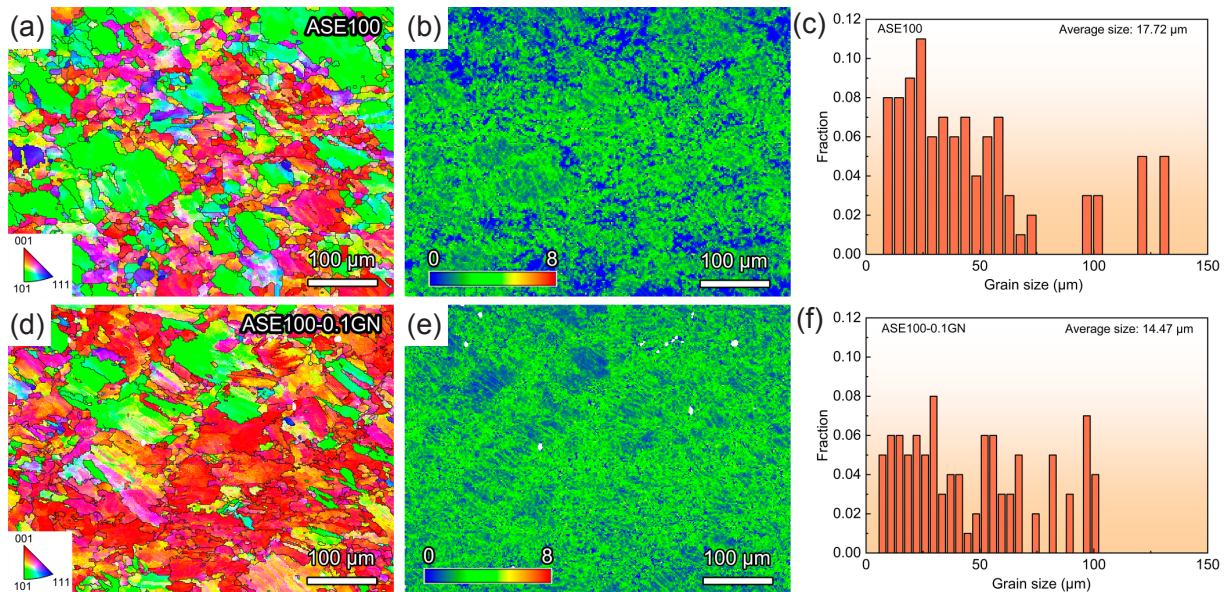


Fig. 11: EBSD analysis of longitudinal section after room temperature tensile tests: (a) IPF with grain boundaries of ASE100; (b) KAM map of ASE100; (c) grain size distribution of ASE100; (d) IPF with grain boundaries of ASE100-0.1GN; (e) KAM map of ASE100-0.1GN; and (f) grain size distribution of ASE100-0.1GN

exhibits a distinct tendency toward CET compared with the as-built ASE100. Finer cellular structures are observed in the ASE100-0.1GN.

(2) In-situ formed nano-carbides are uniformly distributed in the matrix of the as-built ASE100-0.1GN. Residual GNs are also detected by the SANS and DSC in the as-built ASE100-0.1GN.

(3) Both the hardness and YTS (346 HV and 842 MPa) of the as-built ASE100-0.1GN are increased compared with those of the as-built ASE100 (315 HV and 756 MPa). The synergistic effect of the in-situ formed nano-carbides and the residual GNs is primarily responsible for the enhancement of hardness and strength.

(4) The fracture surface of the as-built ASE100-0.1GN exhibits more coarse voids compared to the as-built ASE100. These coarse voids may restrict the strengthening effect of the nanoparticles. The optimization of process parameters to reduce the defect density in ASE100-0.1GN has significant potential for further enhancing the strength and ductility.

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Conflict of interest

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

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