

Effects of laser scanning conditions on high-temperature tensile properties of IN738LC fabricated by laser powder bed fusion

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ABSTRACT

Recent studies have demonstrated that crack-free components of IN738LC, a nickel-based superalloy susceptible to solidification cracking, can be produced by laser powder bed fusion (L-PBF) through optimization of laser scanning parameters. Building on this progress, this study clarifies the process–microstructure–property relationships of IN738LC fabricated by L-PBF without hot isostatic pressing (HIP) by examining its tensile behavior and fracture morphology over a broad temperature range (200–1000 °C) under loading parallel to the build direction. Samples fabricated under conventional and optimized scanning conditions were directly compared in the as-built state, and the crack-free optimized material was further evaluated after standard heat treatment. The resulting mechanical responses were benchmarked against those of precision-cast IN738LC within the same analytical framework. In the as-built state, the optimized condition produced a directional-solidification-like microstructure with uniformly distributed fine carbides, resulting in high-temperature ductility with elongations exceeding 10%. In contrast, the conventional condition produced V-shaped grain boundaries that led to early failure within the elastic regime at tensile temperatures of 800 °C and above. Standard heat treatment of the optimized material promoted the formation of fine γ' precipitation with a bimodal size distribution while retaining the directional-solidification-like microstructure, leading to higher yield and ultimate tensile strengths up to 600 °C than those of precision-cast IN738LC. Consequently, both the as-built and heat-treated specimens fabricated with optimized scanning parameters showed tensile properties comparable to or better than those of precision-cast material.

1. Introduction

Nickel-based superalloys are valuable as energy-related structural materials because of their outstanding properties under high-temperature environments, and have been widely used as components in jet engines for aircraft and gas turbines for power generation equipment. Since raising the turbine inlet temperature can improve the efficiency of gas turbine engines and reduce CO₂ emissions, manufacturers and research institutes have intensively developed novel nickel-based superalloys with superior properties since the mid-20th century [1–3]. Among such a wide variety of superalloys, IN738LC, which is the model material for this study, was designed in the 1960s to provide the gas turbine industry with a suitable vacuum casting alloy [4]. Due to the strengthening by precipitation of γ' of 50%, the superalloy exhibits excellent high-temperature creep-rupture strength up to 980 °C combined with the ability to withstand long-time exposure to the hot

corrosive environments associated with the engine.

Although IN738LC is a classical superalloy designed for the casting process, fundamental research on its processability and material properties has been conducted in the field of additive manufacturing (AM) for the past decade. Laser powder bed fusion (L-PBF), which is one of the AM processes, alternates between spreading fine metal powder onto a build platform and scanning a high-power laser over it to selectively melt and solidify the powder according to the sliced data derived from the product design. Since there are few restrictions on designed geometry in L-PBF, it is expected to achieve topology optimization and part consolidation of conventional components used in the aerospace industry. On the other hand, practical application of AM is still constrained by processing difficulties for nickel-based superalloys with high γ' content and by limited datasets for high-temperature mechanical properties. In components built by L-PBF, solidification cracking is likely to occur due to tensile force applied to the mushy zone during solidification

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after laser scanning [5–7]. Since IN738LC is particularly susceptible to solidification cracking, the specimens fabricated by L-PBF were usually post-processed with hot isostatic pressing (HIP) before the high-temperature tensile test [8–14]. According to the studies by Rickenbacher et al. [8] and Chen et al. [14], the HIPed IN738LC specimen achieved strength and ductility comparable to those of precision-cast material when loaded along the build direction, even at temperatures above 800 °C (e.g., 942 MPa yield strength and 14.3% in elongation at 800 °C [14]). It should be noted that HIP not only reduces internal defects such as cracks, but also partially coarsens the as-built grains [13], or the microstructure may completely recrystallize during the HIP process at 1200 °C [12]. In contrast, there are limited studies on the high-temperature tensile properties of non-HIPed IN738LC specimens fabricated by L-PBF. Xu et al. [12] reported that, in the absence of HIP, the standard heat-treated specimen showed $0.6 \pm 0.3\%$ elongation at 800 °C, while the as-built specimen fractured in the elastic regime.

As an alternative approach to post-processing by HIP, several studies have investigated the fabrication of crack-free IN738LC by L-PBF [15–19]. The primary parameters regarding the occurrence of solidification cracking are laser power P , scanning velocity v , hatch spacing h , scanning strategy, and the preheating temperature. While Grange et al. [15] concluded the crack density decreased with lower linear energy density $E_l (= P/v)$ and narrower h , Hafezi et al. [16] found that medium volumetric energy density $E (= P/vht; \text{ powder layer thickness } t)$ around 90 J/mm^3 and faster v resulted in higher relative densities and lower defects. According to the study on the scanning strategy by Xu et al. [17], the as-built IN738LC sample with the rotation angle R of 67° between layers had higher density than those with the rotation angle of 0° and 90° . In the previous research with data-driven approach [18], more than 100 cubic samples of IN738LC were fabricated with different laser scanning conditions, and the process window for crack density and defect ratio was created based on their microscopic observations. As a result, the optimized laser scanning conditions were identified to suppress the crack density to less than 0.005 mm^{-1} . Chen et al. [19] suppressed cracking by raising the build platform temperature from 200 to 700 °C during fabrication. The as-built IN738LC exhibited a strongly textured columnar microstructure with finely dispersed carbides along grain boundaries, showing an ultimate tensile strength (UTS) of 1612 MPa and an elongation of 18% at room temperature. According to Jena et al. [20], crack-free IN738LC specimens fabricated at $E = 108 \text{ J/mm}^3$ and subjected only to the standard heat treatment (without HIP) achieved excellent high-temperature tensile properties, with UTSs of 765 MPa and 538 MPa, and elongations of 17% and 14% at 850 and 927 °C, respectively.

Beyond their influence on solidification cracking, the laser scanning conditions also significantly affected the as-built microstructure. In the previous study [18], the grain morphology in the XY plane (normal to the build direction) changed from a typical checkerboard-pattern reflecting the laser scanning path to equiaxed-like fine grains with serrated grain boundaries when hatch spacing h was narrowed from 100 μm to 50–70 μm . In the XZ plane, the microstructure consisted of columnar grains growing along the build direction, with $\langle 001 \rangle$ orientation strongly aligned with the build direction under the crack-free conditions. These features are similar to the microstructures by unidirectional solidification in casting and were also observed in other crack-free IN738LC samples [14,19,20]. These previous studies established the feasibility of fabricating crack-free IN738LC by L-PBF and demonstrated promising tensile properties without HIP.

Building upon these findings, the present study examines the tensile and fracture behavior of non-HIP L-PBF-fabricated IN738LC over a broad temperature range using samples with distinct scan-parameter-induced as-built microstructures. Two samples fabricated under conventional and optimized laser scanning conditions were directly compared in the as-built state. In addition, the crack-free material fabricated under the optimized condition was evaluated after standard heat treatment. The cross sections and fracture surfaces of the tensile-

tested specimens were observed microscopically to assess the effects of as-built microstructures and γ' precipitates on tensile behavior. These results were further compared with those of precision-cast IN738LC within the same analytical framework. Thus, this study extends previous reports on crack-free, non-HIP L-PBF-fabricated IN738LC by clarifying the process–microstructure–property relationships through systematic comparisons of conventional and optimized as-built samples, the heat-treated optimized sample, and precision-cast material over a broad temperature range.

2. Materials and methods

As shown in Fig. 1 (a), rectangular samples with dimensions of 15 mm $\times$ 10 mm $\times$ 130 mm were fabricated by an L-PBF machine (SLM 280; SLM Solutions GmbH) with gas-atomized IN738LC powder (AMPERPRINT®0151.074; Höganäs AB) as a feedstock material. The median particle diameter (D50) was 27.92 μm , and the chemical composition of the raw metal powder is listed in Table 1. The powder was spread over the platform, and the stainless-steel baseplate was lowered by 30 μm after each layer was scanned by the laser with 1070 nm in wavelength and 80 μm in beam diameter. Two types of samples were prepared using either conventional or optimized laser scanning conditions. The parameters referred to here as “conventional” were laser power $P = 300 \text{ W}$, scan velocity $v = 1300 \text{ mm/s}$, hatching space $h = 100 \mu\text{m}$, and rotation angle $R = 90^\circ$, which are similar to the conditions used in the previous studies on IN738LC [21] and Hastelloy X [22] and were adopted as the reference condition in the present study. In contrast, the optimized parameters to minimize solidification cracking were $P = 150 \text{ W}$, $v = 1000 \text{ mm/s}$, $h = 50 \mu\text{m}$, and $R = 67^\circ$, as reported in the previous study [18]. The build chamber was purged with argon gas to maintain an oxygen concentration below 0.01 vol% during the fabrication.

Then, the rectangular samples were removed from the baseplate, and some of them were subjected to a standard heat treatment (SHT) recommended for cast IN 738LC: solution treatment at 1120 °C for 2 h followed by air cooling, and subsequent aging at 850 °C for 24 h followed by air cooling [4]. After the solution treatment, the sample fabricated with the conventional condition exhibited severe cracking, over 10 mm in length and oriented parallel to the XZ or YZ plane (see Fig. 1(b and c)). Thus, we prepared two forms for the optimized condition (as-built and standard heat-treated) and only the as-built form for the conventional condition. Hereafter, we refer to these as Opt-AB, Opt-SHT, and Conv-AB, respectively. Accordingly, the effect of the laser scanning condition was evaluated by comparing Conv-AB and Opt-AB, whereas the effect of SHT on the optimized sample was evaluated by comparing Opt-AB with Opt-SHT. The rectangular samples were then machined into tensile specimens so that they would be loaded along the build direction, as illustrated in Fig. 1 (d).

High-temperature tensile tests were conducted using a universal testing machine (INSTRON 5982) in accordance with ASTM E21-20 in air at six temperatures: 200, 400, 600, 800, 900, and 1000 °C. One tensile specimen was tested for each material condition and temperature. The temperatures were monitored with R-type thermocouples at the top, middle, and bottom of the gauge section. Specimens were heated to the test temperature within 30 min and then held at that temperature until fracture (typically $\sim 30 \text{ min}$). The applied strain rate was 0.5%/min up to the yield point, and then increased to 5%/min until fracture. After testing, fracture surfaces and cross-sections were observed using an optical microscope (VHX-2000; KEYENCE Corporation), scanning electron microscopes (SEMs) (JSM-6010LA and JSM-7200F; JEOL) equipped with electron backscatter diffraction (EBSD). The accelerating voltage was set to 20 kV and the scanning step size was 1.5 μm . Here, the misorientation angle of HAGBs was defined as above 15° . The γ' precipitate size distribution and area fraction were evaluated from each SEM image. Similarly, the grain size distribution was also obtained from each EBSD map. The mean values and standard

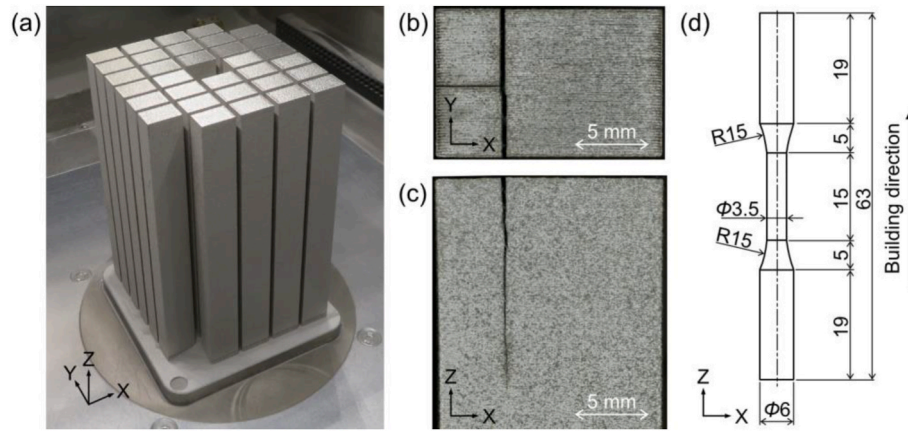


Fig. 1. (a) As-fabricated rectangular samples on the baseplate, (b) top and (c) side views of the sample fabricated with the conventional laser scanning condition followed by the heat treatment at 1120 °C, and (d) dimensions of tensile specimen.

Table 1

Chemical composition of AMPERPRINT 0151.074 (wt. %, Ni = balance).

Co	Cr	Mo	W	Al	Ti	Nb	Ta	C	B	Zr	N	O	Fe	Mn	Si	P	S
8.5	15.9	1.7	2.5	3.5	3.5	0.88	1.8	0.10	0.007	0.024	0.008	0.017	0.02	0.01	0.02	0.005	0.002

deviations of γ' precipitate size and grain size were calculated from the respective distributions. The hardness was measured with a microhardness tester (AMT-7FS; Matsuzawa Co., Ltd.) under a load of 1.962 N and a dwell time of 15 s. For each sample condition, more than five indentations were performed, and the mean value and standard deviation were calculated.

In addition, these AM results were compared with precision-cast IN738LC reported in a creep datasheet [23]. The cast specimen was produced by vacuum casting with a ceramic mold and subjected to HIP at 1176 °C/100 MPa for 2 h, followed by heat treatment at 1176 °C/2 h/gas fan cooling (GFC) + 843 °C/24 h/GFC. Refer to Ref. [23] for details of the casting and heat-treatment procedures, the alloy composition, and the tensile testing conditions. As with the AM specimens, the fracture surface and cross section were observed by microscopy.

3. Results

3.1. Microstructure of as-built and heat-treated samples

Figs. 2 and 3 show microscopic images of XY and YZ cross-sections of Conv-AB and Opt-AB, respectively. The optical micrographs in Fig. 2 (a, d) and 3 (a, d) were binarized using the image analysis software Fiji [24] to highlight internal defects within the sample. As shown in Fig. 2(a–d), numerous microcracks were observed in Conv-AB, particularly near the edges. Such concentration of cracks near the edges may be attributable to thermal accumulation for turning around of the laser scan at the sample boundaries. A more detailed SEM observation of the regions indicated by blue rectangles in Fig. 2(a, d) revealed cracks more than 50 μm in size, as shown in Fig. 2(b, c, e, f). In particular, the dendritic structures observed on the fracture surface in Fig. 2(f) suggest that the cracks were formed as a result of solidification cracking [5,18]. In contrast, such microcracks were rarely observed in Opt-AB as shown in

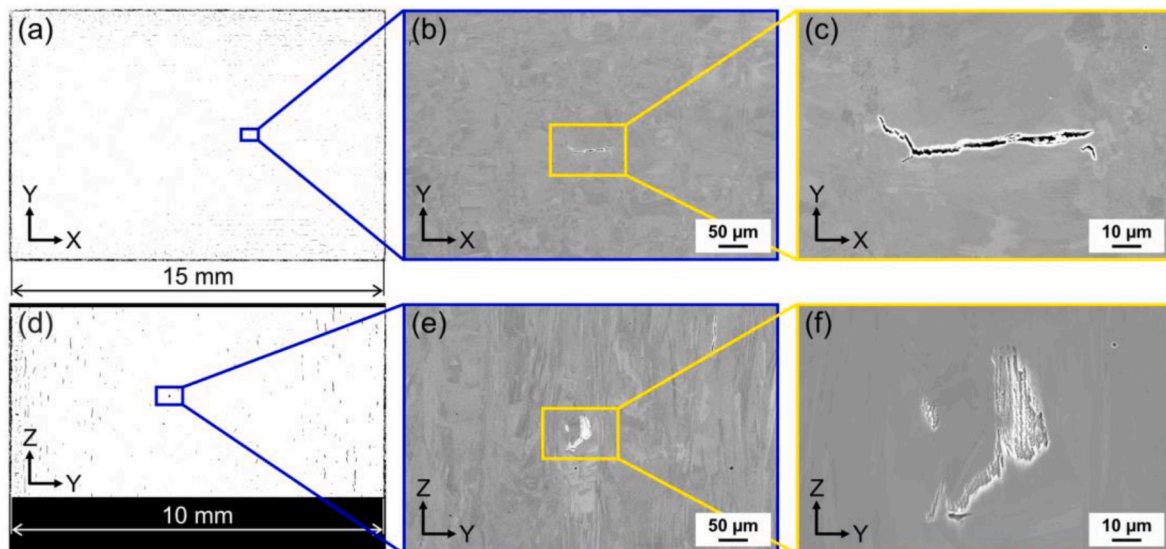


Fig. 2. (a, d) Optical micrographs and (b, c, e, f) SEM secondary electron images of Conv-AB. (a–c) XY plane; (d–f) YZ plane.

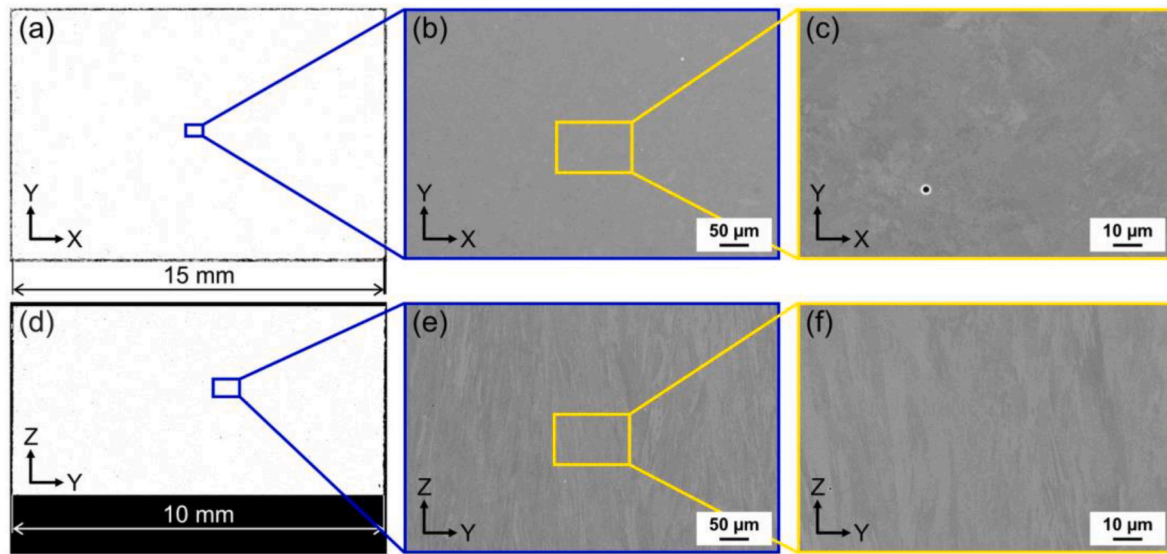


Fig. 3. (a, d) Optical micrographs and (b, c, e, f) SEM secondary electron images of Opt-AB. (a–c) XY plane; (d–f) YZ plane.

Fig. 3(a–d). Even under detailed SEM observation in Fig. 3 (b, c, e, f), only spherical pores smaller than 10 μm were found inside the sample. Using the same procedure as described in the previous study [18], the crack density, defined as the total crack length per unit observed cross-sectional area, was evaluated in the region located 1 mm inward from the sample edges in Figs. 2 (a) and 3 (a). The crack densities of Conv-AB and Opt-AB were 0.277 mm^{-1} and 0.007 mm^{-1} , respectively.

Fig. 4(a–c) shows higher-magnification SEM secondary electron images of the sub-grain structures of Conv-AB, Opt-AB, and Opt-SHT, respectively. As shown in Fig. 4(a and b), cellular substructures were present in both as-built samples. The average cell width over 35 substructures, corresponding to the primary dendrite arm spacing (PDAS), was $0.58 \pm 0.21 \mu\text{m}$ for Conv-AB and $0.54 \pm 0.20 \mu\text{m}$ for Opt-AB, indicating substantial overlap within the experimental scatter and thus comparable spacing. According to previous studies [25,26], PDAS and the cooling rate GR are related by the empirical relation $PDAS = a(GR)^n$, where a was 50 μm , and n was $1/3$ for nickel-based alloys including IN738LC. Using this equation, the cooling rates for Conv-AB and Opt-AB were estimated to $6.4 \times 10^5 \text{ K/s}$ and $7.9 \times 10^5 \text{ K/s}$, respectively. However, considering the scatter in PDAS, both conditions should be regarded as having comparable cooling rates on the order of 10^5 K/s .

As indicated by the white arrows in Fig. 4(a and b), white particles were found in the cellular boundaries in both Conv-AB and Opt-AB, and are considered to be MC carbides, as reported in previous studies [12, 27]. Such white particles are also apparent in the backscattered electron (BSE) image in Fig. 5 (a). As shown in Fig. 5(b and c), EDS maps showed slight enrichment of Ti, Nb, and Mo along the cellular boundaries in the

as-built samples. In contrast, as shown in Fig. 4 (c), the cellular substructures disappeared and precipitated γ' particles were visible after the heat treatment. The γ' precipitates showed a bimodal morphology, consisting of cuboidal secondary and spherical tertiary particles, with average sizes of $441 \pm 187 \text{ nm}$ and $109 \pm 28 \text{ nm}$, respectively. These values, together with the γ' area fraction, are listed in Tables 2 and 3. The white spherical particles were also present in Opt-SHT (see the white arrows in Figs. 4 (c) and 6 (a, b)) and were relatively larger than those in Opt-AB. As shown in Fig. 6(e and f), these particles were clearly enriched in Ti and Nb. In addition, as indicated by black arrows in Fig. 6 (b), dark-gray curved precipitates appeared along grain boundaries, and were identified as Cr-rich compounds by EDS (see Fig. 6 (c)). Mn, Mo, and O were also detected at the same locations. According to the EDS maps of Al and Ti in Fig. 6(d and e), coarse γ' particles were also present at grain boundaries.

Fig. 7 shows inverse pole figure referenced to the Z direction (IPF-Z) and grain reference orientation deviation (GROD) maps of the cross-sections analyzed by EBSD. Hereafter, the black lines in the IPF map represent HAGBs, whereas gray scale map of image quality (IQ) was overlaid on the GROD maps. As shown in Fig. 7 (a), Conv-AB exhibited a checkerboard-like texture on the XY cross-section, where each square region was separated by grains oriented in the $\langle 001 \rangle$ direction along the build direction. The arrangement and spacing of the checkerboard pattern were consistent with the laser scan path directions along the X and Y axes, as well as the hatch spacing of 100 μm . On the YZ plane, as shown in Fig. 7 (e), a microstructure composed of columnar grains with non-uniform size, morphology, and crystallographic orientation was observed, including several V-shaped grains as indicated by black

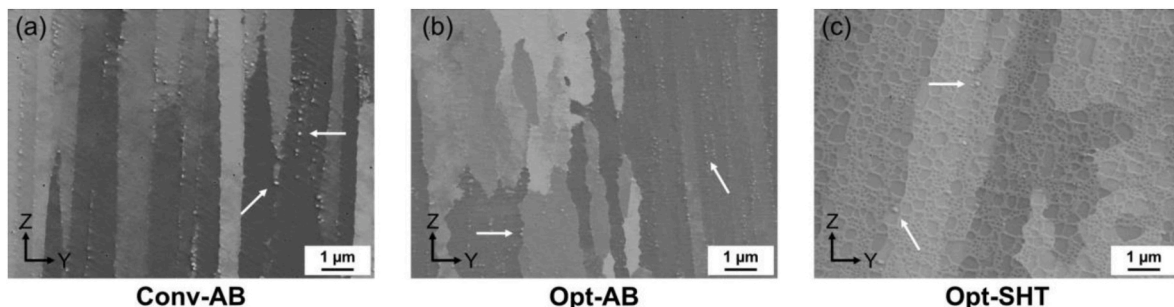


Fig. 4. High-magnification SEM secondary electron images of (a) Conv-AB, (b) Opt-AB, and (c) Opt-SHT.

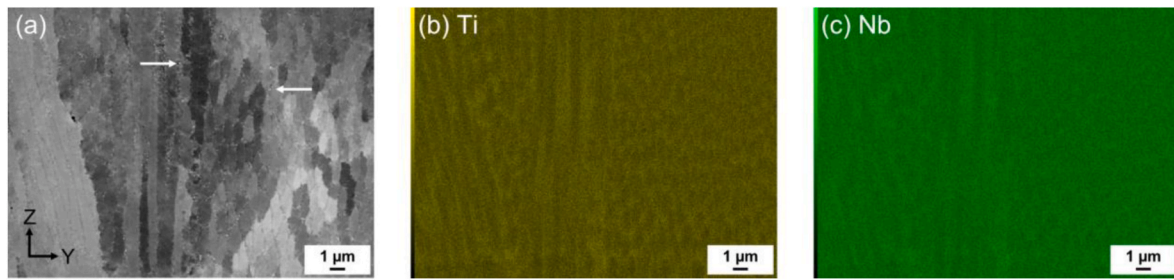


Fig. 5. (a) High-magnification backscattered-electron SEM image and EDS element maps of (b) Ti and (c) Nb for Conv-AB.

Table 2

γ' fraction of Opt-SHT and precision-cast materials.

Sample condition	Secondary γ'	Tertiary γ'	Total
Opt-SHT	0.375	0.146	0.521
Precision-cast	0.461	-	> 0.461

Table 3

γ' size of Opt-SHT and precision-cast materials.

Sample condition	Secondary γ' , nm	Tertiary γ' , nm
Opt-SHT	441 $\pm$ 187	109 $\pm$ 28
Precision-cast	301 $\pm$ 78	-

arrows in the figure. In contrast, as shown in Fig. 7(c–g), Opt-AB had a strong (001) crystallographic texture throughout the entire microstructure. On the XY plane in Fig. 7(c), the microstructure consisted of fine grains divided by serrated boundaries, while on the YZ plane in Fig. 7(g), columnar grains elongated along the build direction were observed. For Opt-SHT, the grain morphology and crystallographic orientation were identical to those of Opt-AB in Fig. 7(c, d, g, h). Table 4 lists the area-weighted average grain diameter of these samples.

In the GROD maps of Fig. 7(b–d, f, h), the value at each point represents the misorientation relative to the mean orientation of the grain. In the figure, defects such as pores and cracks appear black due to low IQ values (see the white arrows in Fig. 7(b)). As confirmed by the corresponding IPF map in Fig. 7(a), the cracks exist along HAGBs. Compared with Conv-AB in Fig. 7(b–f), Opt-AB shows more regions of higher GROD

(see Fig. 7(d–h)). The high GROD is considered to arise from plastic strain generated by local solidification shrinkage during L-PBF. Such overall high GROD was also observed in Opt-SHT, suggesting that recrystallization did not occur during the standard heat treatment. This observation is consistent with previous studies [12,27], which report that recrystallization requires temperatures above 1200 °C.

Compared with these L-PBF samples, the precision-cast IN738LC material was characterized by coarse equiaxed-like grains with non-texture, and the GROD values were uniformly low (see Fig. 8(a and b)). The area-weighted average grain diameter was added to Table 4 and is significantly coarser than in the L-PBF samples. As shown in Fig. 8(c), carbides (black arrows) and γ/γ' eutectics (green arrows) were found not only at grain boundaries but also within grains. Higher magnification images of γ/γ' eutectics at grain boundaries and within a grain are shown in Fig. 8(e) and (f), respectively. These constituents were likely formed during the final stage of slow solidification at grain boundaries and in interdendritic regions [28]. By contrast, in regions free of such constituents, only nearly cuboidal secondary γ' precipitates were observed (Fig. 8(d)), the γ' area fraction and particle size were measured (Tables 2 and 3). Although the secondary γ' fraction of 0.461 is lower than the total γ' fraction of Opt-SHT (0.521), the overall γ' fraction of the precision-cast material may be comparable when the γ' contained in γ/γ' eutectics is taken into account.

3.2. High-temperature tensile properties

Fig. 9 shows the engineering stress-strain curves obtained from the high-temperature tensile tests. Each curve represents the individual result for one specimen tested at each material condition and

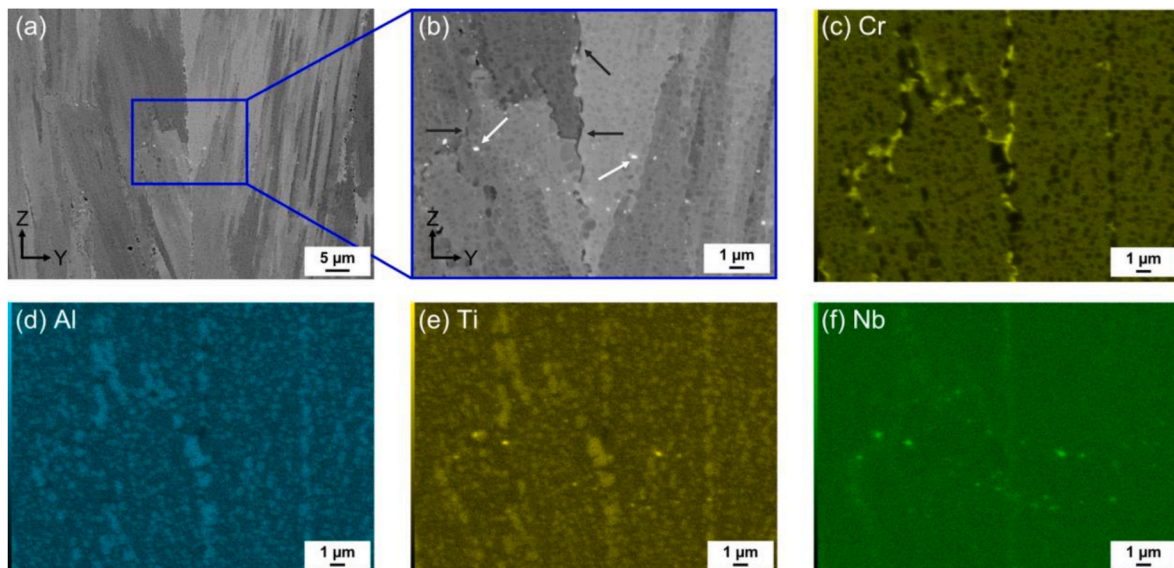


Fig. 6. (a, b) High-magnification backscattered-electron SEM images and EDS element maps of (c) Cr, (d) Al, (e) Ti, and (f) Nb for Opt-SHT.

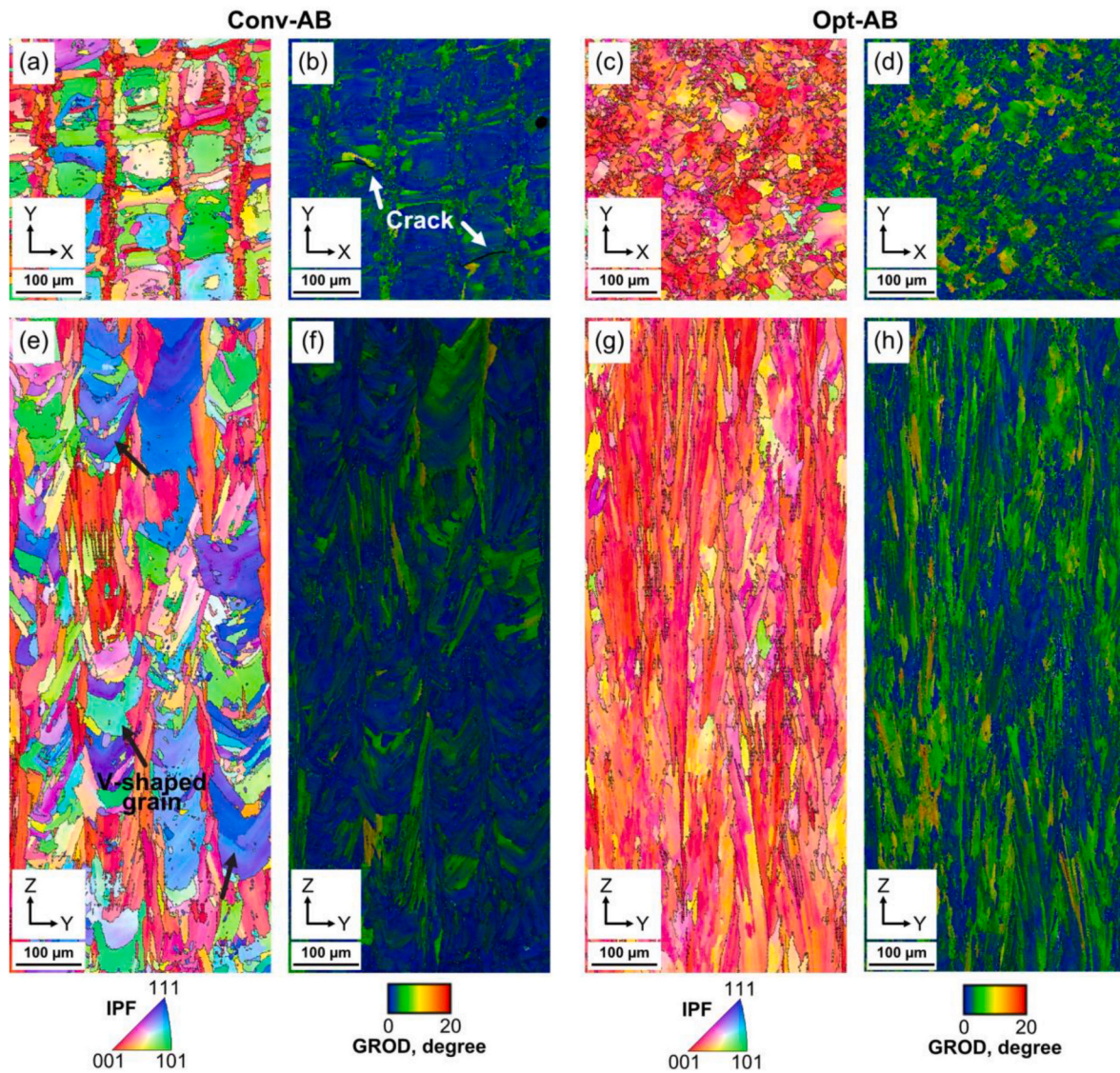


Fig. 7. IPF-Z and GROD maps of (a, b, e, f) Conv-AB and (c, d, g, h) Opt-AB. (a, b, c, d) XY plane; (e, f, g, h) YZ plane. Cracks and V-shaped grains are indicated by white and black arrows in (c) and (e), respectively.

Table 4

Area-weighted grain diameter of L-PBF and precision-cast materials evaluated by EBSD.

	Area-weighted grain diameter, μm	
	XY plane	YZ plane
Conv-AB	36.1	91.5
Opt-AB	30.7	63.9
Opt-SHT	31.3	71.9
Precision-cast	646.3	

temperature. The discontinuity in the curves indicated by the black arrows results from changing the applied strain rate from 0.5%/min to 5%/min after the yield point. As shown in the curves at lower temperatures ranging from 200 °C to 600 °C in Fig. 9(a–c), both Conv-AB and Opt-AB exhibited excellent ductility, with elongation after fracture exceeding 20%. In addition, serrated flow was observed in the plastic deformation region for both samples. This behavior may indicate dynamic strain aging (Portevin–Le Chatelier–type behavior), which has been reported in Ni-based superalloys at intermediate temperatures [29]. The occurrence of serrations suggests interactions between mobile

dislocations and diffusing solute atoms during plastic deformation. In contrast, no serrations were observed in Opt-SHT in the lower temperature range, implying that differences in solute distribution and γ' precipitation state may affect the development of serrated flow. This interpretation is also consistent with the higher strength, exceeding 1250 MPa, but lower ductility in Opt-SHT. At higher temperatures ranging from 800 °C to 1000 °C in Fig. 9(d–f), while both Opt-AB and Opt-SHT had adequate ductility above 10% in elongation, Conv-AB exhibited poor ductility and fractured early with minimal plastic deformation. It is also noted that the strength of all samples decreased as the test temperature increased from 800 °C to 1000 °C.

Fig. 10(a–c) shows the 0.2% yield strength (YS), ultimate tensile strength (UTS), and elongation of each sample condition as a function of test temperature. For comparison, the tensile properties of precision-cast IN738LC are also plotted as blue crosses in Fig. 10. In addition, data from previous works on L-PBF IN738LC [8,9,14,17,19,20,29,30] are also overlaid as white markers outlined in red, light gray, and green, corresponding to as-built, heat-treated (HT), and HIPed conditions, respectively. It should be noted that these reference data are limited to tensile properties measured with the load applied parallel to the build direction. Below 600 °C, the tensile properties of the samples fabricated

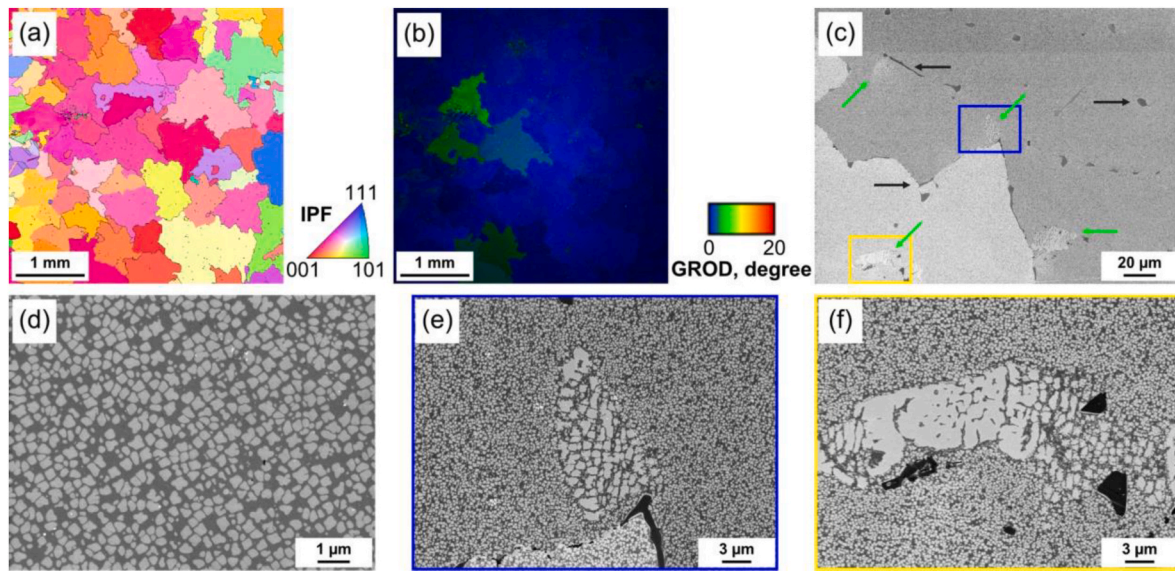


Fig. 8. (a) IPF and (b) GROD maps, and (c–f) SEM secondary electron images of precision-cast IN738LC. In (c), black and green arrows indicate carbides and γ/γ' eutectics, respectively. Higher-magnification images of γ/γ' eutectics are shown in (e) at a grain boundary and (f) within a grain, corresponding to the blue and yellow regions in (c). A region free of such constituents is shown in (d). (For interpretation of the references to color in this figure legend, the reader is referred to the Web version of this article.)

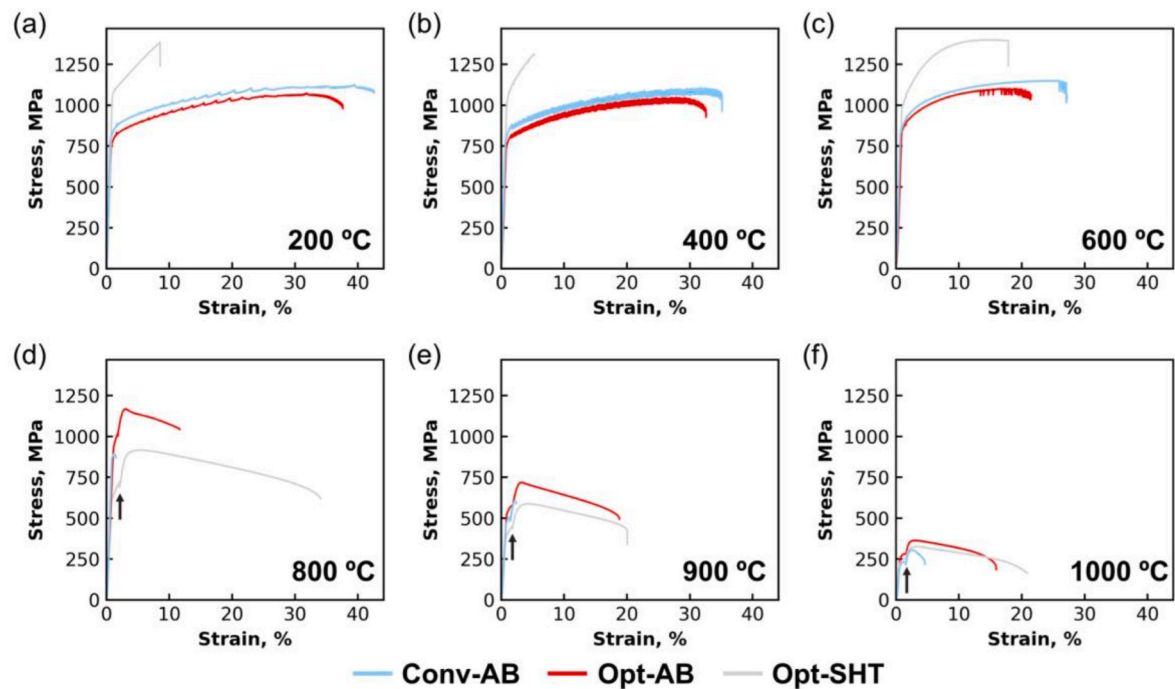


Fig. 9. Engineering stress-strain curves at different test temperatures: (a) 200 °C, (b) 400 °C, (c) 600 °C, (d) 800 °C, (e) 900 °C, and (f) 1000 °C. The discontinuities in the curves, indicated by black arrows, are due to a change in the applied strain rate after the yield point.

by L-PBF were comparable to those of the cast material; in particular, Conv-AB and Opt-AB exhibited significantly better ductility, while Opt-SHT showed higher strength than the precision-cast material. Above 800 °C, Opt-AB and Opt-SHT maintained tensile properties similar to those of the cast material and the L-PBF specimens subjected to HIP [8, 14], whereas Conv-AB showed significantly lower elongation. Since one specimen was tested for each material condition and temperature, these comparisons are intended to indicate overall temperature-dependent trends rather than statistically rigorous differences.

The UTS values of the lower temperature range below 600 °C were

plotted against the elongation (Fig. 10 (d)). A trade-off is apparent, with heat-treated and HIPed specimens showing higher strength at the expense of the superior ductility of the as-built condition. Overall, the L-PBF specimens were found to have a better balance between strength and ductility than the precision-cast materials.

3.3. Microstructure and microhardness after tensile test

After the high-temperature tensile tests, the fracture surfaces and cross-sections of the specimens were further characterized using the

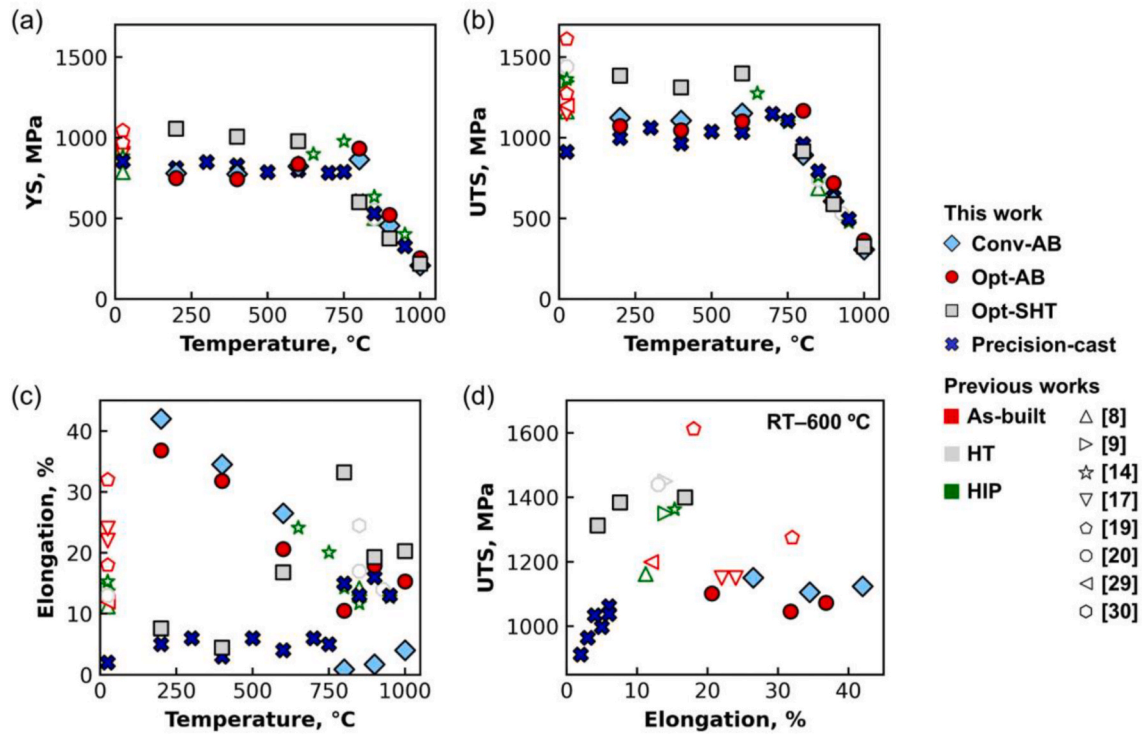


Fig. 10. Tensile properties as a function of test temperatures: (a) 0.2% yield strength (YS), (b) ultimate tensile strength (UTS), and (c) elongation after fracture. (d) UTS against elongation over the range room temperature to 600 °C. Blue crosses represent the tensile properties of IN738LC produced by precision casting [23]. White markers outlined in red, light gray, and green represent the data from previous works on L-PBF IN738LC [8,9,14,17,19,20,30,31], corresponding to as-built, heat treated (HT), and HIPed conditions, respectively. (For interpretation of the references to color in this figure legend, the reader is referred to the Web version of this article.)

microscopes. As representative cases for the lower (200 °C–600 °C) and higher (800 °C–1000 °C) temperature ranges, micrographs of the specimens tested at 400 °C and 800 °C are shown in Figs. 11–13, and Figs. 14–18, respectively. After testing at 400 °C, fine dimples were generally observed on the fracture surfaces of all samples (see Fig. 11). On the surface of Opt-SHT, cracks ranging in size from several micrometers to several hundred micrometers were extensively observed (see

green arrows in Fig. 11 (c)). These fracture surface characteristics were similarly observed in specimens tested at 200 °C and 600 °C.

As shown in the cross-sectional optical micrograph in Fig. 12 (a), Conv-AB contained relatively large defects on the section, compared to Opt-AB and Opt-SHT in Fig. 12(b and c). In the IPF-Z map in Fig. 1 Fig. 2 (d), even though their grains are predominantly oriented between <001> and <111>, locally <101>-oriented regions appear within the grains. In

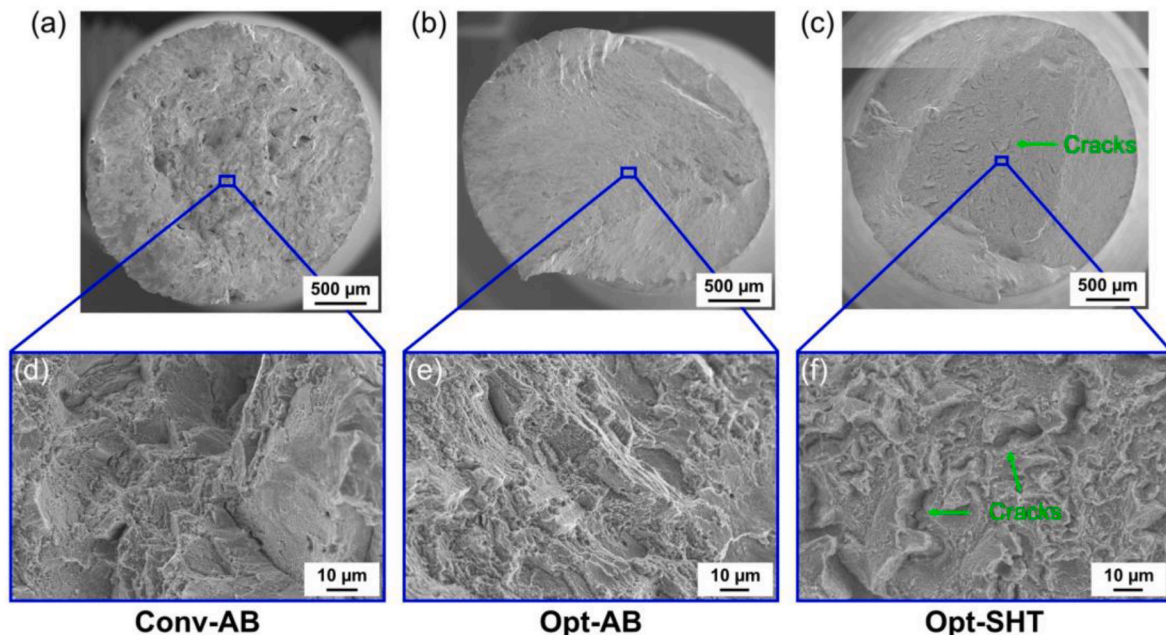


Fig. 11. Fracture surfaces after tensile testing at 400 °C; (a, d) Conv-AB, (b, e) Opt-AB, and (c, f) Opt-SHT.

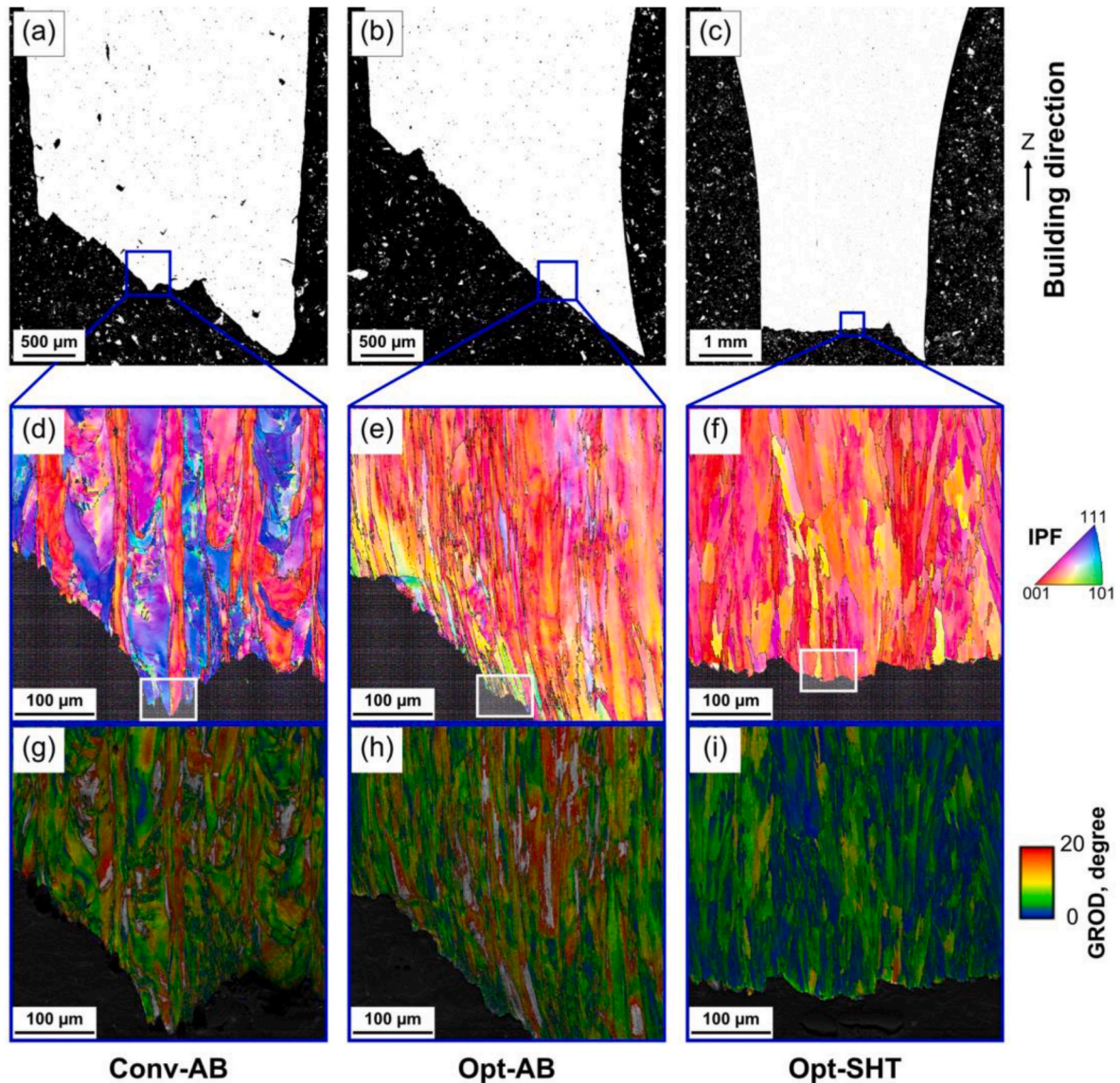


Fig. 12. (a–c) Optical micrographs, (d–f) IPF-Z maps, and (g–i) GROD maps of the cross-sections of the specimens after testing at 400 °C; (a, d, g) Conv-AB, (b, e, h) Opt-AB, and (c, f, i) Opt-SHT. The regions outlined by white rectangles in (d–f) are shown in Fig. 13(a–c), respectively, at higher magnification by SEM.

contrast, the Opt-AB specimen showed yellow regions within approximately 100 μm ahead of the fracture tip in the IPF-Z map (Fig. 12 (e)), indicating an orientation between $\langle 001 \rangle$ and $\langle 101 \rangle$. As shown in Fig. 12 (g and h), the GROD values of these as-built specimens were significantly higher overall than those of the pre-test (Fig. 7(f–h)). Fig. 13 (a–d) and (b, e) show BSE images of the fracture tip within the white rectangles in Fig. 12(d) and (e), respectively. As indicated by the red arrows in Fig. 13 (d and e), voids were found along the boundaries of the cellular substructure in both Conv-AB and Opt-AB.

For the Opt-SHT specimen, the fracture tip region in the IPF map in Fig. 12 (f) indicates that fracture occurred within grains, rather than along grain boundaries. The GROD map in Fig. 12 (i) was essentially unchanged from the pre-test state. These results indicate greater plastic deformation in Conv-AB and Opt-AB than in Opt-SHT during tensile testing at 400 °C, consistent with fracture elongations exceeding 30% for Conv-AB and Opt-AB and less than 5% for Opt-SHT (Fig. 10 (c)). As shown in Fig. 13(c–f), voids were observed along grain boundaries parallel to the Z direction. As indicated by the red arrows in Fig. 13 (f), such voids formed at the interface between the γ matrix and the white spherical particles (white arrow) or between the γ matrix and the dark-gray Cr-rich compound (black arrow). At such grain boundaries near the

specimen tip, cracks originating from voids were observed as indicated by green arrows in Fig. 13 (c), which likely correspond to the cracks seen on the fracture surface in Fig. 11(c–f).

After testing at 800 °C, checkerboard patterns were observed on the fracture surface of Conv-AB, as shown in Fig. 14 (a). The spacing of this pattern corresponded to that observed in the IPF map in Fig. 4 (a). While cellular substructures were identified in the interior regions of the checkerboard pattern, as indicated by the green arrows, quasi-cleavage facets were observed in the surrounding areas, as indicated by the yellow arrow. Such cleavage facets were also observed on the fracture surface of Opt-AB, whereas the fracture surface of Opt-SHT showed dimples (see Fig. 14(b and c)).

As shown in Fig. 15 (a), the Conv-AB specimen after testing at 800 °C shows numerous V-shaped cracks distributed throughout the cross section around the fracture surface. The corresponding IPF-Z and GROD maps in Fig. 15 (d, g) indicate that these cracks propagated along HAGBs of the V-shaped grains, consistent with the features in Fig. 7 (e). In Fig. 15 (d), although fracture appears to have initiated and propagated at the V-shaped grain boundaries, transgranular rupture was also observed in the columnar grains with $\langle 001 \rangle$ aligned parallel to the Z-direction (see yellow arrows in the figure). This is consistent with the

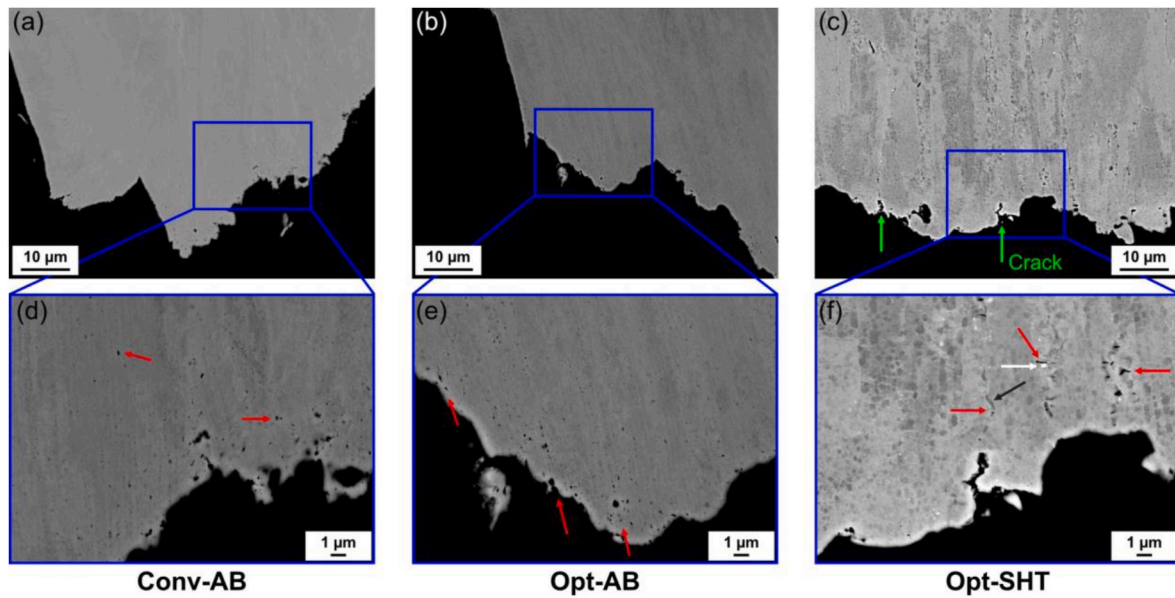


Fig. 13. Backscattered electron images of specimen cross sections after testing at 400 °C, taken from the white rectangles in Fig. 12(d–f): (a, d) Conv-AB, (b, e) Opt-AB, and (c, f) Opt-SHT. The red, green, black, and white arrows indicate voids, cracks, white carbides, dark-gray compounds, respectively. (For interpretation of the references to color in this figure legend, the reader is referred to the Web version of this article.)

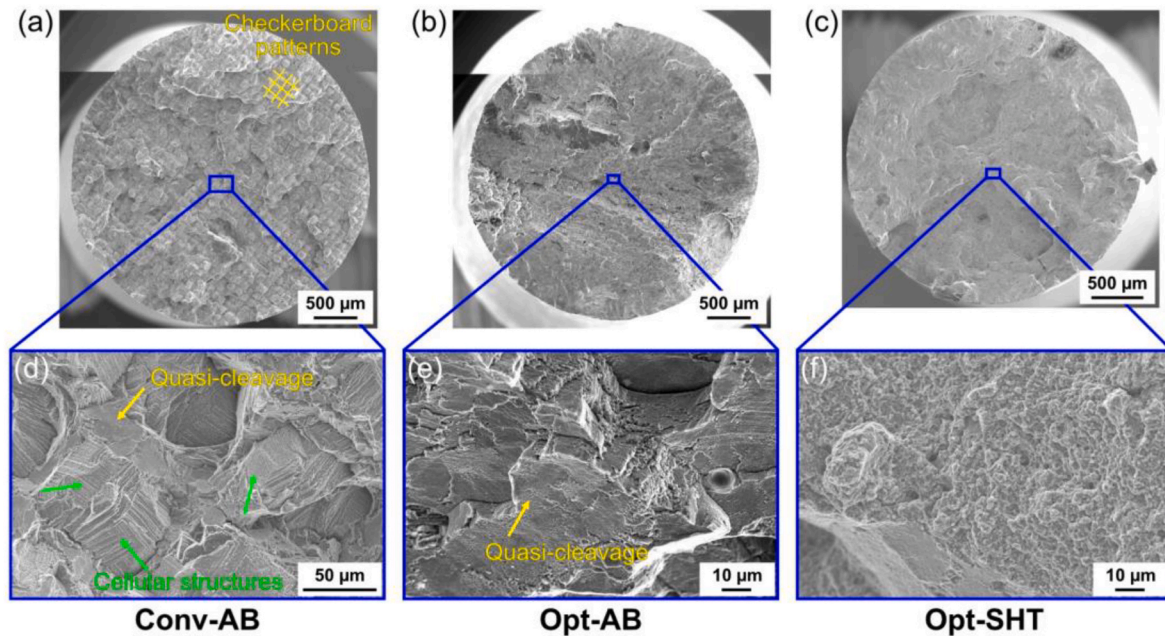


Fig. 14. Fracture surfaces after tensile testing at 800 °C; (a, d) Conv-AB, (b, e) Opt-AB, and (c, f) Opt-SHT.

cellular structures and the surrounding quasi-cleavage observed on the fracture surface in Fig. 15 (a). At higher magnification of the fracture tip and the crack edge within the white and black rectangle in Fig. 15 (d), voids were observed along HAGBs, as shown in Fig. 16 (a, d) and Fig. 17 (a and b). According to EDS analysis, Ti and Nb segregated at cellular boundaries, as in the pre-test condition (cf. Fig. 5 (b, c) vs. Fig. 17(e and f)). Cr was slightly segregated along the HAGBs (Fig. 17 (c)). However, neither Mn nor O was detected at these boundaries, in contrast to the Cr-rich compounds observed in the heat-treated samples.

As for the Opt-AB specimen after testing at 800 °C, several relatively large cracks were distributed throughout the cross-section (Fig. 15 (b)). As indicated by the white arrow in Fig. 15 (h), these cracks occurred primarily at boundaries of locally misoriented grains. However, their

propagation was arrested by grains with $\langle 001 \rangle$ parallel to the Z-direction, and the tensile fracture subsequently propagated transgranularly (Fig. 15 (e)). At the fracture tip, slip lines were observed (yellow arrows in Fig. 16 (e)), and the voids (red arrows) appeared to form along those lines.

On the cross section of Opt-SHT after tensile testing at 800 °C, fine defects formed along grain boundaries elongated parallel to the Z-direction (see Fig. 15 (c)). The IPF-Z map in Fig. 15 (f) indicates that this specimen also fractured transgranularly. These trends are consistent with the higher-magnification BSE-SEM observations in Fig. 16 (c, f) and Fig. 18(a and b). As in the pre-test condition (Fig. 6), the precipitates indicated by the black arrows in Fig. 18 (b) are Cr-rich compounds, as shown by the Cr map in Fig. 18(c), and Mn, Mo, and O were also

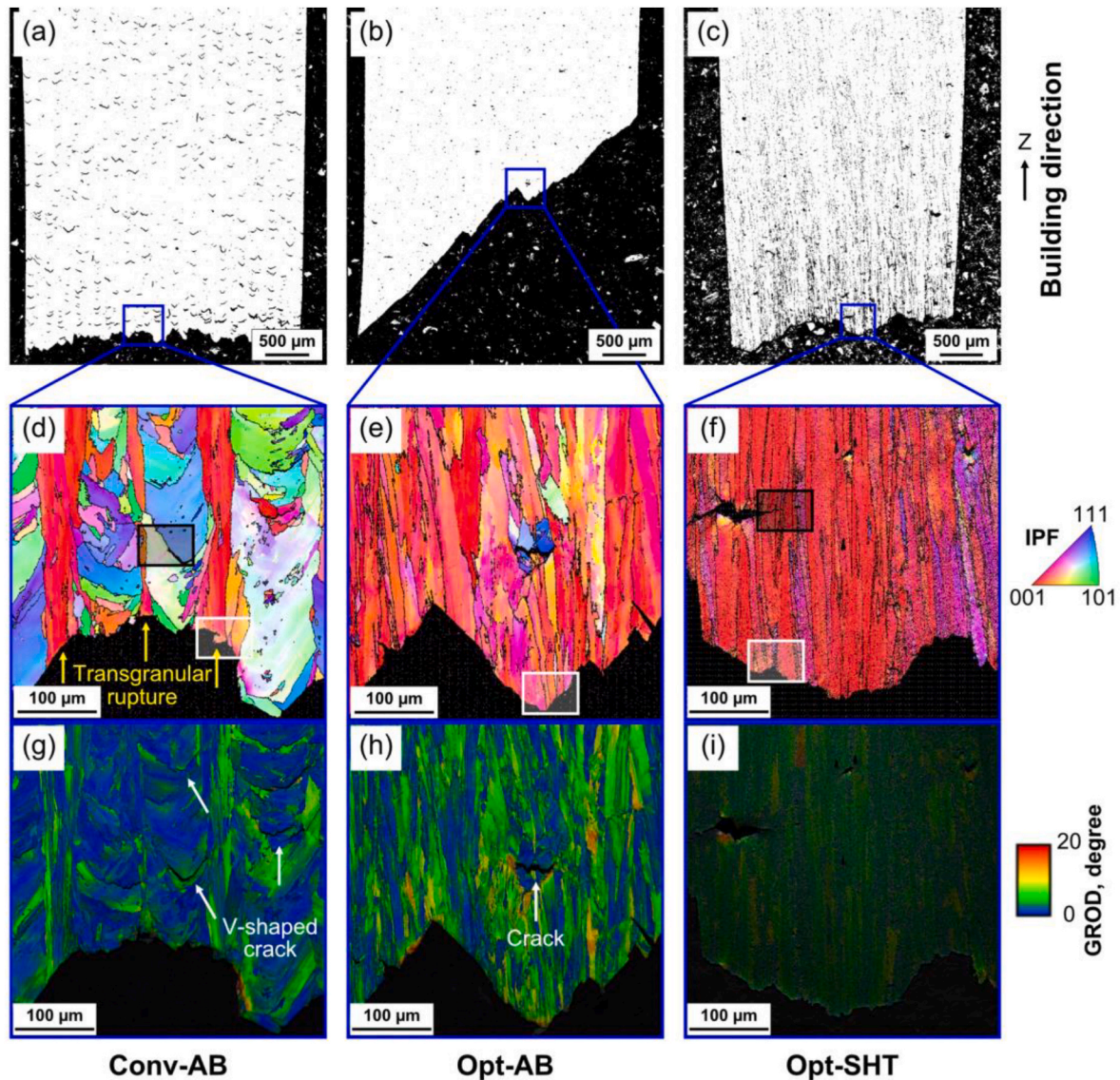


Fig. 15. (a–c) Optical micrographs, (d–f) IPF-Z maps, and (g–i) GROD maps of the cross-sections of the specimens after testing at 800 °C; (a, d, g) Conv-AB, (b, e, h) Opt-AB, and (c, f, i) Opt-SHT. The regions outlined by white rectangles in (d–f) are shown in Fig. 16(a–c), respectively, at higher magnification by SEM. The edges of the cracks enclosed by the black rectangles in (d) and (f) are observed by SEM and EDS in Figs. 17 and 18, respectively.

detected at the same locations. In contrast, the white precipitates were considered Ti/Nb-rich carbides (see Fig. 18(e and f)). The Al and Ti maps in Fig. 18 (d, e) also indicate that relatively large γ' precipitates are aligned in the vertical direction. As shown in Fig. 18 (b), voids formed at the interfaces between the γ matrix and these precipitates at HAGBs, and the crack propagated rightward through the grains by linking these voids.

Fig. 19 shows high-magnification SEM images of the shoulder region of tensile specimens after tensile testing at 800 °C, 900 °C, and 1000 °C. As shown in Fig. 19(a and b), numerous fine light-gray dots were distributed throughout the cross-section of Conv-AB and Opt-AB after testing at 800 °C. Fine spherical tertiary γ' precipitates formed clearly at 900 °C (Fig. 19(d and e)), and larger ones appeared at 1000 °C (Fig. 19(g and h)). For Opt-SHT, the bimodal γ' precipitates formed by the heat treatment remained almost unchanged after tensile testing at 800 °C (cf. Fig. 4 (c) vs. Fig. 19 (c)). However, as the test temperature elevated to 1000 °C, cuboidal secondary precipitates coarsened, while the finer spherical tertiary ones decreased in number (Fig. 19 (i)).

The γ' area fraction and particle size were quantified by image analysis, and plotted as a function of test temperature in Fig. 20(a and b).

As shown by the yellow solid line in Fig. 20 (a), the γ' fraction predicted by Thermo-Calc decreased with the temperature to 0.45 at 800 °C and 0.31 at 1000 °C. In contrast, the measured γ' fraction decreased only slightly, likely because the high-temperature exposure during tensile testing was too short to reach equilibrium. In Opt-SHT, a similar trend seen in the open and light-gray square symbols in Fig. 20 (a) indicates that the reduction of the tertiary γ' fraction largely accounts for the decrease in the total fraction with increasing test temperature.

As for particle size in Fig. 20 (b), Conv-AB and Opt-AB showed significant coarsening with increasing test temperature, reaching about 50 nm at 900 °C and exceeding 100 nm at 1000 °C. It should be noted that the values at 800 °C (dotted circle) are approximate, because very fine particles with unclear boundaries, as shown in Fig. 19(a and b), prevented reliable classification and segmentation in the statistical image analysis. In Opt-SHT, the secondary γ' coarsened to about 600 nm at 1000 °C, while the tertiary γ' size was nearly constant over 800–1000 °C (see the light-gray and open square symbols in Fig. 20 (b)).

Fig. 20 (c) shows the Vickers hardness of the shoulder section of the tensile specimens after testing, plotted as a function of tensile-test temperature. Owing to γ' precipitation induced by the standard heat

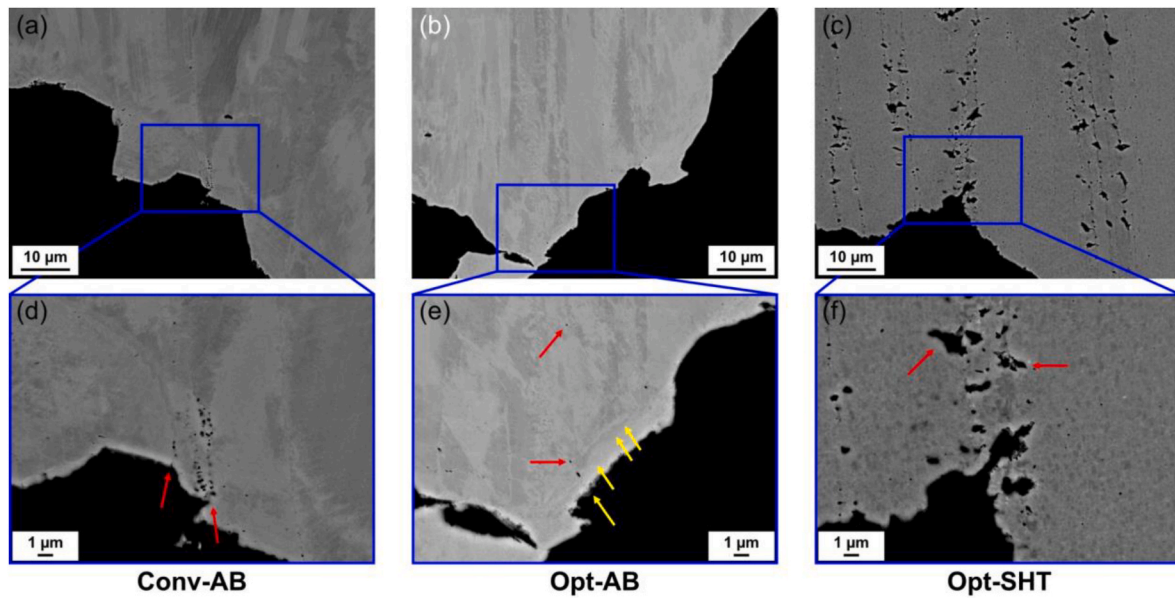


Fig. 16. Backscattered electron images of specimen cross sections after testing at 800 °C, taken from the white rectangles in Fig. 15(d–f): (a, d) Conv-AB, (b, e) Opt-AB, and (c, f) Opt-SHT. The red and yellow arrows indicate voids and slip lines, respectively. (For interpretation of the references to color in this figure legend, the reader is referred to the Web version of this article.)

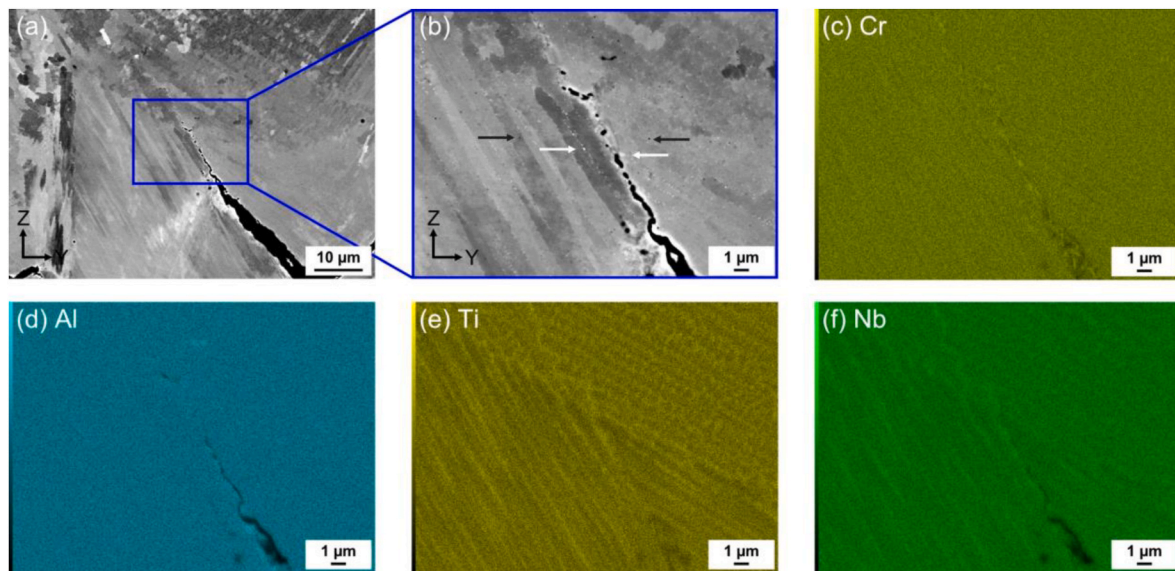


Fig. 17. (a, b) High-magnification backscattered-electron SEM images and EDS elemental maps of (c) Cr, (d) Al, (e) Ti, and (f) Nb at the crack edge in Conv-AB after testing at 800 °C, taken from the black rectangle in Fig. 15 (d).

treatment, Opt-SHT showed a higher hardness of 432 HV in the pre-test state, compared with 395 HV for Conv-AB and 359 HV for Opt-AB, and it increased slightly to 450 HV after testing at 800 °C. By contrast, both Conv-AB and Opt-AB hardened markedly after testing above 800 °C, reaching maxima of 529 HV (Conv-AB) and 551 HV (Opt-AB) at 900 °C.

Similarly, the precision-cast specimens were observed by SEM, EDS, and EBSD on fracture surfaces and cross-sections after tensile tests at 400 °C and 800 °C (Figs. 21 and 22). At higher magnification of the fracture surfaces, brittle regions were found between dimple areas as indicated by white arrows in Figs. 21 (c) and 22 (c). EDS identified these regions as Ti-, Nb-, Si-, and Mo-rich carbides (see Fig. 21 (d, e) and Fig. 22(d and e)). In contrast to the L-PBF specimens, Cr was uniform within the γ matrix, and no Cr segregation was observed in the precision-cast specimens. The cross sections at the crack tip indicate intergranular

fracture not only at 800 °C (Fig. 22 (g)) but also at 400 °C (Fig. 21 (g)). In addition, cracks were observed at the carbides along HAGBs (Fig. 21 (h, i) and Fig. 22(h and i)). Although γ/γ' eutectic regions also appeared along the crack (Fig. 22 (i)), their direct role in crack initiation is not supported, as cracks were not observed around some of these regions as indicated by green arrows in Figs. 21 (h) and 22 (j). Also, as in Opt-SHT, no significant change in the intragranular γ' fraction or size was observed after testing at 800 °C (Fig. 22 (k)).

4. Discussion

Based on the above microstructural observations, Fig. 23 summarizes the dominant fracture modes for L-PBF specimens under different laser scanning conditions with and without post-heat treatment, as well as for

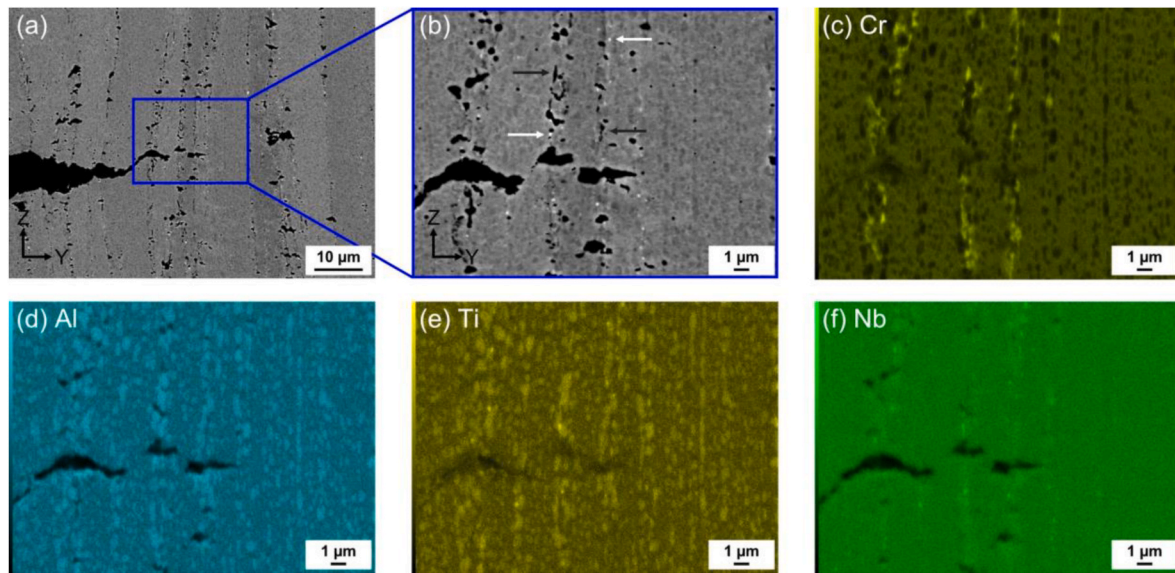


Fig. 18. (a, b) High-magnification backscattered-electron SEM images and EDS elemental maps of (c) Cr, (d) Al, (e) Ti, and (f) Nb at the crack edge in Opt-SHT after testing at 800 °C, taken from the black rectangle in Fig. 15 (f).

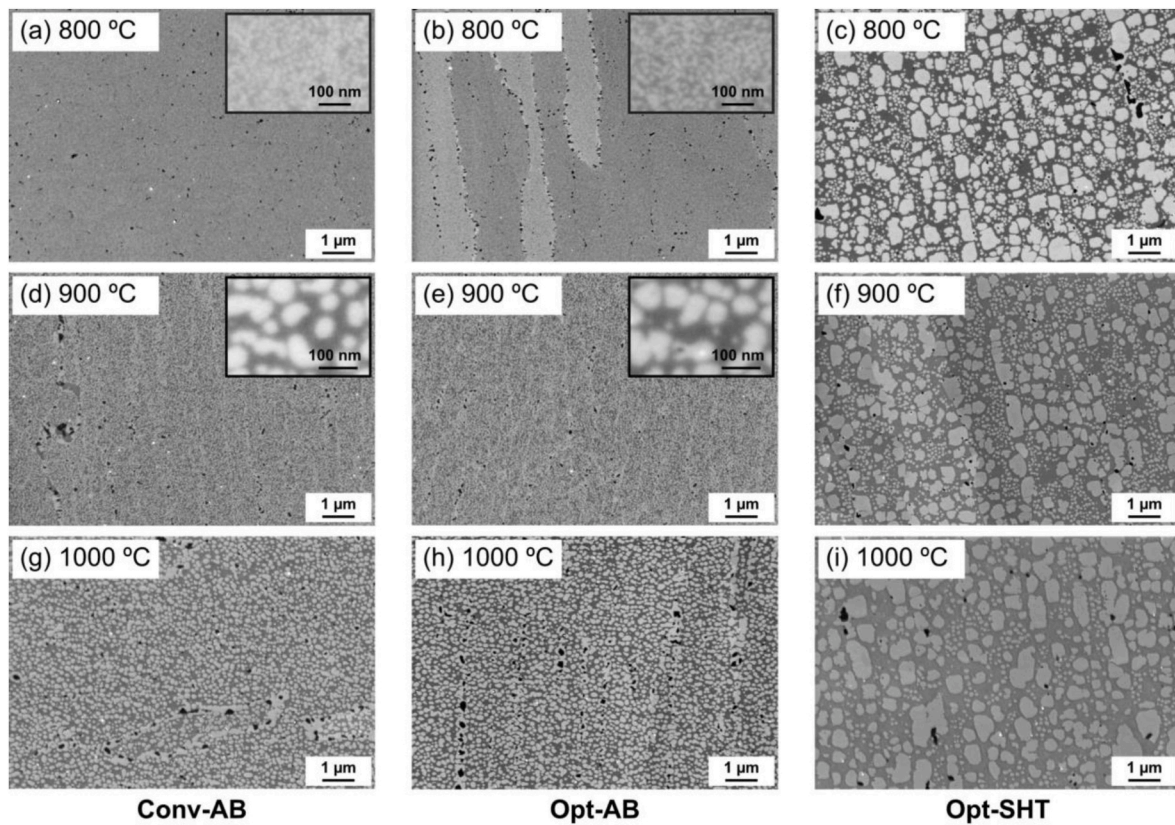


Fig. 19. High-magnification SEM images of the specimens after tensile testing at (a–c) 800 °C, (d–f) 900 °C, and (g–i) 1000 °C; (a, d, g) Conv-AB, (b, e, h) Opt-AB, and (c, f, i) Opt-SHT.

the precision-cast reference. Since the trends differ markedly between the lower and higher tensile test–temperature ranges (200–600 °C and 800–1000 °C), the following sections discuss the fracture mechanisms for each range.

4.1. Fracture mechanisms in the lower-temperature range

At lower test temperatures (top row in Fig. 23), all L-PBF specimens fractured primarily transgranularly. However, the as-built specimens (Conv-AB and Opt-AB) showed substantial plastic deformation with elongations exceeding 20%, whereas the heat-treated one (Opt-SHT) showed little deformation and thus lower elongation (Fig. 10 (c)). This

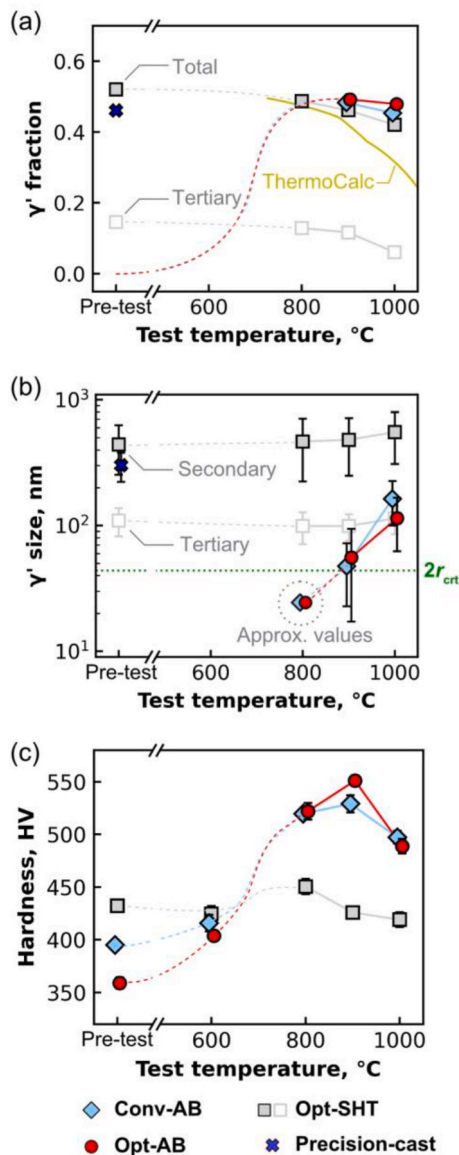


Fig. 20. (a) Area fraction and (b) particle size of γ' precipitates, and (c) Vickers hardness, plotted as a function of tensile-test temperature. The γ' fraction from thermodynamic calculations using Thermo-Calc is shown in (a). In (b), the green dotted line represents $2r_{crit}$, the theoretical size that maximizes the increment of CRSS by spherical precipitates, as discussed in Sec. 4.2. (For interpretation of the references to color in this figure legend, the reader is referred to the Web version of this article.)

difference is mainly attributed to the absence or presence of γ' precipitates.

In the as-built samples, no γ' precipitates were observed within the resolution limit of SEM (Fig. 4(a and b)). The nanoscale chemical state of as-built IN738LC samples has also been investigated in previous studies [32], and atom probe tomography (APT) and transmission electron microscopy (TEM) revealed the presence of γ' precursors or nanoscale compositional clustering, i.e., Cr-depleted and Al- and Ti-enriched regions without significant ordering, with characteristic sizes of less than 10 nm. Such nanoscale features may contribute to the mechanical response of the as-built material. However, their contribution cannot be quantitatively assessed based on the characterization performed in the present study. Instead, cellular substructures with Ti- and Nb-enriched boundaries were observed, with a measured width (PDAS) of 0.58 μm and 0.54 μm for Conv-AB and Opt-AB, respectively. According to the

previous study on feedback control of interlayer temperature during L-PBF [26], the PDAS of as-built IN738LC decreased from 1.2–1.5 μm to 0.8–1.0 μm as the interlayer temperature was reduced from 700 °C to 400 °C. This suggests that the interlayer temperature in the present builds was below 400 °C, which is too low to permit γ' precipitation at that temperature through the process. Furthermore, as estimated in Sec. 3.1, the rapid cooling rate at solidification (6.4×10^5 K/s for Conv-AB and 7.9×10^5 K/s for Opt-AB) also implies the suppression of γ' precipitation during cooling from 1100 °C to 700 °C. The absence of γ' precipitates in the as-built specimens accounts for its lower YS and UTS compared with Opt-SHT in the lower temperature range, and also explains the lower Vickers hardness of the pre-test state (Fig. 20 (c)). As a trade-off, the as-built specimens presented superior ductility (Fig. 10 (d)).

Compared with Opt-AB, Opt-SHT had 140–250 MPa higher YS and 260–310 MPa higher UTS at the lower temperature range (Fig. 10(a and b)), owing to γ' precipitation. In addition, while voids were observed at cellular boundaries in the as-built specimens, they nucleated at interfaces between coarsened γ' precipitates and the γ matrix, or between Cr-rich compounds and the γ matrix at HAGBs in Opt-SHT (Fig. 13(c–f)). Although these precipitates did not appear to contribute directly to the tensile properties in this temperature range, tensile loading perpendicular to the build direction may promote fracture along HAGBs with these precipitates, potentially degrading the tensile performance.

By contrast, the precision-cast material had similar YS and UTS to the as-built samples despite strengthening by γ' precipitates. This is attributed to γ/γ' eutectics at HAGBs and in interdendritic regions formed during the last stage of solidification (Fig. 8(b and c)), which deplete the intragranular γ' fraction. Consistent with this, the intragranular γ' fraction for precision-cast material was lower than that of Opt-SHT (Table 2). Previous studies [28,33] likewise reported that a high fraction of γ/γ' eutectic degrades the mechanical properties by producing γ' -deficient regions. In addition, differences in the γ' size distributions between Opt-SHT and the precision-cast material also likely affected the tensile properties. The γ' precipitates in Opt-SHT exhibited a bimodal distribution with average sizes of about 441 nm and 109 nm, whereas the cast material showed a unimodal distribution with an average size of about 301 nm (Table 3). In a study on cast IN738LC by Balikci et al. [34], in which different aging treatments were used to control the γ' size distribution, the alloy with a bimodal γ' distribution (≈ 50 nm and ≈ 450 nm) showed a tensile strength more than 100 MPa higher than that of an alloy with a unimodal medium-sized γ' distribution (~ 450 nm), supporting the high strength observed for Opt-SHT in the present study. Furthermore, as shown in Table 4, the precision-cast material has larger grains than the L-PBF materials and lacks the fine cellular substructure present in the as-built ones. Consequently, no Hall–Petch strengthening would be expected for the precision-cast material. These multiple factors explain why its strength was only comparable to that of the as-built materials without precipitation strengthening. In contrast, the ductility of the precision-cast material was inferior, likely because intergranular fracture initiated at carbides at HAGBs (Fig. 21 (h)). This places the precision-cast material outside the expected strength-ductility trade-off (Fig. 10 (d)).

In summary, within the lower temperature range, L-PBF specimens generally have tensile properties comparable to or better than those of the precision-cast material. Rapid solidification in L-PBF suppresses the formation of γ/γ' eutectics and carbides during the last stage of solidification. Subsequent heat treatment precipitates γ' as intended by the alloy design, thereby theoretically contributing to strength. Given the strength-ductility trade-off, Opt-SHT is recommended when strength is prioritized, whereas Opt-AB is preferable when ductility is required. Furthermore, even when fine solidification cracks were present in Conv-AB, the tensile properties were nearly the same as the defect-minimized one (Opt-AB).

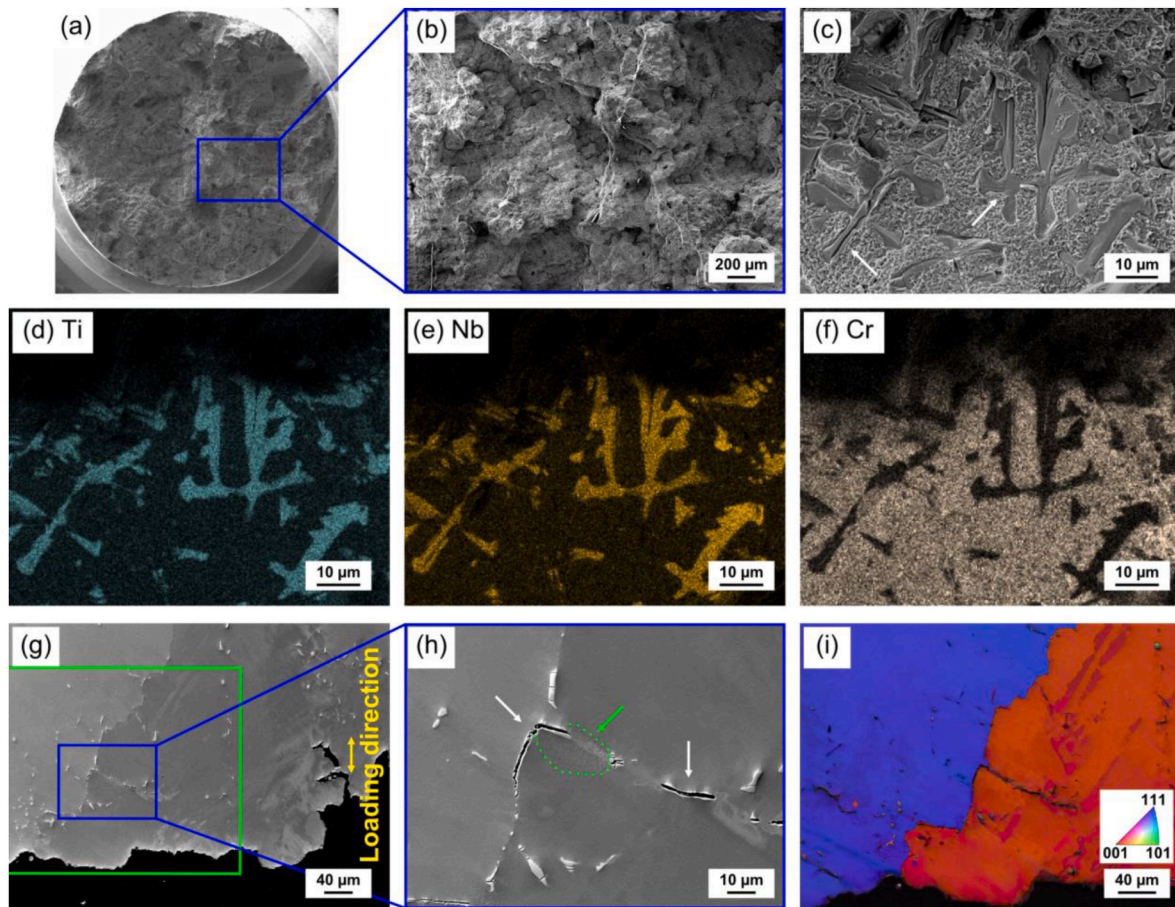


Fig. 21. Observation of (a–f) the fracture surface and (g–h) cross-section of the precision-cast material after testing at 400 °C. (a–c, g, h) SEM images; (d–f) EDS maps of Ti, Nb, and Cr corresponding to (c); (i) IPF map referenced to the loading direction, corresponding to the green-outlined region in (g). White arrows indicate carbides (c) on the fracture surface and (h) along cracks. Green arrow indicates a γ/γ' eutectic region at HAGBs in (h). (For interpretation of the references to color in this figure legend, the reader is referred to the Web version of this article.)

4.2. Fracture mechanisms in the higher-temperature range

At higher temperatures of 800 °C and above (bottom row in Fig. 23), the fracture mode differed markedly between Conv-AB and Opt-AB (and Opt-SHT) because their as-built grain morphologies and HAGBs were distinct, resulting in either intergranular or transgranular fracture. As described in Sec. 3.1, laser scanning conditions, classified as conventional (“Conv”) or optimal (“Opt”) on the basis of microcrack suppression, produced distinct crystallographic textures. As shown in Fig. 7, Conv-AB showed V-shaped grains on the YZ plane and a checkerboard pattern on the XY plane, whereas Opt-AB was dominated by columnar grains elongated along the Z direction with serrated grain boundaries, and its $\langle 001 \rangle$ direction was strongly aligned with the build direction. The Opt-AB microstructure is analogous to that of the directionally solidified (DS) cast material [1].

In Conv-AB, cracking initiated at V-shaped HAGBs nearly perpendicular to the loading axis (Fig. 15 (a–d, g)), and the specimen failed prematurely with little plastic strain at the elevated temperatures. At such crack tips, Cr was segregated along the HAGBs, and voids had formed. The remaining columnar grains with $\langle 001 \rangle$ aligned with the Z direction (yellow arrows in Fig. 15 (d)) then bore the load, leading to transgranular rupture with little plastic deformation. Such HAGBs nearly perpendicular to the loading axis were sparsely observed in Opt-AB, and cracks initiated at these boundaries (white arrow in Fig. 15 (h)). However, crack propagation was arrested by adjacent columnar grains with $\langle 001 \rangle$ aligned with the Z direction. Consequently, the Opt-AB specimen fractured transgranularly with dislocation slip (Fig. 16 (e)),

and maintained elongation exceeding 10% even in the high-temperature range. This improvement is similar to the superior high-temperature mechanical properties of DS cast alloys compared with conventional castings, which are attributed to columnar grains having fewer transverse grain boundaries than equiaxed grains [1]. These results should, however, be interpreted as specific to loading parallel to the build direction. The DS-like microstructure in Opt-AB also implies anisotropy, with lower ductility expected under loading perpendicular to the build direction. Indeed, Chen et al. [14] reported lower perpendicular ductility in L-PBF IN738LC due to build-aligned columnar grains. According to the study by Alabort et al. [35], in-situ SEM tensile tests combined with crystal-plasticity finite-element modeling at 750 °C on a nickel-based polycrystalline superalloy showed that grain boundaries nearly perpendicular to the loading axis act as damage nucleation sites, after which cracks propagate along the boundary and strain accumulates in a neighboring grain until transgranular failure occurs. In Opt-SHT, fracture predominantly within grains (Fig. 15 (c–f, i)), as in Opt-AB, and the tensile properties were comparable to those of the precision-cast material. However, in Opt-SHT, microvoid nucleation was pronounced at interfaces between precipitates and the γ matrix along HAGBs (Fig. 16 (c–f)). The crack then appeared to advance by shearing through the grains, linking these voids (Fig. 18).

In addition, as observed in Fig. 19, fine spherical γ' precipitation in these as-built specimens during high-temperature tensile testing should be considered, as it can affect the measured tensile properties. However, since the microstructural observations were performed after the test, it is difficult to clearly separate the contributions from thermal aging,

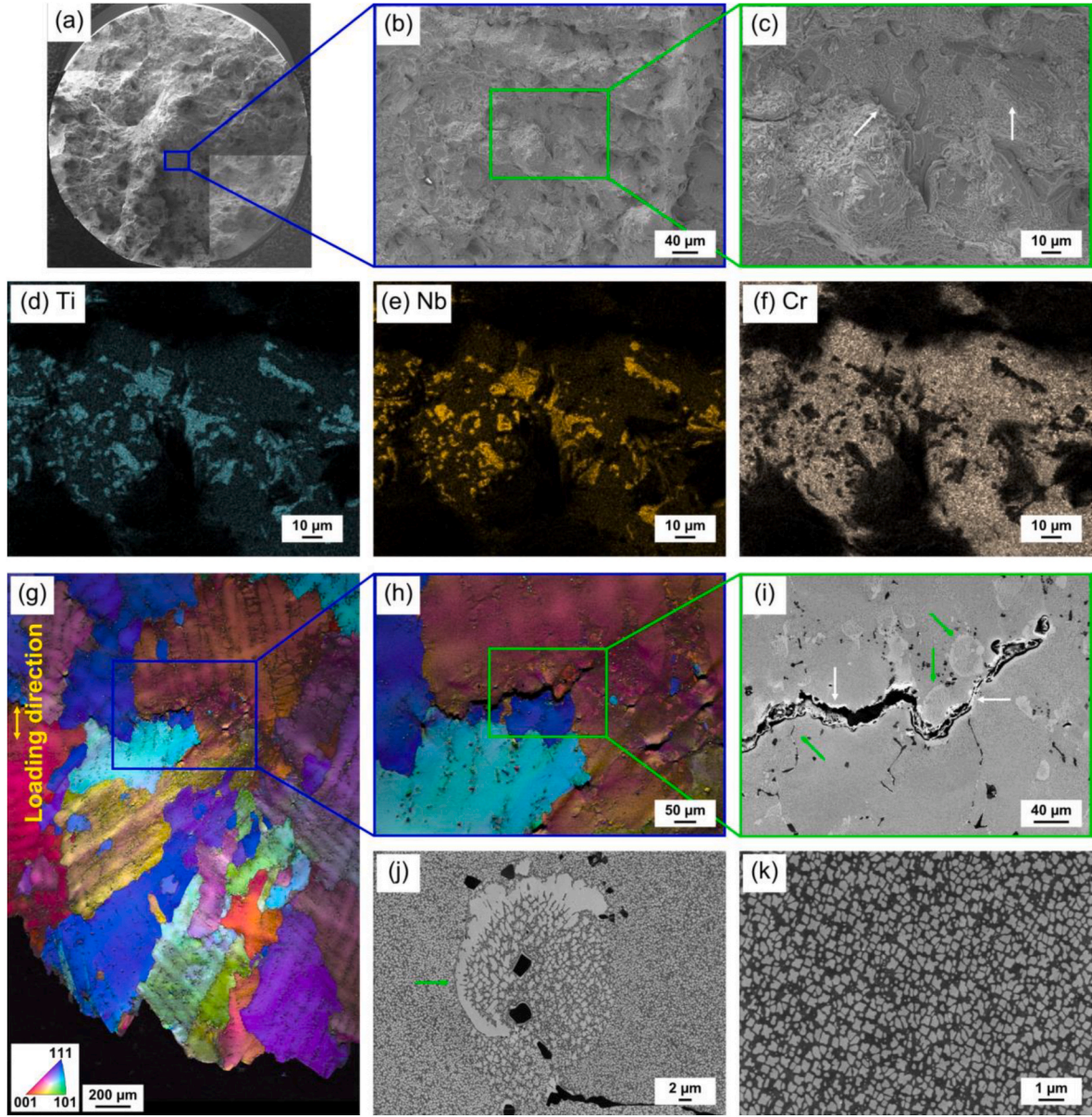


Fig. 22. Observation of (a–f) the fracture surface and (g–k) cross-section of the precision-cast material after testing at 800 °C. (a–c, i–k) SEM images; (d–f) EDS maps of Ti, Nb, and Cr corresponding to (c); (g, h) IPF maps referenced to the loading direction. White arrows indicate carbides (c) on the fracture surface and (i) along cracks. Green arrows indicate γ/γ' eutectic regions in (i, j). (For interpretation of the references to color in this figure legend, the reader is referred to the Web version of this article.)

deformation-assisted processes, and any residual work-hardening effects. Thus, the following discussion should be interpreted as indicating a likely contribution of γ' evolution to the observed mechanical response. As described in Sec. 2, more than 30 min elapsed from reaching the test temperature to fracture in the tensile test. According to Wang et al. [36], even short time heat treatments of electron beam weld IN738LC (e.g., 800 °C for 30 min or 900 °C for 1 min) were sufficient to precipitate very fine γ' particles. In the study, the hardness exceeded 550 HV when the γ' particle size was around 50 nm, and fell below 500 HV as the particles coarsened beyond 100 nm. This relationship arises from a competition between dislocation cutting and Orowan bypass strengthening. The critical resolved shear stress (CRSS) associated with dislocation cutting increases as the γ' size approaches a critical value (≈ 60 nm in Wang et al. [36]), whereas the CRSS for bypassing varies inversely with the γ' size above that critical value. The increment $\Delta\tau_p^s$ in CRSS due to strong pair-coupling of dislocations by spherical γ'

precipitates of radius r can be modeled as follows [37–39];

$$\Delta\tau_p^s = \frac{2\gamma_{APB}}{b} \frac{1}{L} \left\{ \frac{(2\gamma_{APB}r)^{\frac{1}{2}}}{(Gb^2)^{\frac{1}{2}} + (2\gamma_{APB}r)^{\frac{1}{2}}} \right\} \left\{ r^2 - \left(r - \frac{2w}{\gamma_{APB}} \left(\frac{Gb^2}{2\pi} \right) \right)^2 \right\}^{\frac{1}{2}}, \quad (1)$$

with the mean particle spacing

$$L = \left(\frac{2\pi}{3f_v} \right)^{\frac{1}{2}} r. \quad (2)$$

Here γ_{APB} is the anti-phase boundary energy, b is the Burgers vector, G is the shear modulus, w is the fitting parameter, and f_v is the γ' volume fraction. Substituting $\gamma_{APB} = 0.11$ J/m², $b = 0.254$ nm, $G = 77.8$ GPa [40], and $w = 1.0$ gives a critical particle diameter $2r_{crit} = 43.8$ nm that maximizes $\Delta\tau_p^s$ (green dotted line in Fig. 20 (b)). The value is consistent

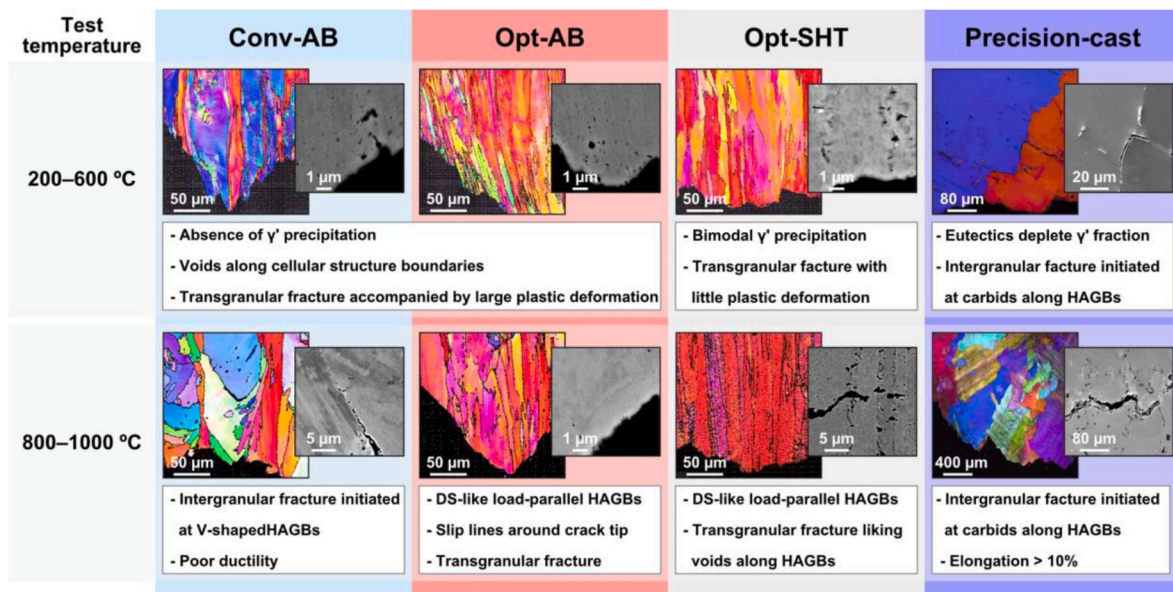


Fig. 23. Fracture modes at lower and higher temperature ranges for Conv-AB, Opt-AB, Opt-SHT, and precision-cast material.

with the pronounced hardening of Conv-AB and Opt-AB after high-temperature tensile testing, especially the peak hardness at 900 °C (Fig. 20 (c)). Furthermore, this may also help explain their higher YS relative to Opt-SHT at 800 °C and 900 °C (Fig. 10 (a)).

In the precision-cast material, the high-temperature fracture behavior appears similar to that at lower temperature. Cracks initiated at carbides and propagated along HAGBs. However, unlike the ductility degradation seen in Conv-AB, intergranular fracture in the precision-cast material still results in elongation above 10%. This difference may be attributed to the segregating elements and the nature of the grain boundary precipitates as well as the internal defects formed during the process.

It should be noted that all tensile tests in the present study were conducted in air so that oxidation may also have contributed to crack initiation and propagation. For example, Fournier et al. [41] reported that, in tensile tests of IN718 at 500–600 °C, oxidation changed the fracture mode from transgranular under vacuum to intergranular in air. Chen et al. [14] also attributed the decrease in strength of L-PBF-fabricated IN738LC at the temperature above 750 °C mainly to chromium oxide along grain boundaries. In contrast, Hilaire et al. [29] reported that, for L-PBF-fabricated IN718 tested at 650 °C, inter-dendritic fracture initiation was not significantly affected by whether the tensile tests were conducted in air or under vacuum. In the present study, Cr-rich compounds were already observed along the grain boundaries of the heat-treated samples (Opt-SHT) before tensile testing, with oxygen also detected at the same locations. This suggests that at least some of the Cr-rich features observed in Opt-SHT after tensile testing originated from pre-existing grain-boundary precipitates. The color change of the fracture surface suggests that oxidation occurred during the tensile tests. However, EDS mapping on the cross section did not provide clear evidence of oxidation at the crack tip in either the as-built or heat-treated specimens, probably because the oxide layer was too thin to be detected by EDS. Further tensile tests under controlled atmospheres together with detailed elemental analysis are required to clarify this issue.

Assuming that turbine engine components of IN738LC are manufactured by L-PBF and put into service, their fatigue properties and creep properties should be assessed alongside their tensile properties. In our previous studies [42–44], the mechanical properties of L-PBF-fabricated Hastelloy X with a microstructure similar to that of Conv-AB were comprehensively evaluated. The gigacycle fatigue strength was comparable to that of the conventionally cast material, and no significant

difference was observed between specimens loaded in the build and transverse directions [42]. In contrast, the creep rupture time of the L-PBF-fabricated Hastelloy X under loading parallel to the build direction was comparable to that of conventionally cast material, whereas it was reduced by approximately one order of magnitude under loading perpendicular to the build direction because crack propagation was facilitated along high-angle grain boundaries preferentially aligned with the build direction [43,44]. Based on these findings, the DS-like microstructures in Opt-AB and Opt-SHT are expected to benefit creep resistance for loading parallel to the build direction. However, the very fine grains in the XY plane may conversely degrade creep performance due to vacancy generation and accumulation, void and crack initiation at grain boundaries, and grain-boundary sliding. Actually, according to Song et al. [45], L-PBF IN738LC with coarse recrystallized grains produced by unconventional post-heat treatments showed a 1.5–4× increase in creep rupture life relative to SHT material, as well as reduced secondary creep rates. Therefore, further microstructural optimization through both L-PBF processing and subsequent heat treatments should be pursued in future studies, together with evaluations of creep behavior, fatigue strength, and fracture toughness.

5. Conclusion

This study evaluated the tensile properties of IN738LC fabricated by L-PBF under loading parallel to the build direction, considering different as-built microstructures related to laser scanning parameters, and the additional fine γ' precipitates introduced by subsequent standard heat treatment. The main findings are as follows:

- Optimized laser scanning parameters formed a columnar microstructure with $\langle 001 \rangle$ aligned to the build direction, similar to a directionally solidified microstructure. This as-built microstructure provided excellent tensile ductility, with elongations exceeding 10% under loading parallel to the build direction.
- Conventional laser scanning parameters produced V-shaped grains whose boundaries became sites for intergranular fracture at tensile temperatures of 800 °C and above. As a result, tensile elongation was significantly reduced at these temperatures.
- With standard heat treatment, bimodal γ' precipitates formed within the directional-solidification-like microstructure, leading to yield and ultimate tensile strengths exceeding 978 MPa and 1312 MPa,

respectively, at the lower temperatures up to 600 °C. However, the elongation decreased to a level comparable to that of precision-cast IN738LC.

- The presence of γ/γ' eutectics and carbides, formed under the slower solidification rates of precision casting, tends to reduce both strength and ductility at the lower temperatures up to 600 °C. Such constituents are suppressed under the rapid solidification of L-PBF, which is another characteristic advantage of the process.
- Both the as-built and heat-treated specimens fabricated with the optimized scanning parameters showed excellent tensile properties across the tested temperature range that were comparable to or better than those of precision-cast material under loading parallel to the build direction.

CRediT authorship contribution statement

Masahiro Kusano: Data curation, Formal analysis, Funding acquisition, Investigation, Validation, Visualization, Writing – original draft, Writing – review & editing. **Tomotaka Hatakeyama:** Formal analysis, Investigation, Writing – review & editing. **Toshio Osada:** Formal analysis, Validation, Writing – review & editing. **Kota Sawada:** Investigation, Writing – review & editing. **Makoto Watanabe:** Funding acquisition, Project administration, Writing – review & editing.

Declaration of competing interests

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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Data availability

Data will be made available on request.

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