High-Temperature Creep Properties of an Additively Manufactured Y₂O₃ Oxide Dispersion-Strengthened Ni–Cr–Al–Ti γ/γ’ Superalloy

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The effects of the modification of a high-γ’ Ni–8.5Cr–5.5Al–1Ti (wt%) model superalloy with 0.5 wt% Y₂O₃, manufactured by laser powder bed fusion, on the microstructure and compressive creep resistance, are investigated. Compared with the base alloy, the oxide dispersion-strengthened (ODS) alloy exhibits 8–10 times slower creep rates at 800 °C, over a wide range of compressive stresses (35–250 MPa). Two creep regimes are observed: diffusional creep, hypothesized to be due to grain boundary sliding, and dislocation creep, with stress exponents n of 2 and 5, respectively. Compared with the horizontal direction, the vertical build direction is characterized by lower creep rates, due to the vertically elongated grain structure. Nonetheless, the ODS alloy’s weakest (horizontal) direction shows better creep resistance than the non-ODS alloy in its strongest (vertical) direction, despite a higher number of as-built defects (slag, cracks) in the ODS alloy. The strengthening potential of ODS modification of additively manufactured nickel-based superalloys is thus demonstrated. The development of successful commercial ODS nickel superalloys for additive manufacturing processing requires further additions of grain boundary strengthening elements as the dispersoids do not significantly increase the grain boundary strength, as their cracking is prevalent in the ODS alloy.

1. Introduction

Oxide dispersion-strengthened (ODS) alloys combine superior mechanical properties, corrosion, and creep resistance at temperatures beyond the coarsening and dissolution limits of classical precipitation strengthening.[1–3] However, alloys that rely on dispersed strengthening alone suffer from low strength at intermediate temperatures, where they are being outperformed by conventionally cast precipitation-hardened Ni-based superalloys.[2–4] This led to the development of ODS Ni-based superalloys combining three types of strengthening (matrix solid solution, γ’ precipitates, and oxide dispersoid), such as MA6000 (Ni–15Cr–4.5Al–2.5Ti–4W–2Ta–2.5Fe–1.1Y₂O₃, wt%), PM3030 (Ni–17Cr–6Al–2Mo–3.5W–2Ta–0.9Y₂O₃), and their variants.[7–11] The contributions of γ’ precipitates and oxide dispersoids to the creep resistance of these ODS alloys are found to be additive, making them superior to their solely precipitation-hardened counterparts.[12] However, when the precipitates are then dissolved at high temperatures, similar creep resistance is observed again between γ’ and non-γ’ strengthened ODS alloys, as only the dispersoids provide resistance to creep.[8] Ultimately, ODS alloys provide superior performance compared with γ’/γ Ni-based superalloys only at medium stresses, at temperatures above γ’ coarsening or dissolution temperature and when high microstructural stability is required.[13] The need for additional strength at lower temperatures stems from applications where ODS alloys are desired for the hottest part of a component but still have to fulfill specifications in sections at lower temperatures.[7] While technical feasibility has been demonstrated, ODS alloys remain a niche solution due to the costly powder metallurgy processing route, heat treatments required for microstructure adjustment, difficulties in achieving net-shaped parts, and competition from continued optimization of conventional alloys.[14,15]

The challenges in processing and part complexity can be overcome by melt-based additive manufacturing (AM), either by laser powder bed fusion (L-PBF) or electron beam melting (EBM) to melt and consolidate the ODS powder, which features very short times in the liquid phase, thus helping to maintain isotropic dispersoid distribution within the solidified alloy.[16–19] As shown...
by several studies, oxide dispersoids can be created during AM (from oxygen contamination present as inclusions or oxide layers), or they can be deliberately added to the alloy powders prior to AM. By comparison, ODS Ni-based alloys receive much less attention. Spierings et al. manufactured an ODS variant of IN625 (Ni–20Cr–9Mo–3.75Nb, wt%) modified with Y2O3 dispersoids. This ODS alloy showed superior oxidation resistance compared with its non-ODS IN625 counterpart. Guo et al. added Y2O3 to IN738LC to successfully scavenge Zr from the grain boundaries to prevent cracking. However, the high-temperature creep properties are so far absent from published literature for additively manufactured ODS Ni-based alloys.

2. Experimental Section

2.1. PBF of Creep Specimens

Prealloyed Ni–8.5Cr–5.5Al–1Ti (wt%) powders were fabricated by gas atomization by Nanoval (Germany) and sieved to 63 μm. ODS-modified powders were produced by rotational ball milling for 24 h of these gas-atomized powders together with 0.5 wt% Y2O3 nanoparticles (10 nm, 99.99%, US Research Nanomaterials), using polypropylene vials with ZrO2 media (5:1 ball-to-powder ratio). The composition of both powders is reported in Table 1. Two blocks were produced, one from the unmodified, “base” powders and the other from the ball-milled, ODS powders, on a Sisma MySint 100 (Sisma S.p.A., Italy) PBF machine equipped with a 200 W fiber laser operating in the continuous-wave mode with a 55 μm spot size and 1070 nm wavelength. A bidirectional scan strategy (90° rotation between layers) was used for each layer, followed by a contour scan, at 150 W power, 1000 mm s\(^{-1}\) scanning speed, with 30 μm layer height, and 75 μm hatch spacing, under Ar shielding gas (O\(_2\) < 0.01%). These parameters were optimized in previous work to maximize density (Base = 7.965 g cm\(^{-3}\) and ODS = 7.925 g cm\(^{-3}\)) and minimize processing-induced defects such as pores, lack of fusion, oxide slag, and microcracks.

After PBF, each consolidated block was separated from the stainless steel baseplate by electrodischarge machining and solutionized at 1260 °C for 2 h in air, followed by air cooling. Cylindrical compressive creep specimens (diameter: 5 mm, height: 11 mm) were then machined from the heat-treated blocks by EDM in the vertical (build direction z parallel to the load direction) and horizontal directions (scan direction x parallel to the loading direction), yielding four cylinders per direction from a single block (Figure 1a,b). The material exhibited a <100> texture aligned along the three printing directions x, y, and z. All cylinders were thus tested along a <100> direction. The remaining material was used for microstructural analysis. Prior to testing, creep specimens were subsequently aged at 850 °C for 24 h in air, followed by air cooling. The described heat-treatment sequence was subsequently optimized and is reported in our earlier investigation summarized in the study by De Luca et al. which also confirmed the high-temperature stability of the dispersoids, with radii of \(\approx 10.4 \pm 3.8\) nm. Measurement of the \(\gamma\) size evolution during isothermal aging does not reveal profound interaction between the presence of

![Figure 1](https://www.aem-journal.com/)

**Figure 1.** Creep specimens and testing apparatus. a) Cylindrical specimens (5 × 11 mm) in vertical and horizontal orientation are electrodischarge machined from a block of alloy directly built onto a steel baseplate. Laser scan directions x and y and build direction z are indicated. Vertical specimens have z along the compression axis and horizontal ones x. b) Consolidated block of ODS Ni–Cr–Al–Ti alloy with the location of the extracted specimens indicated. c) Compressive creep geometry with the specimen between BN-lubricated SiC blocks in a MarM246 creep cage located in a three-zone tube furnace (shown opened). Inset: as-machined and crept specimen.
dispersoids and the $\gamma'/\gamma$ microstructure itself. The dispersoids appeared to sit within the $\gamma$ channels[47] (Figure 2).

2.2. Compressive Creep Testing

Constant-load compressive creep was performed at 800 °C using a dead-load frame equipped with a MarM246 compression cage surrounded by a three-zone furnace (Figure 1c). The creep testing temperature was selected to be below the $\gamma'$ aging temperature (850 °C) in order to avoid further precipitate coarsening and dissolution. The cylindrical specimen was placed between boron nitride (BN) lubricated SiC plates to reduce friction. A dead load was applied to the specimen (in the compressive creep cage) via a fulcrum with a lever arm ratio of 10:1. The applied load was calibrated, prior to the experiment, with a load cell inserted in the cage instead of a specimen. During testing, the loads were converted to engineering stress using the cross-sectional area of the cylinder samples, which was updated after each stress change based on the instantaneous sample height, assuming no barreling. The temperature was controlled, within ±1 °C, by a thermocouple (K-type, Omega) in direct contact with the upper SiC plate. Displacement was measured with a linear-variable displacement transducer (LVDT) with a resolution of 10 μm. After stress-free heating to 800 °C, stepped load creep experiments were performed, where each load step was maintained until the specimen reached a linear steady-state secondary creep regime after which the load increased. The minimum strain rate for each stress level was obtained from the slope of the strain versus time curves. The total strain was kept below 3% to minimize grain boundary cracking. Cooling was performed stress free in the compression cage.

2.3. Microstructure and Defect Characterization

Prior to the creep experiments, microstructural stability at the creep temperature was studied on specimens cut from the same block as the creep specimens. 850 °C heat treatment was used to form a high fraction of $\gamma'$ precipitates, which provide strength and creep resistance to the alloy. The aging duration of the 850 °C step was varied between 24 h (end of peak-aged plateau[47]) and 48 h (overaged) to identify which condition provided better $\gamma'/\gamma$ microstructural stability during subsequent aging at 800 °C for 120 h, corresponding to a typical creep time. After polishing down to 20 nm with colloidal silica, the $\gamma'/\gamma$ microstructure was revealed by etching with Glycerol (HCl:HO:glycerol 15:5:10). The precipitate size was measured by image analysis using ImageJ. Selected creep samples were cut longitudinally along the compression direction, polished to a 20 nm colloidal silica finish, and vibropolished for postcreep microstructure analysis by scanning electron microscopy (SEM). Then, the specimens were Ar-ion beam polished (Leica TIC) for grain orientation mapping by electron backscatter diffraction (EBSD). EBSD was performed at a step size of 2 μm at 30 kV acceleration voltage on a JEOL 4700F FIB-SEM with a Symmetry S2 (Oxford Instruments) detector. Data analysis was performed in AztecCrystal (Oxford Instruments, UK). The HV1 Vickers hardness was measured using a Wilson VH3100 (Buehler, USA) at 1 kgf. Data visualization was performed in Python (Anaconda, Continuum Analytics).

3. Results

3.1. Microstructural Stability

To confirm that the $\gamma'/\gamma$ microstructure remains stable during the duration of the creep test, samples of the base and ODS alloys were peak or overaged at 850 °C for 24 and 48 h and subsequently exposed to a 800 °C treatment for 120 h, the creep test temperature, and typical duration. Figure 3 shows the microstructure of the base and ODS alloys after aging at 850 °C and subsequent exposure at 800 °C for 120 h. The $\gamma'$ precipitate size and alloy hardness were measured after each aging step (Table 2). The specimens aged at 850 °C for 24 h show a higher average hardness and smaller $\gamma'$ size compared with those overaged for 48 h. After exposure to 800 °C for 120 h, both base and ODS alloys, in both their 24 h- and 48 h-aged states, show coarsening of $\gamma'$ (≈10–25 nm) and a slight increase in hardness (≈1–6%), albeit within experimental error. The smaller $\gamma'$ size measured after aging for 24 h remains slightly smaller after exposure at 800 °C for 120 h compared with aging for 48 h for both alloys. Figure 4 shows a comparison of the measured hardness and $\gamma'$ size with the hardening and coarsening curves obtained by De Luca et al. for PBF Ni–8.5Cr–5.5Al–1Ti–(0.5Y2O3) 5 × 5 × 5 mm cubes.[47] Aging for 24 h is at the end of the peak aged plateau, while 48 h exposure leads to overaging. The difference in hardness between the two studies is likely caused by the different setup for the hardness measurement, which involves a higher load for the new datapoints (HV0.2 vs HV1). Hardness indents made at higher loads are known to reduce the overall measured hardness, but in this case they reveal a difference between the ODS and base alloys.[46] This difference is likely caused by the larger deformed volume at higher load, expected to provide a better measure of the strengthening effects of the ODS dispersoids in the grains and at grain boundaries. While the measured $\gamma'$ sizes after 48 h aging is in good agreement with the coarsening curve obtained by De Luca et al.,[47] the values...
after 24 h appear slightly larger. This difference is probably caused by the slower cooling from solutionizing temperature, due to the larger part size. A slower cooling rate leads to a coarser initial population of secondary $\gamma'$ precipitates formed on cooling and thus a shift of the coarsening curve to larger $\gamma'$ sizes at short ageing times. As the 24 h aging condition exhibits a more stable $\gamma'$ population, this condition was employed for the creep study.

### 3.2. Creep Performance and Mechanisms

Figure 5 shows the minimum strain rate as a function of applied stress at 800 °C for the ODS and base alloys, aged at 850 °C for 24 h. The ODS alloy shows, depending on the applied stress, creep rates up to 8–10 times lower compared to the base alloy, illustrating a marked strengthening effect from the dispersoids.

| Ni–Cr–Al–Ti | 24 h | 3020 ± 80 | 136 ± 26 (77 ± 15) | 3050 ± 120 (+1%) | 142 ± 34 (80 ± 19) |
| Ni–Cr–Al–Ti–0.5Y$_2$O$_3$ | 24 h | 3160 ± 130 | 138 ± 31 (78 ± 17) | 3190 ± 140 (+1%) | 161 ± 35 (91 ± 20) |
| | 48 h | 3010 ± 100 | 158 ± 38 (89 ± 21) | 3090 ± 160 (+2%) | 184 ± 38 (104 ± 22) |
Regardless of the building direction, two distinct creep regimes are present for both alloys, as evidenced by the change of slope for strain rates above and below \(\approx 5 \times 10^{-5} \text{ s}^{-1}\). The relationship between minimum strain rate \(\dot{\varepsilon}\) and applied stress \(\sigma\) is given by

\[
\dot{\varepsilon} = A(\sigma)^n \exp\left(-\frac{Q}{k_B T}\right) \tag{1}
\]

where \(A\) is a constant, \(n\) is the stress exponent, \(Q\) is the creep activation energy, \(k_B\) is the Boltzmann constant, and \(T\) is the absolute temperature. The use of the logarithmic scale allows to easily assess the power law creep exponent \(n\) and thus the associated active deformation mechanism. At low stresses, \(n \approx 2\) indicates a diffusion creep mechanism, suspected to be controlled by grain boundary sliding, while at high stresses, \(n \approx 5\) indicates creep controlled by dislocations. Due to the limited deformation of the grain structure during the course of the creep test (maximum strain \(= 3\%\)), it is not possible to unequivocally confirm the grain boundary sliding mechanism.

Creep rate anisotropy is observed between vertical and horizontal test directions (Figure 5a,b). Testing along the vertical build direction results in lower strain rates compared with the horizontal test directions (Figure 5a,b). Testing along the vertical direction, the latter being along one of the laser scan directions. For the ODS alloy, the reduction in creep resistance is consistent over the whole stress range. For the base alloy, the horizontal creep rates approach vertical values at creep rates \(>2 \times 10^{-7} \text{ s}^{-1}\). At all stresses, the weaker, horizontal direction of the ODS alloy still exhibits lower creep rates than the non-ODS alloy tested along its stronger, vertical direction.

### 3.3. Postcreep Microstructure

Figure 6 shows the microstructure of the base and ODS Ni–Cr–Al–Ti alloys after creep deformation at 800 °C to a total accumulated strain of 3%, tested either in the vertical or horizontal direction with respect to the build geometry. The postcreep microstructure of the two alloys is similar and consists of a bimodal grain size distribution: a matrix of large grains (\(\approx 100–200 \mu\text{m}\) wide and up to 1 mm long) containing pockets of much finer grains (Figure 6a,b,d,e). The fine-grained regions are wider and more numerous in the ODS alloy. Various defects are observed: 1) cracks at grain-boundaries oriented vertically in regard to testing direction, as the grains separate during radial expansion of the cylinder sample subjected to uniaxial longitudinal deformation; 2) gas pores present from the PBF process and expected to originate from gas-filled pores in the gas-atomized powder; 3) elongated, dumb-bell shaped, sintered former microcracks; and 4) oxide slag particles. The initial vertical microcracks are formed during PBF but are greatly reduced in size and number by sintering occurring during the solutionizing heat treatment.[47] The cracks are too large to be fully eliminated within the solutionizing duration form elongated pores with bulbous ends. None of these fabrication defects (gas pores, slag particles, or partially sintered cracks) were observed to initiate creep damage in the alloy when located within bulk grains. Damage is solely confined to the grain boundaries, as discussed later.

After creep, the cuboidal \(\gamma'\) precipitates are slightly coarsened along the grain boundaries (Figure 6c,f). Grain boundaries ahead of a crack tip are often observed to be a \(\gamma'\) matrix devoid of \(\gamma'\) precipitates. The strongly localized stress thus appears to destabilize \(\gamma'\) and concomitantly locally reduce the alloy’s strength. Raising of \(\gamma'\) precipitates within grains is not observed. Specimens crept in the horizontal orientation also exhibit cracks parallel to the loading direction with grain boundaries at \(>45^\circ\) to the loading axis remaining uncracked, which agrees with the suspected grain boundary sliding mechanism (Figure 6g,h). The \(\gamma'/\gamma\) structure within the grains is unaffected by the testing orientation. The formation of Ni–Y-rich intermetallic phases at triple points is observed (Figure 6i), likely caused by coarsening of a Y-rich film present on the grain boundaries after PBF.[48]

Figure 7 shows orientation and misorientation maps obtained by EBSD for the base and ODS alloys after creep at 800 °C. The inverse pole figure orientation maps and the 100 pole figures show no strong texture of the specimens along the loading direction. Texture in these alloys is typically observed along the build
direction\[28,46\] which is parallel to the loading axis in Figure 7a,b, as well as along the scan directions, parallel to the loading direction for the horizontal test direction (Figure 7c). Solutionizing leads to recrystallization and thus reduced as-solidified texture.\[47\] The grain shape is anisotropic and typically elongated along the build direction and shortened along the laser scan direction. When specimens are tested vertically, the grains are thus loaded along their longer axis, and for horizontal tests, they are loaded along their short direction. The shape anisotropy is also observed for the fine-grained regions appearing bright in misorientation maps (Figure 7d–f). These fine-grained regions, which result from grain boundary pinning by the dispersoids during solutionizing (Al\textsubscript{2}O\textsubscript{3} in the base alloy\[47\], are in a higher prevalence (and cover a larger area) in the ODS alloy. Strain accumulation in the grains and along grain boundaries is observed in the misorientation maps for all conditions. Strain accumulation within the fine-grained regions is likely but cannot be distinguished, in the misorientation maps, from the high density of

**Figure 6.** Microstructure of ODS Ni–Cr–Al–Ti after creep at 800 °C. Postcreep microstructure for vertical a–c) Ni–Cr–Al–Ti base alloy and d–f) Ni–Cr–Al–Ti–0.5Y\textsubscript{2}O\textsubscript{3} alloy, and for horizontal g–i) Ni–Cr–Al–Ti–0.5Y\textsubscript{2}O\textsubscript{3} alloy after stepwise creep to 3% strain under increasing stress. a,d,g) Overview micrographs, b,e,h) defects present in the structure with retained fine-grained structure shaded in blue, and c,f,i) γ'/γ' structure in the grains and at boundaries. Testing direction is vertical in all micrographs.
pre-existing small angle grain boundaries (Figure 7e inset). Strain accumulation along the boundaries of the large grains indicates deformation throughout the whole specimen, rather than strain localization in the fine-grained regions. While these fine-grained regions appear equiaxed in Figure 7f, this is likely due to the build direction being rotated out of plane due to the rotational symmetry of the cylindrical specimen. It is not possible to know the original build direction, leaving only the scan direction as a defined axis. Based on the data obtained perpendicular to the z-plane in Figure 7d,e, the fine-grained regions are thus elongated along the build direction within the specimens tested, rendering the structure of horizontal specimens rotationally anisotropic.

4. Discussion
4.1. Enhanced Creep Resistance of Additively Manufactured ODS Alloy

The introduction of nanometric oxide dispersoids into an AM \( \gamma/\gamma' \) strengthened Ni-based model superalloy is shown here to be a viable approach to improve creep resistance at 800 °C, across a wide range of stresses, for both regimes of diffusional and dislocation creep. This indicates that the dispersoids not only lead to dispersion strengthening by interacting with mobile dislocations in the \( \gamma/\gamma' \) grains but also enhance, when located at or near grain boundaries, resistance to diffusional creep due to grain boundary sliding. This deformation mechanism relies on the motion of dislocations along or adjacent to the boundary, via a combination of dislocation climb and glide, with a strain rate proportional to the square of stress \( (n = 2) \) and inversely proportional to the grain size.\(^{[50]}\)

The small volume fraction (0.5 wt%) of \( Y_2O_3 \) dispersoids achieves a reduction in creep rate by a factor of \( 8 - 10 \), at a given stress or doubling in stress at a given minimum strain rate. The dispersoids do not change the dominant creep mechanism nor the minimum strain rate at which the transition occurs \( (\approx 5 \times 10^{-8} \text{s}^{-1}) \) but they shift the creep curves between base and ODS alloys along the stress axis; this further demonstrates that the combination of \( \gamma' \) precipitates and oxide dispersoids is synergistic, rather than the most potent obstacle (precipitate or dispersoid) being dominant. While synergistic strengthening from two populations of second phases has been utilized in the past for advanced ODS powder metallurgy superalloys, fabricated by sintering, it is shown here for an AM superalloy processed through rapid melting and solidification of powders.
The AM ODS Ni–Cr–Al–Ti superalloy, studied here as a model material, shows lower compressive creep rates compared with cast binary Ni–8Al (n = 5.8) but creeps faster than the highly alloyed commercial superalloy In792 (n = 8.8) (Figure 5c). In792 contains solid-solution strengtheners for both the γ matrix (Cr, Co, W) and γ′ precipitates (Ta) in combination with a high Al and Ti content as well as elements that segregate to the grain boundaries (B, Zr) to avoid grain boundary sliding; in contrast, the model quaternary alloy studied here relies solely on the oxides dispersoids for additional strengthening beyond the γ′ precipitates. The presence of a low-stress regime with n≤2 indicates that the grain boundaries in our ODS Ni–Cr–Al–Ti model alloy remain the weakest point at low stresses; this can be addressed by obtaining larger grain sizes and/or by adding grain boundary strengtheners such as B and Zr. Despite the microstructural changes induced by the ODS addition, oxide slaggling, grain refinement after PBF, and increased cracking (the latter being eliminated by a proper HIPing procedure) (57), the ODS alloy demonstrates improved creep resistance in compression. It is thus expected that ODS strengthening can also be achieved with other commercial alloys processed by PBF, such as CM247LC (from which the model Ni–Cr–Al–Ti alloy was derived) or Al-free alloys (i.e., Hastelloy X) to prevent Y₄Al₂O₉ coslagging.

4.2. Orientation Anisotropy

Both base and ODS alloys display increased creep rates when deformed along the horizontal direction compared with the vertical direction, with respect to the original build direction. While creep rate anisotropy can be influenced by the crystallographic orientations of the grain structure in regard to the testing direction, that is, testing parallel to <100>, <111> or <110> grains, the <100> grain texture in our study is oriented along all three printing directions x, y, and z (47). This creep rate anisotropy likely originates from the anisotropic, partially recrystallized microstructure (combining coarse, vertically elongated grains and pockets of aligned fine grains), reminiscent of the as-processed microstructure. When loaded along their short direction, the large grains and fine-grain pockets show a comparably higher number of grain boundaries perpendicular to the load path. The two tested directions, vertical and horizontal, thus represent the extremes in terms of microstructure alignment with the load direction. Furthermore, the observed creep rate anisotropy indicates that, despite extensive solutionizing (which results in significant recrystallization) and aging, the initial texture created during AM creates anisotropy during creep deformation. To achieve isotropic creep properties, the alloys might require heat treatment optimization to increase the grain size, eliminate the fine-grain regions, and reduce texture through more extensive recrystallization. However, the creep anisotropy is not necessarily undesirable in parts that can be built along their loading axis, for example, conventional nickel-based superalloys show this as well (6,52). However, for multiaxial or unknown stress states and for geometries that need to be rotated off-axis within the PBF build chamber, this anisotropy may become an engineering challenge that applies to many other PBF alloys operating at high temperatures.

4.3. Grain Boundary Cracking

Grain boundary cracking is the main creep damage mechanism for the present alloys. Existing defects (i.e., gas pores, sintered cracks, and embedded slag) do not initiate damage if they are located within grains. Under uniaxial compressive stress, the specimen must accommodate radial expansion. Due to the elastic and plastic anisotropy of the alloy and friction at the specimen faces, tensile stresses develop onto vertical grain boundaries, while grain boundaries oriented ∼45° to the loading axis are subjected to shear and horizontal grain boundaries are under compressive stresses. Similar observations were made by Estrin et al. (51). The PM1000 ODS exhibits cracking of transverse grain during tensile test, thus perpendicular to the tensile load. The nature of the stress thus dictates which grain boundaries are susceptible to cracking. With a highly strengthened γ/γ′ matrix, the grains resist deformation at low stresses, which in turn is accommodated by movement along the grain boundaries. To reveal the influence of the dispersoids, the alloy studied here is a Ni–8.5Cr–5.5Al–1Ti model Ni-based superalloy, free of additional solid solution strengtheners (e.g., Co, W), γ′ strengtheners (e.g., Ta), or, crucially, grain boundary strengtheners (e.g., B, Zr).

Without the presence of grain boundary strengthening of intermetallic phases, carbides and/or borides, the grain boundaries provide little resistance to deformation. This is also evident from the extended stress range where grain boundary sliding is observed. Based on the results obtained here, future alloy development for AM ODS Ni-based superalloys must incorporate strategies for additional grain boundary strengthening, beyond that provided by the oxide dispersoids.

5. Conclusion

The compressive creep resistance of an AM high-γ′ Ni–Cr–Al–Ti model alloy, with and without 0.5% Y₂O₃ additions, was investigated at 800°C to elucidate the effect of oxide dispersoids on the high-temperature creep properties and the microstructure of these alloys.

When compared with the base, non-ODS superalloy, the ODS superalloy exhibits 8–10 times slower creep rates at any given stress over a wide range of stresses ranging from 35 to 250 MPa. Two distinct creep regimes are observed for both ODS-free and ODS alloys, with the creep exponents allowing to identify diffusional creep, speculated to be due to grain boundary sliding (n = 2) and dislocation glide (n = 5) at strain rates below or above ∼5 × 10⁻⁸ s⁻¹, respectively. In the ODS alloy, this transition point is shifted to higher stress. For both alloys, the creep resistance is anisotropic with regard to the orientation of the applied stress, which arises from the elongated grain structure aligned with the build direction. This microstructural anisotropy originates from the solidification step and is not fully removed during recrystallization.

The results obtained here on a model quaternary superalloy demonstrate that ODS modification of additively manufactured commercial Ni-based superalloys can provide additional strengthening at high temperatures. The Y₂O₃ dispersoids, however, do not provide a sufficiently strong grain boundary strengthening effect to avoid grain separation. Grain boundary
stabilizing elements must thus be added to achieve a successful ODS Ni-based superalloy for AM processing.

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Conflict of Interest

The authors declare no conflict of interest.

Data Availability Statement

The data that support the findings of this study are available from the corresponding author upon reasonable request.

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additive manufacturing, creep, laser powder bed fusion, nickel superalloy, oxide dispersion strengthening

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