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Characteristics and Mechanical Properties of Polycrystalline CM 247 LC Superalloy Casting

Hsin-Erh Huang*1 and Chun-Hao Koo*2

Department of Materials Science and Engineering, National Taiwan University, Taipei, 106 Taiwan, R. O. China

The CM 247 LC superalloy was remelted and cast to obtain the desired polycrystalline test bars by controlling casting parameters, followed by the investigation of precipitation morphology and mechanical properties. The experimental results show that by well-controlled casting parameters the CM 247 LC owns excellent castability to form a superalloy with fine grain structure, high strength as well as superior creep resistance.

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1. Introduction

A forged disc of fine grains has tensile and fatigue properties superior to those of a cast disc. Conversely, the rupture and creep properties of a cast airfoil are superior to those of a forged airfoil. Cast DS or single crystal airfoils and forged or P/M discs are currently bonded together to yield both benefits. Unfortunately, some problems are associated with the "dual-property wheel". The aggressive environment of a turbine engine can lead the turbine wheel to crack in the high stress regions. These cracks are attributed to a low cycle fatigue (LCF) mechanism either thermally or mechanically induced. The bonded areas between the wheel rim or hub and blade are particularly sensitive to crack, resulting from the difference of the grain sizes at the bonding interface. Production of a turbine blade/disk rotor as a single piece casting for small gas turbine engine would offer a significant cost reduction, compared with "dual-property wheel", and may avoid certain stress problems caused by the bonding of blade attachment[1] An integral casting process is used herein to produce the desired fine grain hub and cast airfoils with optimal properties.

Three economical processes, Fine Grain Process (FGP), Grainex and Microcast-X, have been developed to produce integrally cast turbine wheels with fine grain structures.[1–4] Fine Grain Process and Microcast-X are the methods of the variable casting parameters (VCP), and Grainex is a dynamic method. Fine grain alloy has more uniform properties, chemical composition and precipitate morphology than conventionally cast alloys, and provide several advantages, including superior tensile properties, fatigue properties, weldability, and inspectability.

Integrally cast fine grain turbine wheels have been used extensively in small turboprop and auxiliary power gas turbine engines for many years.[1] These wheels have been extensively cast with IN 713 LC, IN 718, and Mar-M247 nickel-base superalloys.[1,4,5] Usually, the turbine disk undergoes moderate temperature/high stress and the airfoil undergoes high temperature/low stress in the operational environment of a gas turbine. Integral turbine wheels must combine the properties of both a blade and a disk. Therefore, the rigid requirements of increasing the operating temperature of integral turbine wheels, increasing rotational speeds and extending the lives of the components forces designers of turbine engine to raise their criteria for selecting alloys.

CM 247 LC is a cast nickel-base superalloy with low carbon content.[6–8] The alloy is a modified superalloy with a chemical composition of Mar-M247, specifically designed for making the directionally solidified (DS) turbine blade. Unlike the parent Mar-M247 superalloy,[5,9–11] the modified alloy exhibits an exceptional resistance to grain boundary cracking during DS casting, combined with improved carbide microstructure and ductility. Cannon-Muskegon Corporation[10] found that the CM 247 LC superalloy is also suitable for casting fine grain integral wheels. However, the effect of microstructures and grain size on the tensile strength and creep performance of polycrystalline CM 247 LC superalloy casting are still equivocal.

The variable casting parameters (VCP), producing a controlled polycrystalline structure of test specimens of CM 247 LC superalloy were employed for this work. The objective of this paper is to determine the microstructural characteristics and investigate the tensile and creep performance of the polycrystalline CM 247 LC casting. Furthermore, some mechanisms associated with microstructural changes and improved properties are discussed.

2. Experimental Procedure

The alloy used in herein was originally melted by vacuum induction melting (VIM) at the Cannon-Muskegon Corporation in USA. The alloy was remelted in an alumina crucible in a VIM furnace and poured at 1653∼1713 K into a ceramic shell mold with a CoO facecoat, at Chung Shan Institute of Science and Technology in Taiwan. The ceramic shell mold was maintained at 1373 K. Table 1 presents the nominal composition of the alloy, which is compared with that of the Mar-M247 alloy. After casting, all test bars were subjected to hot isostatic pressing (HIP) at 1458 K/173 MPa/4 h in an atmosphere of argon. The specimens were then heat-treated in a vacuum. The heat treatment processes consisted of

*1Graduate Student, National Taiwan University

*2Corresponding author, E-mail: chkoo@ccms.ntu.edu.tw

Table 1 Nominal Composition (mass%).

Alloys	C	Cr	Co	W	Mo	Ta	Al	Ti	Hf	B	Zr	Ni
Mar-M247	0.15	8.4	10.0	10.0	0.7	3.0	5.5	1.0	1.5	0.015	0.05	BAL
CM 247 LC	0.07	8.1	9.2	9.5	0.5	3.2	5.6	0.7	1.4	0.015	0.015	BAL

solution-treatments at 1494 K/2 h, followed by argon gas fan quenching (GFQ), and subsequently aging treatments at 1044 K/20 h, and furnace cooling (F.C.). The solutions used to etch the grain boundary and the microstructure were 70% Marble's + 30% Alcohol and 30 mL lactic acid + 10 mL HNO3 + 5 mL HCl, respectively. The grain size was measured using the linear intercept method. The microstructure was elucidated by optical microscopy (OM) and scanning electron microscopy (SEM). Tensile tests were performed using an Instron 1125 mechanical testing machine. The creep test was conducted at 1033 K/725 MPa and 1255 K/200 MPa using SATEC M3 creep testers. The gauge size of the test bars was 6.3 mm in diameter by 25.4 mm long.

3. Results and Discussion

3.1 Grain size and microstructures

The microstructures shown in Fig. 1 are for CM 247 LC superalloy under various casting conditions. The grain sizes in Figs. 1(a–b) are maintained at 80∼90 µm and 150∼200 µm, by controlling the pouring temperature at about 20 K and 40 K, respectively, above the liquidus. At a pouring temperature about 80 K above the liquidus, a transition to conventional dendritic microstructure occurred, after which the grain size rapidly increased. Figure 1(c) shows that the grain size is increased to 2∼3 mm by controlling the pouring temperature to about 80 K above the liquidus. The experimental results indicate that, by controlling casting parameters, the microstructural grains of CM 247 LC superalloy can be modified and refined, from conventional columnar grains of 6∼20 mm to equiaxed grains of 80∼90 µm. The alloy produced by the VCP process, which results in a substantially cellular and nondendritic microstructure, achieves an average grain size of ASTM 3-5.

The low pouring temperature caused the liquid metal in this study to be more viscous than that in conventional casting, allowing pores to form in the casting structure. Therefore, castings produced by the VCP process also must undergo Hot Isostatic Pressing (HIP) to eliminate all possible micropores.[2)]

Figure 2 depicts the typical as-cast microstructure of equiaxed grain CM 247 LC superalloy. The structure consists of γ' intermetallic phase dispersed in a γ matrix. The primary γ-γ' eutectic phase is mainly at the grain boundaries, with only a small portion within the grains. The as-cast MC carbide has a irregular shape. Figures 3(a)–(b) show the morphology of the γ-γ' eutectic phase, the γ' phase and MC carbide following HIPing and heat treatment at 1494 K for 2 h

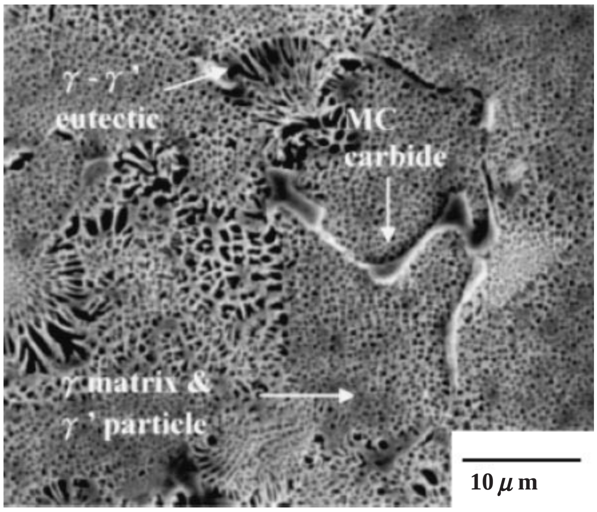

Fig. 2 SEM micrograph showing the as-cast microstructure of fine grain CM 247 LC superalloy (80∼90 µm).

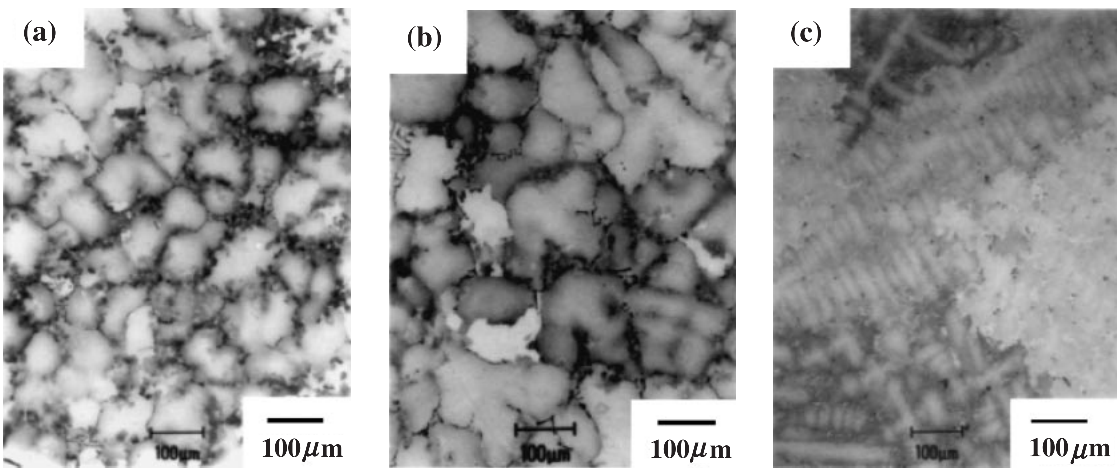

Fig. 1 Optical micrographs showing the grain size of CM 247 LC at various casting temperature: (a) 80∼90 µm, at a pouring temperature about 20 K above the liquidus. (b) 150∼200 µm, at a pouring temperature about 40 K above the liquidus. (c) 2∼3 mm, at a pouring temperature about 80 K above the liquidus.

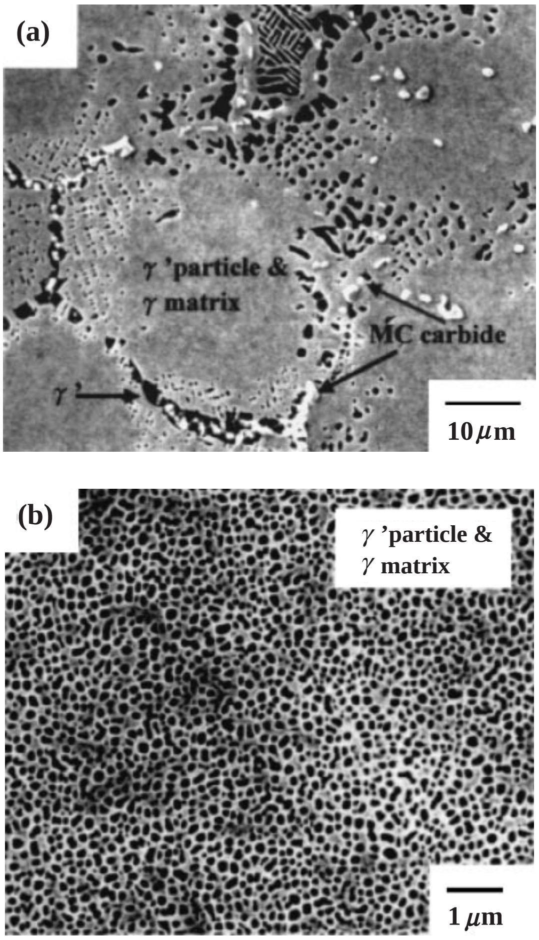

Fig. 3 SEM micrographs showing the distribution and size of γ' phase of CM 247 LC superalloy after heat treatment (HIPing + 1494 K/2 h + 1044 K/20 h) and the observation at (a) grain boundary, and (b) the center of grain.

GFQ + at 1044 K for 20 h F.C. Partial solution-treatment at 1494 K for 2 h reduces the content of the coarse γ-γ' eutectic phase. Figures 2 and 3(a) present the carbide microstructures in the as-cast superalloy and in the heat-treated alloy specimens, respectively, illustrating that the polycrystalline CM 247 LC castings exhibit a low density of carbides with a tendency toward script-like carbide morphology in the as-cast condition. Restated, the content of script-like carbides is less than that in the conventional alloys. Moreover, Fig. 3(a) reveals that the morphology of the carbides changes from script to globular and discrete after heat treatment at 1494 K for 2 h GFQ + at 1044 K for 20 h F.C. The VCP process is not essential to affect the carbide morphology at pouring temperature of 20 to 80 K above the liquidus. The change of carbide morphology from script to globular and discrete after heat treatment at 1494 K/2 h GFQ + 1044 K/20 h F.C. was observed not only in the fine grain CM 247 LC but also in the coarse grain CM 247 LC. These thermal cycles also facilitate the decomposition of primary MC carbides, which may transform to more stable chemical compounds. Chemical analysis conducted in this research (Table 2) by SEM/EDS indicates that the MC carbides presented in the as-cast CM 247 LC superalloy are predominantly tantalum/titanium-rich carbides, and are transformed by thermal processing into more stable hafnium/tantalum carbides. The Hf/Ta-rich

Table 2 The SEM-EDS analysis of carbide in fine grain CM 247 LC superalloy.

Condition	GB carbide	Matrix carbide
Casting	Ta, Ti-rich	Ta, Ti, Hf-rich
	(MC-2, 1)	(MC-2, 1, 3)
HIPing	Ta, Hf, Ti-rich	Ta, Hf-rich
	(MC-2, 3,1)	(MC-2, 3)
Solution	Ta, Hf-rich	Hf, Ta-rich
Treatment	(MC-2, 3)	(MC-3, 2)

carbides typically have a more discrete morphology and frequently are wrapped by the γ' phase. Furthermore, relatively small discrete Hf-rich carbides precipitate in the regime of the primary γ-γ' eutectic phase. Most of the carbon in most nickel-based superalloys below 1255 K is in the form of MC carbide at high temperature. However, during heat treatment, MC decomposes slowly to produce carbon, which permeates the alloy and triggers a number of important reactions. The dominating carbide reaction in many alloys is the formation of $M_{23}C_6$ by the following reaction:[12)]

$$MC + \gamma \rightarrow M_{23}C_6 + \gamma' \tag{1}$$

or

$$(Ti,Mo)C + (Ni,Cr,Al,Ti) \rightarrow Cr_{21}Mo_2C_6 + Ni_3(Al,Ti) \tag{2}$$

Reaction 1 or 2 occurs at about 1255 K or even as low as 1033 K. Under isolated conditions, the reaction is reversible. M_6C is formed similarly,

$$MC + \gamma \rightarrow M_6C + \gamma' \tag{3}$$

or

$$(Ti,Mo)C + (Ni,Co,Al,Ti) \rightarrow Mo_3(Ni,Co)_3C + Ni_3(Al,Ti) \tag{4}$$

The carbon content of CM 247 LC superalloy is reduced to half of its original value in Mar-M247 to improve the microstructure and stability of the carbide, illustrating that the polycrystalline CM 247 LC exhibits a low density of carbides and a fine carbide morphology (Fig. 2). Additionally, the W and Mo levels of the alloy are slightly reduced to compensate for the lowering C and Ti levels and to thereby minimize the formation of deleterious secondary $M_{23}C_6$ platelets, μ phase or alpha W platelets or needles due to the degeneration or breakdown of primary carbides during high temperature exposure. The Cr is lowered from 8.4 to 8.1% and combined with the lowering of Ti, W and Mo levels more than compensates for the lowering of carbon to further ensure CM 247 LC alloy being free from sigma (σ) phase and M_6C carbide. Solution heat-treatment procedures must be optimized to stabilize the carbide morphology. High-temperature exposure may cause extensive carbide degeneration, resulting in grain-boundary carbide saturation and compromised mechanical properties. The above mentioned situation in fine grain CM 247 LC may not be serious because of the carbon content of CM 247 LC superalloy is reduced to half of its original value in Mar-M247. Clearly, after heat treatment, fine grain CM 247 LC superalloy exhibits neither $M_{23}C_6$ nor M_6C formation (Table 2). Therefore, the dominant carbide reaction in polycrystalline CM 247 LC superalloy proceeds as follows during heat treatment.

$$\text{MC-1,2,3} + \gamma \rightarrow \text{MC-1,3} + \gamma' \quad (5)$$

or

$$\text{(Ta,Ti,Hf)C} + \text{(Ni,Al,Ti)} \rightarrow \text{(Ta,Hf)C} + \text{Ni}_3\text{(Al,Ti)} \quad (6)$$

where MC-1, MC-2 and MC-3 represent TaC, TiC and HfC, respectively.

3.2 Tensile properties at 300 K and 1033 K

The properties of conventional cast nickel-base superalloys are principally functions of the content and morphology of the γ' phase, grain size and carbide distribution. Table 3 and Fig. 4 present the tensile properties of CM 247 LC superalloy with various grain sizes. The yield strength at room temperature increases from 939 MPa for test bars with coarse grains of size 2∼3 mm to 1010 MPa for test bars with equiaxed grains of size 80 µm. Furthermore, the ultimate strength at room temperature increases from 1069 MPa for test bars of grains of size 2∼3 mm to 1242 MPa for test bars of equiaxed grains of size 80 µm. Similar results are also obtained for increased elongation from 5.7 to 8.1% when grain sizes are reduced from 2∼3 mm to 80 µm. In this research, the fine grain CM 247 LC superalloy exhibits higher ultimate strength, yield strength and elongation over those values of coarse grain CM 247 LC superalloy at room temperature.

Table 3 Tensile test results of the CM 247 LC superalloys with various grain sizes.

Grain size	Test Temp. (K)	U.T.S. (MPa)	Y.S. (MPa)	Elongation (%)
EMS-55447	RT	>966	>725	>4.0
80∼90 µm	RT	1242	1010	8.1
150∼200 µm	RT	1192	985	7.2
2∼3 mm	RT	1069	939	5.7
80∼90 µm	1033	1200	1016	8.2
150∼200 µm	1033	1160	990	7.8
2∼3 mm	1033	1110	960	5.1

These results can be explained by the Hall-Petch equation; $\sigma = \sigma_0 + Kd^{-1/2}$, where σ is the yield stress; σ_0 is the yield stress of a single crystal; K is a constant, and d is the grain size. Grain size strongly affects the strength of polycrystalline CM 247 LC alloy. For an alloy with coarse grains, edge dislocations may pile up at grain boundaries, increasing the number of stress concentrations, such that the coarse grains hinder the accommodation of internal stress by the alloy and increase the resistance to deformation. The ability of fine grains to spread the deformation decreases the stress concentration at the grain boundary, such that a cavitated grain boundary can not be easily de-bonded. Therefore, the strength and elongation of a tensile test specimen increase as the size of the grains is reduced at room temperature.

At a temperature of 1033 K, the equiaxed grain structure

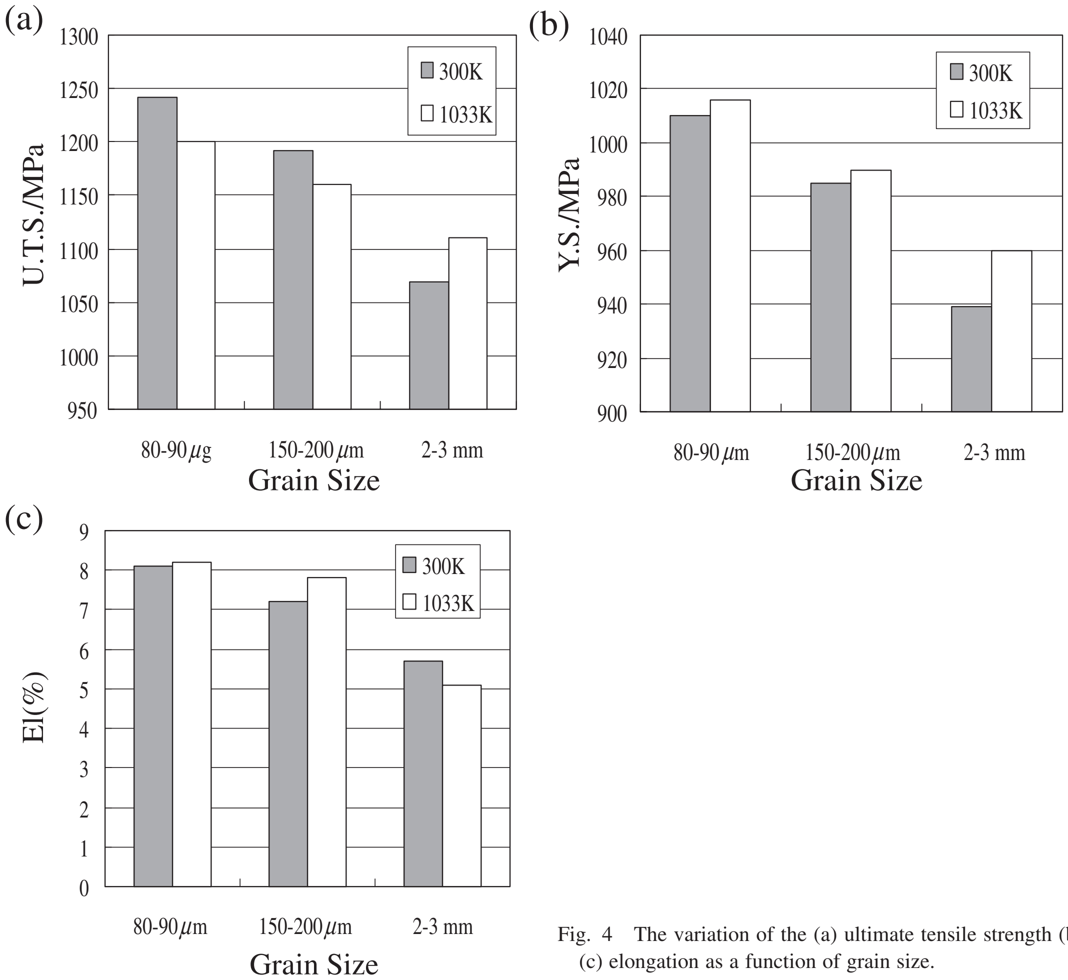

Fig. 4 The variation of the (a) ultimate tensile strength (b) yield strength (c) elongation as a function of grain size.

retains its strength and elongation. The yield strength of CM 247 LC increases notably as the grains become smaller. As shown in Table 3 and Fig. 4, the yield strength increases from 960 MPa for test bars with coarse grains of sizes 2∼3 mm to 1016 MPa for test bars with fine grains of size 80 µm; the elongation increases from 5.1 to 8.2% as the grain size is reduced from 2∼3 mm to 80 µm. Similar results are also obtained as the ultimate strength increases from 1110 to 1200 MPa when the grain size is reduced from 2∼3 mm to 80 µm. Thus, tensile test results reveal that the test bars of fine grains have higher yield strength, ultimate strength and elongation than those of coarse grains at either room temperature or 1033 K. Both tensile strength and elongation change only slightly as the temperature rises to 1033 K for any specified grain sizes. Notably, equiaxed CM 247 LC superalloy after heat treatment is stronger than equiaxed Mar-M247, IN 713 LC, and IN 718 nickel-base superalloy.[1,4,5] The strength of grain boundaries determines the high-temperature strength of polycrystalline alloys, because the grain boundary is weaker than the inside of the grain at elevated temperature. Heat-treated CM 247 LC does not exhibit any σ, μ or M_6C plate formation, and no detrimental effect on either strength or ductility of the alloy is apparent. Under typical operating condition (1033 K) of the hub, fine grain CM 247 LC alloy with fine globular MC carbide at the grain boundaries may retard rapid crack propagation and inhibit grain boundary sliding. Therefore, fine grain CM 247 LC (80 µm) is more elongated than large grains at either room temperature or 1033 K, because crack propagation takes longer in a fine grain structure than coarse-grain structure. Consequently, refining the grains by controlling the temperature of pouring drastically improves the tensile properties of polycrystalline CM 247 LC at room temperature and 1033 K. Briefly, experimental results demonstrate that a fine grain CM 247 LC casting may improve the operating performance of small gas engine with high tensile strength and high elongation at the operating temperatures of the hub.

3.3 Creep test at 1033 K/725 MPa

CM 247 LC superalloy has no particular polycrystalline specification. According to Honeywell EMS-55447, the rupture life of a polycrystalline casting superalloy must be longer than 23 hours and the elongation must not be lower than 2% at 1033 K/725 MPa. EMS-55447 is one of the materials specifications for the mechanical properties of Mar-M 247 superalloy. Table 4 shows that reducing the grain size can remarkably enhance the creep life and ductility at 1033 K/725 MPa. In particular, when the grain size is 80 µm, the rupture elongation and life reach 4.2% and 95 hours, respectively. According to the creep deformation map of a CM 247 LC-related superalloy,[13] the creep mechanism at 1033 K and 725 MPa is dislocation glide; that is, the creep behavior is dominated by the interaction between dislocations and obstacles, such as grain boundaries, precipitates or carbides. Hence, refining grains by controlling the pouring temperature greatly improves the creep life of polycrystalline CM 247 LC at 1033 K/725 MPa.

Table 4 The creep test results of CM 247 LC superalloy at 1033 K/725 MPa.

Grain size	Creep life (h)	Elongation (%)	1% Creep life (h)
EMS-55447	>23	>2	—
80∼90 µm	85∼95	3.3∼4.2	36
150∼200 µm	70∼81	3.3∼3.6	29
2∼3 mm	51∼72	3.0∼3.4	27

Usually, conventional cast nickel-base superalloy is grain-boundary-strengthened by carbide precipitation at grain boundaries, so carbide morphology and distribution are crucial in determining the creep behavior and ductility of superalloys. However, coarse or large script-like carbides at grain boundaries may cause moderate temperature brittleness in nickel-base superalloys.[14] Cannon-Muskegon modified Mar-M247 into DS CM 247 LC alloy. The primary modification in the alloy design was the reduction of the carbon content by about one-half to improve the microstructure of carbides. Figure 3 shows that the polycrystalline CM 247 LC castings exhibit a low density of script-like carbides with a tendency toward blocky or discrete carbide morphology after heat treatment, which is beneficial to the retarding of dislocation glide and grain boundary sliding.

Carbide size is important, and reduced carbide volumes and sizes in nickel-base alloys result in a reduction in precracked carbides. The MC carbides, distributed as discrete particles in polycrystalline CM 247 LC superalloy, may still provide sites for the nucleation of microvoids during deformation, but their shape and distribution may not provide preferred paths for crack propagation. A mixed crack propagation mode was observed in both coarse and fine grain microstructures (Fig. 5). The crack propagation mode for the coarse grain structure is found to be principally interdendritic and transgranular, while in the case of a fine grain structure, it is principally intergranular and transgranular. When the alloy is heat-treated for a long time, MC-type carbide may degenerate to produce grain-boundary carbides, and the grain boundary and carbides are coated with a layer of γ' phase (Fig. 3(a)), which may be relatively soft at service temperatures,[15] allowing cross-slip during grain movement, preventing the initiation of fracture through the enhanced local ductility. In contrast, if carbides precipitate as a continuous grain-boundary film, properties can be severely degraded. $M_{23}C_6$ films were reported[16] lowered impact resistance of M252, and MC films were blamed for lowered rupture lives and ductility in forged Waspaloy alloy.[16] At the other extreme, when no grain-boundary carbide precipitate is present, premature failure also will occur, because grain-boundary movement essentially is unrestricted, leading to subsequent cracking at grain-boundary triple points.[16] Under such circumstances, crack initiation is observed at the coarse γ' phase or the γ-γ' eutectic phase, which have superior ductility.[17] Therefore, the fracture mode change from the carbide/matrix interface of the conventional cast nickel-base superalloys[14] to the γ-γ' eutectic region of the polycrystalline CM 247 LC superalloy may cause substantial suppression of the crack initiation and propagation and enhance the rupture life and elongation in the superalloy.

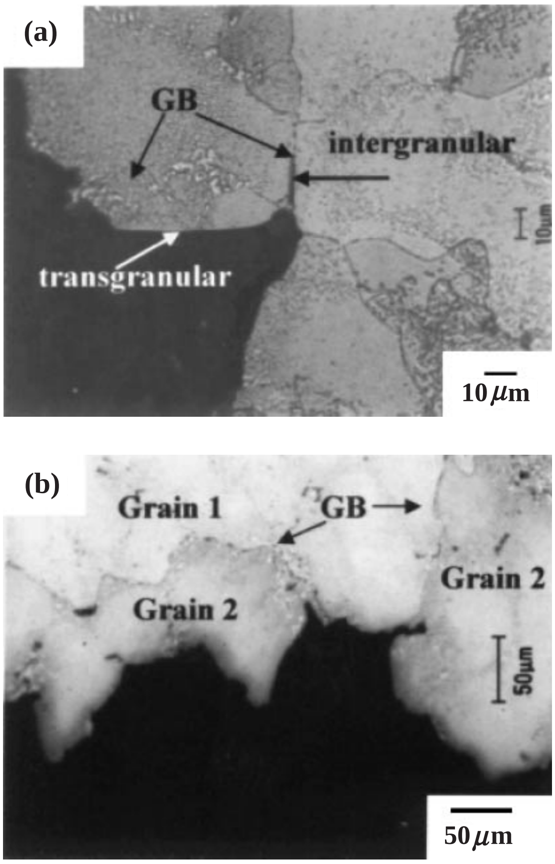

Fig. 5 Optical micrographs of cross-sections of CM 247 LC superalloy tested under 1033 K/725 MPa: (a) fine grain structure (80∼90 μm); (b) coarse grain structure (2∼3 mm).

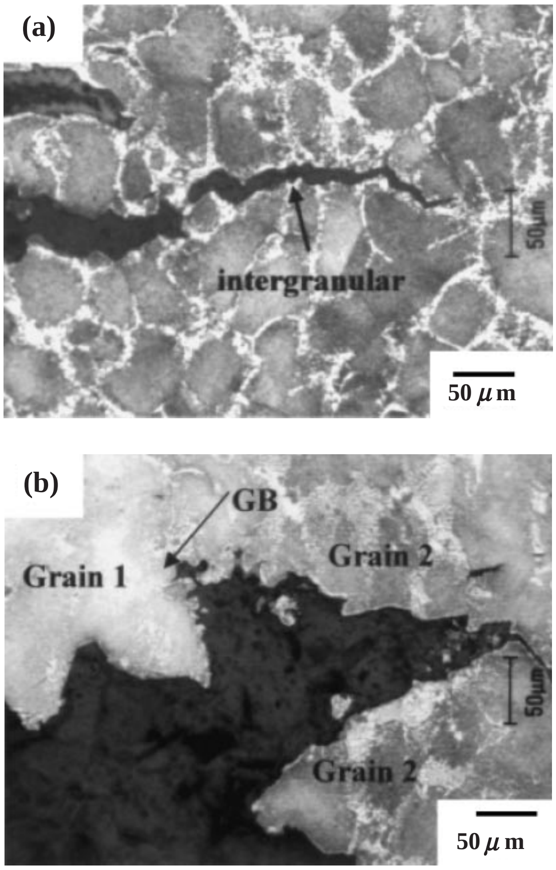

Fig. 6 Optical micrographs of cross-sections of CM 247 LC superalloy tested under 1255 K/200 MPa: (a) fine grain structure (80∼90 μm); (b) coarse grain structure (2∼3 mm).

Table 5 The creep test results of CM 247 LC superalloy at 1255 K/200 MPa.

Grain size	Creep life (h)	Elongation (%)	1% Creep life (h)
EMS-55447	>25	>4	—
80∼90 μm	45∼52	8.9∼9.5	30
150∼200 μm	66∼72	8.0∼8.4	39
2∼3 mm	100∼105	6.0∼6.3	49

3.4 Creep test at 1255 K/200 MPa

According the specification of Honeywell EMS-55447, the rupture life of a polycrystalline casting superalloy at 1255 K/200 MPa must exceed 25 hours and the elongation should not be lower than 4%. As indicated in Table 5, under the creep condition of high temperature and low stress (1255 K/200 MPa), the coarse grain CM 247 LC alloy exhibits a better rupture life (105 h) than that of fine grain CM 247 LC alloy (52 h). Observation of the fracture surface of the cross-sections (Fig. 6) indicates that most of the cracks are formed along grain boundaries at 1255 K/200 MPa in fine grain specimens. However, a mixed crack propagation mode is still observed in a coarse grain microstructure. The crack mode associated with a coarse grain structure is determined to be principally interdendritic and transgranular at 1255 K/200 MPa. According to the creep deformation map of a CM 247 LC related superalloy,[18] the creep mechanism at 1255 K/200 MPa is dislocation climb and grain boundary slip. Therefore, the fine grain structure promotes grain boundary sliding and exhibits a poor resistance to crack propagation. However, the CM 247 LC has a higher volume fraction of the γ' phase approximately 65 vol% precipitated in the matrix, than that of conventional polycrystalline superalloy.[9] The strengthening of nickel-based superalloy clearly depends on the volume fraction of γ'. Moreover, blocky carbide particles at grain boundaries act as "dispersive" obstacles to retard grain-boundary sliding and form barriers to crack propagation, resulting in the extension of the high-temperature rupture life and the improvement in creep rate. Briefly, the experimental results imply that the fine-grain CM 247 LC (80–90 μm) promotes the performance of small turbine engines with high tensile strength and elongation at operating temperatures of the hub for burst protection with superior creep life under the operating conditions of the blade.

4. Conclusions

(1) The experimental results demonstrate that, by controlling casting parameters, the microstructural grains of CM 247 LC superalloy can be modified and refined from conventional columnar grains of 6∼20 mm into equiaxed grains of 80 μm.
(2) This work reveals that solution-treatment at 1494 K changes the morphology of the γ-γ' eutectic phase and

MC carbide, resulting in partially dissolving the coarse γ' and γ-γ' eutectic phases without any incipient melting.

(3) Tensile test results reveal that the fine grain test bars (80 μm) have a higher yield strength and a higher elongation than those of coarse grains (2–3 mm) at either room temperature or 1033 K.

(4) The study also shows that the creep behavior of the fine grain CM 247 LC superalloy is much better than that of coarse grain superalloy at 1033 K/725 MPa because the creep mechanism is dislocation glide. The rupture life and creep rate of coarse-grain superalloy is higher and lower, respectively, than those of fine grain superalloy at 1255 K and 200 MPa because the creep mechanism is dislocation climb.

(5) Creep test results indicate that fine grain CM 247 LC superalloy exhibits longer creep life and higher ductility than those of coarse grain CM 247 LC superalloy at 1033 K/725 MPa, and also meets the requirement of turbine blade at 1255 K/200 MPa. The strong performance of the modified superalloy is attributed to the high volume fraction of the precipitated γ' phase and refinement of carbide, consequently retarding dislocation gliding and climbing.

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