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Journal of Alloys and Compounds 455 (2008) 215–220
Journal of ALLOYS AND COMPOUNDS
www.elsevier.com/locate/jallcom

Effects of cooling rate on the microstructure and mechanical properties of the Ni-base superalloy UDIMET 500

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Received 11 September 2006; received in revised form 18 January 2007; accepted 18 January 2007
Available online 21 January 2007

Abstract

The Ni-base superalloy UDIMET 500 indicates good hot corrosion resistance, high stability and strength at high temperatures and for this reason the alloy is used in manufacturing of gas turbine hot components. UDIMET 500 is employed in manufacturing of the second stage blades of high-power gas turbines. The microstructure consists mainly, of austenite γ-matrix and ordered and coherent γ′ precipitates and also carbides dispersed in the matrix. The alloy gains its appropriate microstructure and strength through precipitation hardening mechanism. The heat treatment cycle consists of three stages: homogenization, partial solution and aging. Heat treatment parameters such as: time and temperature of homogenization, partial solution and aging temperatures, and cooling rate from homogenization and solution temperatures affect the microstructure of the alloy. Among these parameters cooling rate from partial solution is the most effective. Therefore, in this research the effects of cooling rate on microstructure and mechanical properties such as: tensile, creep and hardness were investigated. For this purpose, six different cooling rates were applied on the cast UDIMET 500 specimens after partial solution at 1080 °C for 4 h. Microstructures of the specimens then were studied using optical and electron microscopy. Tension and creep tests were performed at different conditions. It was found out that with increasing cooling rate the volume percent of the γ′ precipitates decreases. Also, it was shown that size, shape and volume fraction of primary γ′-particles are largely influenced by the cooling rate following homogenization and partial solution treatments.

Keywords: Superalloy; UDIMET 500; γ′ Precipitate; Heat treatment; Microstructure

1. Introduction

The Ni-base superalloys with complex and multi-phase microstructures are stable at high temperatures and this characteristic is the main reason for using these materials in critical and severe service conditions.

The Ni-base superalloy UDIMET 500 is used in manufacturing of the second stage blades of high-power gas turbines. The blades work at critical condition of creep, corrosion and fatigue for more than 90,000 h. The alloy requires good physical, mechanical and corrosion properties because of its severity of service conditions.

Considering chemical composition and microstructure of UDIMET 500, it is expected that the alloy with γ′ and carbides exhibits good corrosion resistance and stability at high temperatures [1]. Structural stability of UDIMET 500 has been investigated previously [2].

The alloy consists of carbide-forming, γ′-forming and refractory elements. The roles of the alloying elements in Ni-base superalloys have been discussed in several works [3–7]. The alloy contains refractory elements such as Mo, Cr and Co to prevent local hot corrosion.

UDIMET 500 has a multi-phase microstructure consisting of γ matrix, bimodal γ′ precipitates, γ–γ′ eutectic, carbides and a small amount of deleterious phases such as: δ and σ [1,8]. The γ′ phase is a superlattice possessing the $L1_2$-type structure with a nominal composition of $Ni_3(Al,Ti)$. The primary γ′ cuboids with average edge length of 0.25 μm are produced

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doi:10.1016/j.jallcom.2007.01.091

Table 1
Standard heat treatment cycle for UDIMET 500 [1,9]

Full annealing (homogenization)	1150 °C/(4h air cool)
Partial solution	1080 °C/(4h air cool)
Aging	760 °C/(4h air cool)

during solidification under 1200 °C. Standard heat treatment cycle for UDIMET 500 has been described in literature. Table 1 illustrates the parameters of the cycle [1,9].

The parameters of precipitation hardening heat treatment affect distribution of alloying elements and precipitates and also, their morphologies and volume fractions. Precipitation heat treatment of the alloy consists of two main stages: solution treatment and aging. Solution treatment causes precipitates to dissolve and alloying elements to homogenize in microstructure [10].

Solution heat treatment may be carried out at over 1150 °C. This type of solution is called full solution or homogenization. During homogenization all phases except TiC are dissolved [1,9]. Higher temperatures may have deleterious effects such as: local melting of eutectic and reduction of primary γ' percentage which in turn changes mechanical properties of the blades. Therefore, to prevent the deleterious effects, the solution treatment temperature is decreased. This type of treatment is called partial solution. The partial solution temperature of an alloy depends on the tendency of the alloy to the formation of harmful phases such as: η, σ, and Laves. In an alloy like IN738LC, that this tendency is low, the partial solution temperature is low (1120 °C). However, in an alloy like IN939 having high percentage of Co and Ti and high tendency to the formation of η phase, the treatment is carried out at 1160 °C [11,12]. In the case of UDIMET 500 the partial solution temperature is 1080 °C [1,9]. During partial solution some part of the large γ' precipitates remains undissolved. Bhowal et al. [13] showed that large γ'-particles are lower at higher solution temperatures and their sizes depend on cooling rate after solution treatment. The other effects of partial solution treatment are developing $M_{23}C_6$ carbides in grain boundaries due to the decomposition of MC carbides [1]. Therefore, full solution treatment is a necessary step in heat treatment of the alloy.

Aging treatment in Ni-base superalloys is applied for nucleation and growth of secondary γ' precipitates. Balikci et al. [14] proposed two mechanisms for growth of γ' precipitates: coalescence of the small particles to the larger ones and extraction of dissolved elements, like Al and Ti, from the saturated solid solution matrix and to the precipitates. During aging treatment volume fraction, morphology and distribution of γ'-particles are determined. The heat treatment causes formation of separated $M_{23}C_6$ carbides in the matrix between γ'-particles and also at the grain boundary. Overaging causes precipitate particles to increase and consequently, decreasing their number and increasing their spacing. The growth of the particles causes decreasing creep resistance of superalloys [15].

Table 2
Chemical Compositions of UDIMET 500 superalloy (in wt.%)

Ni	Bal.
Cr	17.9
Co	17.2
Ti	3.0
Al	3.0
Mo	4.0
Fe	2.1
C	0.11

Since the mechanical properties of the alloy are due to its microstructural characteristics and these characteristics are affected by heat treatment, determination of the effects of heat treatment parameters on microstructure is worth of studying.

2. Experimental procedures

The chemical composition of the Ni-base superalloy UDIMET 500 was determined by optical emission spectroscopy. The chemical composition is presented in Table 2. Effects of cooling rate from partial solution temperature on the microstructure of the alloy were investigated. First, some cast specimens from UDIMET 500 ingots were prepared. The specimens were heat treated at different cycles. The temperature was measured through a thermocouple attached to the specimens was used. The different cooling rates applied to the specimens are presented in Table 3. The different cooling rates were selected in the range of furnace and air cooling conditions. The rates were measured using the slope of temperature–time curve at 850 °C for each series of specimens. Microstructures of all samples were examined using optical and electron microscopy. Quantitative analysis of the microstructures was performed by image analyzer.

Constant load–rupture test was carried out according to ASTM E139 [16] at 925 °C and 172MPa. Room and high temperature tensile tests at a constant strain rate of $10^{-4}\,s^{-1}$ were performed on 42 specimens according to ASTM E8 and E21 [17,18]. The temperature was measured with two thermocouples placed on the specimen gage length. The temperature variation of the furnaces during creep and tensile tests was about ±1 °C.

Table 3
Specifications of the different heat treatment cycles

Cycle	Homogenization			Partial solution			Aging		
	Cooling rate (°C/min)	Time (min)	Temperature (°C)	Cooling rate (°C/min)	Time (min)	Temperature (°C)	Cooling rate (°C/min)	Time (h)	Temperature (°C)
A	Air cool	240	1150	(Air cool) 186	240	1080	Air cool	16	760
B	Air cool	240	1150	129.5	240	1080	Air cool	16	760
C	Air cool	240	1150	90.35	240	1080	Air cool	16	760
D	Air cool	240	1150	49.8	240	1080	Air cool	16	760
E	Air cool	240	1150	35.4	240	1080	Air cool	16	760
F	Air cool	240	1150	(Furnace cool) 3.85	240	1080	Air cool	16	760

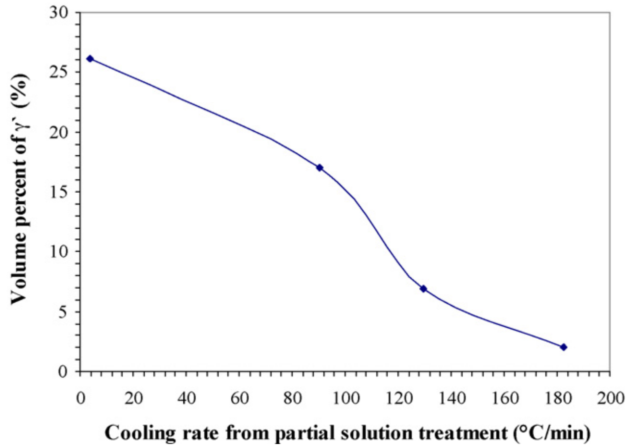

Fig. 1. Weight percent of γ′ precipitates versus cooling rate from partial solution temperature.

3. Results and discussion

Heat treatment is an important step in manufacturing and repairing of gas turbine blades. The aim of heat treatment is producing a suitable microstructure and properties in the alloy and the coating. Improper heat treatment causes formation of deleterious phases, unsuitable properties, crack initiation and oxidation during service.

High-temperature strength of Ni-base superalloys depends mainly on the volume fraction and morphology of γ′ precipitates. Precipitation hardening heat treatment affects these properties.

Measurement of γ′-weight percent in specimens heat treated at different cycles shows that with increasing cooling rate the γ′-weight percent decreases. Fig. 1 shows the relation between γ′-weight percent and cooling rate from partial solution temperature. It is a well-known fact that rapid cooling after solution treatment reduces the number and size of γ′-particles formed during cooling. These particles are so fine that even during long term aging could not grow up or their magnitude could increase. On the other hand, slow cooling causes more precipitation and coarser particles.

Microstructural study of the heat treated specimens confirms the above mentioned results. The microstructure shown in Fig. 2 is a representative of the specimen heat treated by cycle A. The microstructure consists of fine γ′-particles distributed uniformly in the matrix. Weight percent of the precipitates was measured to be 2%.

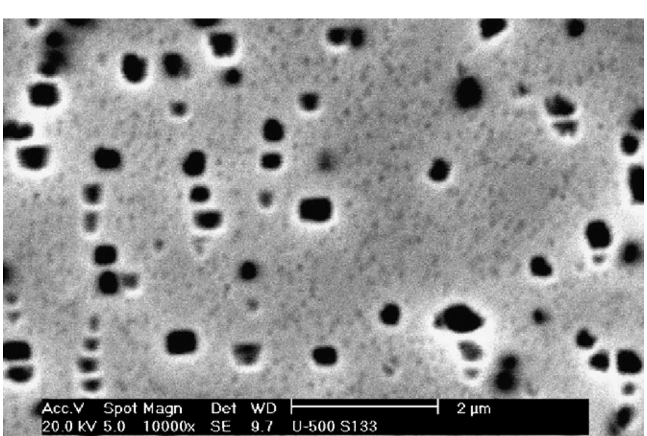

Fig. 2. Microstructures of specimen heat treated with cycle A.

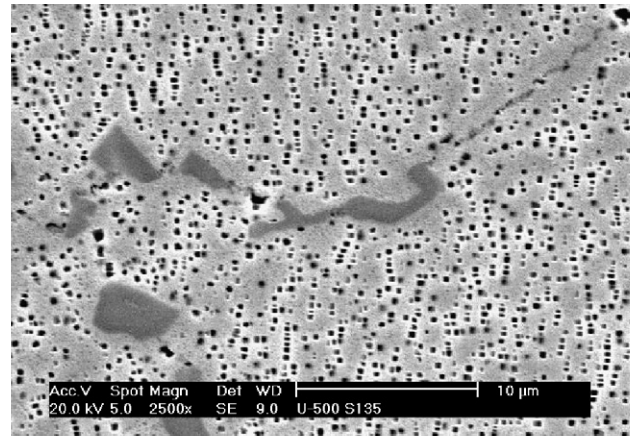

Fig. 3. Carbides precipitated at grain boundary.

Carbides are distributed at grain boundaries and also within the grains (Figs. 3 and 4). Chemical analysis of carbides showed that the carbides formed within grains are mainly MC type carbides, in which M is substituted for Ti. The carbides observed at grain boundaries are $M_{23}C_6$ type carbides, in which M is substituted for Cr and Mo. The serrated grain boundaries prevent grain boundary sliding and enhance creep strength [19].

The γγ′ eutectic is distributed at dendrite boundaries, often near micropores, showing that it is formed during the last stages of solidification.

The superalloy UDIMET 500 is strengthened by two main strengthening mechanisms: solid solution hardening and precipitation hardening. Elements such as: Mo, Ti and Cr are the most potent solid solution strengtheners. Ni, Ti and Al, are γ′ formers and together with significant amounts of Mo and Co strengthen the alloy through a precipitation hardening mechanism. Some factors such as: coherency strains at γ–γ′ interface, elastic moduli difference between γ′ and γ matrix, lattice mismatch, long range ordering of γ′ and anti-phase boundary (APB), produced during the movement of dislocations through γ′-particles,

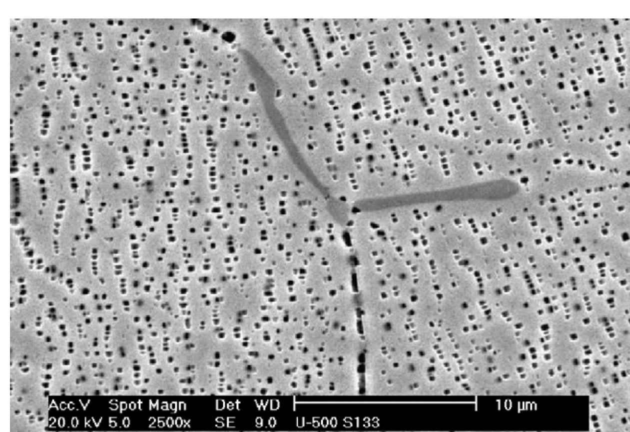

Fig. 4. Carbides precipitated in the interior of a grain.

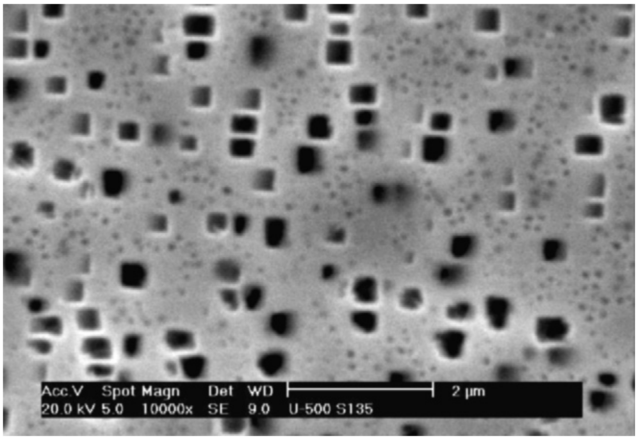

Fig. 5. Microstructure of specimen heat treated with cycle B.

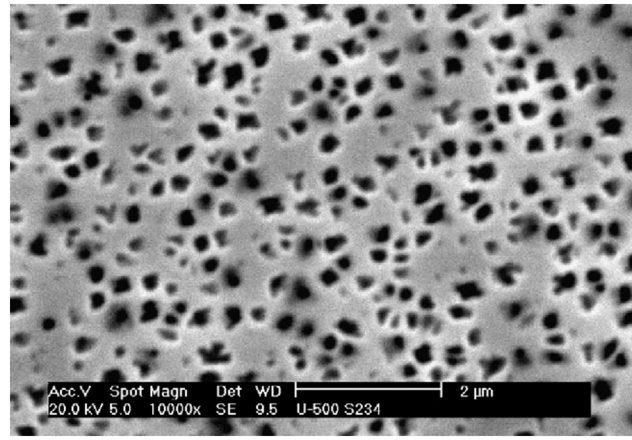

Fig. 8. Microstructure of specimen heat treated with cycle E.

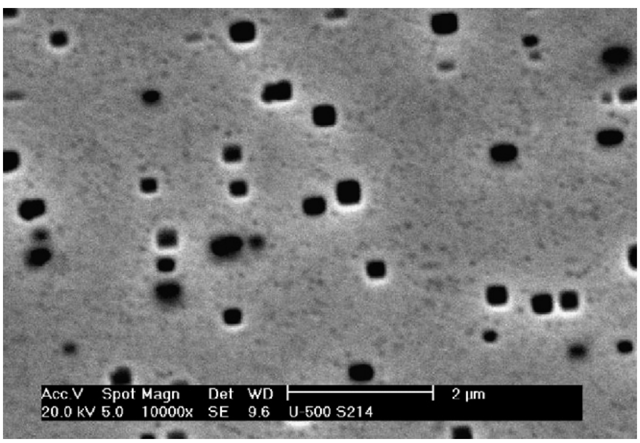

Fig. 6. Microstructure of specimen heat treated with cycle C.

strengthen the Ni-base superalloys via the precipitation hardening mechanism [7,20–22].

Comparing of the microstructures indicates that decreasing cooling rate increases the size and weight percent of γ'-particles. Figs. 5–9 show the microstructures of specimens cooled at different rates from partial solution temperature. One important point found in the study is that fine secondary γ'-particles are formed in the specimen cooled at higher rates. The secondary particles do not appear in the last two cycles with the slowest cooling rate, because during slow cooling they can grow or coalesced with other particles. Inhomogeneous distribution of γ' precipitates in group F specimens is produced due to higher cooling rates from the full solution temperature.

Cooling rate after partial solution treatment also affects particle size, distribution and volume fraction of primary and secondary γ'. Specimens heat treated with cycle F, cooled with lower rate from partial solution treatment posses larger primary γ' with more volume content.

Microstructure reflects the mechanical properties of alloys. The results of tensile tests performed on the specimens heat treated at different cycles show that ultimate tensile strength and yield strength increase with decreasing γ'-particle size and magnitude. γ'-Particles act as barriers against movement of dislocations. Although it is expected that strength of the alloy increases and elongation decreases with increasing weight percent of γ' precipitates. The reverse relations are observed as shown in Figs. 10–12. In fact at high cooling rates γ'-particle size is fine and in spite of the lower magnitude its influence on mechanical properties such as strength and elongation is high.

The difference in γ' volume fraction is due to the higher cooling rate from the partial solution temperature. It was shown that

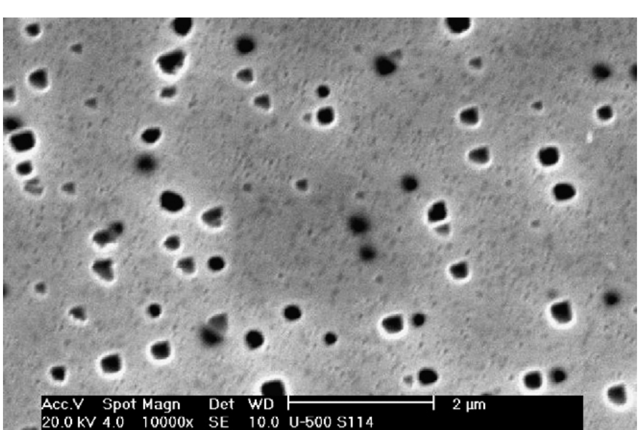

Fig. 7. Microstructure of specimen heat treated with cycle D.

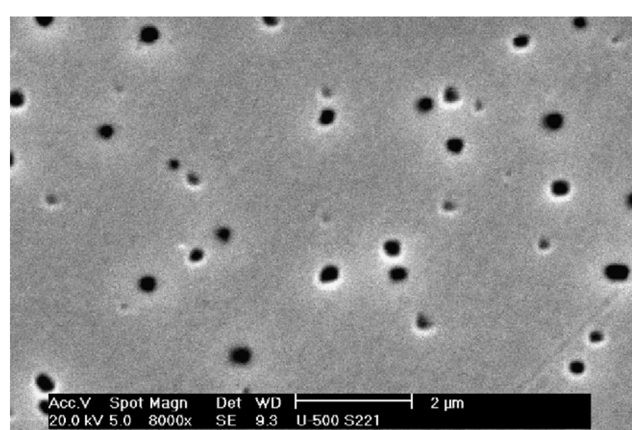

Fig. 9. Microstructure of specimen heat treated with cycle F.

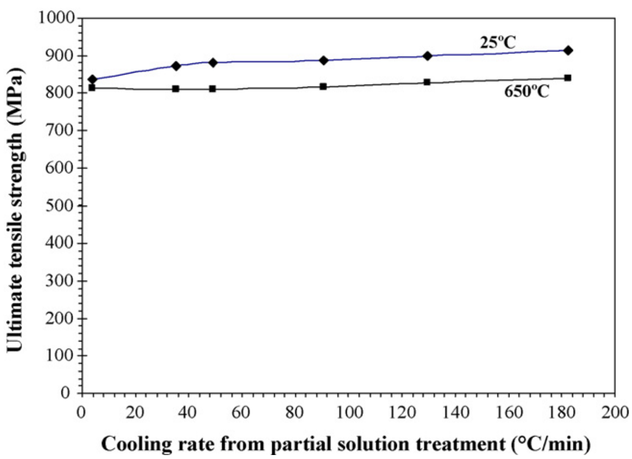

Fig. 10. Ultimate tensile strength vs. cooling rate from partial solution temperature.

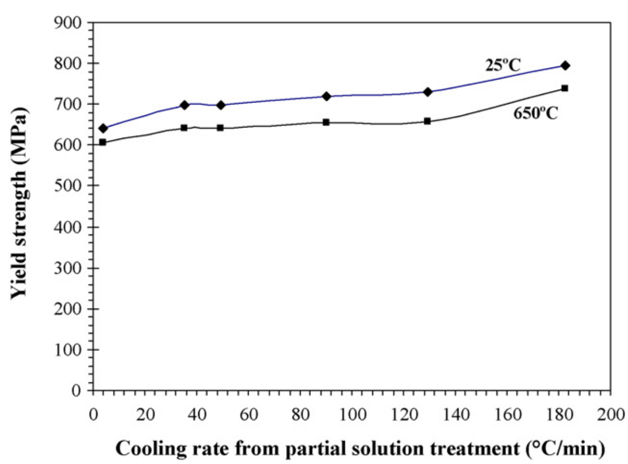

Fig. 11. Yield strength vs. cooling rate from partial solution temperature.

with decrease in cooling rate the primary γ′ size and its volume content increases. Cooling rate after partial solution treatment is an effective parameter in nucleation and growth of γ′ precipitates [23,24]. At low cooling rate, the fine and dispersed γ′-particles joint together and form coarse γ′ precipitates. In conclusion,

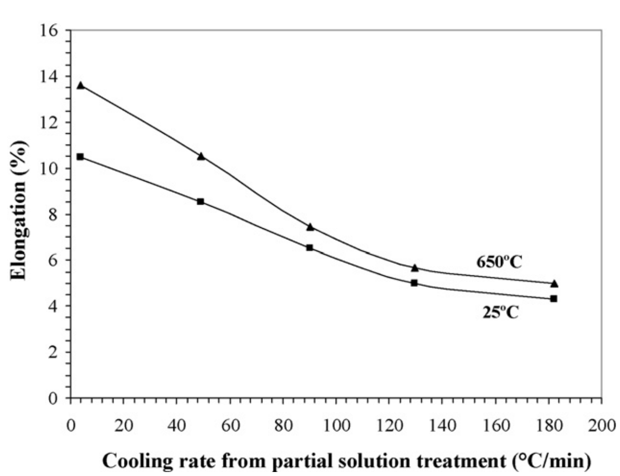

Fig. 12. Elongation vs. cooling rate from partial solution temperature.

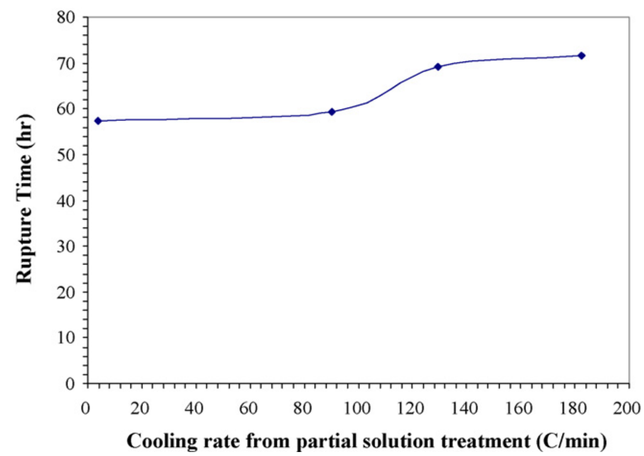

Fig. 13. Rupture life at 925 °C and 172 MPa at different cooling rate from partial solution temperature.

strength increases and elongation decreases. Higher cooling rate, also, causes inhomogeneous distribution of γ′-particles in γ matrix.

The results of stress–rupture test at 925 °C and stress of 172 MPa (Fig. 13) indicates that size of γ′-particles influences rupture life under creep condition, so that with decreasing γ′ dimensions, due to higher cooling rate, the rupture life increases. In fact, fine γ′-particles produced during higher cooling rate are more resistant to stress rupture, although their volume content is lower than the specimens heat treated with slow cooling rate.

4. Conclusions

After partial solution treatment, different cooling rates were applied on UDIMET 500 Ni-base superalloy to examine their influences on microstructure and mechanical properties. The following results were obtained:

(1) Size, shape and volume fraction of primary γ′-particles are largely influenced by the cooling rate following partial solution treatment, so that, as the cooling rate increases the size and volume percent of primary γ′-particles decrease.
(2) Inhomogeneity in the microstructure is observed when the cooling rate increases.
(3) Increasing cooling rate increases strength but decreases ductility.
(4) Improved stress–rupture life was observed at 186 °C/min cooling rate.

Acknowledgements

The authors wish to express appreciation to Mavadkaran Eng. Co. for supporting of this project. Also, Deputy of Research and Technology of Tavanir Co. (Ministry of Energy) is gratefully acknowledged for providing the material.

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