Influences of Cooling Rate After Solution Treatment on Microstructural Evolution and Mechanical Properties of Superalloy Rene 80

Shuang Gao¹ · Jie-Shan Hou¹ · Kai-Xin Dong¹ · Lan-Zhang Zhou¹

Received: 10 July 2016/Revised: 8 September 2016/Published online: 18 October 2016
© The Chinese Society for Metals and Springer-Verlag Berlin Heidelberg 2016

Abstract Influences of the cooling rate after solution treatment on microstructural evolution and mechanical properties of Rene 80 nickel-based superalloy were investigated by optical microscopy, scanning electron microscopy, transmission electron microscopy and mechanical test. Results showed that the high cooling rate decreased the size of secondary γ' particles in the supersaturated matrix, but led to a high coarsening rate of γ' particles during subsequent aging treatment. Despite various coarsening rates, the size and morphology of γ' particles in the final microstructures of all samples were close due to the long enough holding time to an equilibrium state. During the aging of 870 °C/2 h, primary MC started to decompose with the carbide reaction: $MC + \gamma \rightarrow M_6C$ or $M_{23}C_6 + \gamma'$. And a number of observations showed that the coarsening of γ' particles on grain boundaries resulted in the depletion of γ' during aging treatment. The test results indicated that high cooling rate resulted in the presence of quench crack, and the air cooling method following solution treatment was an optimum heat treatment method for Rene 80.

KEY WORDS: Ni-based superalloy; Cooling rate; γ' Precipitates; Ripening theory; Heat treatment; Microstructural evolution

1 Introduction

Nickel-based superalloys have been widely applied to the aerospace field and gas power generation industries due to their excellent high-temperature creep resistance, superior high-temperature corrosion resistance and high-temperature phase stability [1–3]. As service requirements are continuously increasing, the alloying degree of nickel-based superalloys improves gradually [4, 5]. In general, it achieves

the goal of solid solution strengthening by the addition of alloying elements such as Co, Mo and W [6, 7]. The precipitation strengthening of the alloy can be enhanced by adding the formation elements of strengthening phase γ' , e.g., C, Al and Ti [8, 9]. With an excessive increase in alloying degree, microstructural properties were aggravated owing to the formation of as-cast microstructural defects, for example, coarse γ' phases (primary γ'), γ – γ' eutectic, large carbides and detrimental topologically close-packed phases (TCP) which are unfavorable to the properties of alloy [10, 11]. As a leading material for large stationary gas turbine in China, Rene 80 is a precipitation-strengthened cast nickel-based superalloy with a relatively high alloying degree.

For most of the cast alloys, heat treatment includes solution treatment and aging treatment [12–14]. The purpose of solution treatment is to homogenize the as-cast microstructure and eliminate the composition segregation, simultaneously dissolve coarse as-cast γ' particles, γ – γ' eutectic and carbides into the matrix [15]. The role of aging

Available online at <http://link.springer.com/journal/40195>

✉ Lan-Zhang Zhou
lzz@imr.ac.cn

¹ Institute of Metal Research, Chinese Academy of Sciences, Shenyang 110016, China

treatment is to adjust the size and distribution of secondary γ' particles and optimize alloy microstructure and properties [16]. The fine secondary γ' particles are distributed more uniformly than as-cast γ' particles. In the processes of solid solution and aging treatment, carbides also change correspondingly. Most of primary MC carbides usually divide into the stable secondary M_6C or $M_{23}C_6$ carbides [15, 17, 18]. These carbide particles play an important role in hindering the migration of grain boundaries and dislocations and thus improve the creep properties of alloy [9, 19].

At present, the studies on heat treatment of nickel-based superalloys have gained comprehensive attention [20–24]. However, the various parameters of heat treatment have diverse influences on microstructural evolution and mechanical properties. Safari et al. [12] did some investigations on the heat treatment of Rene 80, and they considered that the nucleation rate of secondary γ' precipitates can be improved with increasing cooling rate after aging treatment. For a similar alloy UDIMET 500, Sajjadi et al. [17] held that the size, shape and volume fraction of secondary γ' particles were deeply impacted by the cooling rate following solid solution treatment, and with increasing cooling rate, the size and volume percent of primary γ' particles decreased. Moreover, Bhowal et al. [25] also deemed that the size of the secondary γ' particles depended on the cooling rate after solution treatment. Based on a literature survey, all previous studies focus on the effect of the cooling rate on the solid solution microstructure, but there has little work to be published about the potential effects of the cooling rate on microstructural evolution during subsequent heat treatment and the final microstructures and mechanical properties. It is well known that the parameters of solution treatment not only impact directly the solid solution microstructures and properties, but also have potential influences on the microstructural evolution in subsequent heat treatment. Therefore, this work investigated the influence of the cooling rate after solution treatment on the microstructural evolution and mechanical properties in detail. And the aim is to provide a theoretical basis for designing heat treatment route of the similar chemical composition alloy.

2 Experimental

Table 1 lists the chemical composition of the experimental alloy which was produced by 5-kg vacuum induction melting and casting. The alloy was subjected to heat

treatment with the solution treatment followed by three different cooling methods, corresponding to three cooling rates, as shown in Table 2. Cooling rate was measured by using a thermocouple thermometer. The average cooling rate of water cooling (WC), air cooling (AC) and furnace cooling (FC) is about 138, 20 and 0.5 °C/s, respectively.

Microstructural features were examined by optical microscopy (OM), scanning electron microscopy (SEM), transmission electron microscopy (TEM). Precipitates were analyzed by energy-dispersive spectroscopy (EDS) and selected area electron diffraction (SAED). The size and area fraction of carbides and γ' were measured by using an image analysis software (Image Pro Plus 6.0). At least twenty micrographs from different grains were selected to minimize measurement errors. The specimens were polished and etched with a solution of 10 g $CuCl_2$ + 50 ml HCl + 50 ml C_2H_5OH . Films for TEM observation were mechanically grounded to 50 μm and thinned by twin-jet electropolishing with a solution of 10 ml $HClO_4$ + 90 ml C_2H_5OH at 20 V and -50 °C.

The hardness of the alloy was measured by Rockwell hardness C (HRC) test at room temperature. The HRC measurements were taken on a Rockwell hardness machine using an indentation load of 140 kg, a 120° diamond cone and a dwell time of 10 s. Tensile tests were conducted on the alloy at 25 and 870 °C. Smooth cylindrical specimens have an overall length of 70 mm with gage length and diameter of 25 and 5 mm, respectively.

3 Result

3.1 As-Cast Microstructure

Figure 1 shows the OM and SEM microstructures of as-cast Rene 80. The OM microstructure has obvious dendritic segregation pattern with the average secondary dendrite arm spacing of 93 μm (Fig. 1a). It reveals that as-cast microstructure consists of matrix phase γ , strengthening phase γ' , a limited amount of $\gamma'-\gamma$ eutectic and carbides (Fig. 1b, c). The $\gamma'-\gamma$ eutectics with petal-like morphology, which were solidified as the last parts of the liquid during solidification, were distributed in interdendritic regions while the blocky or script morphology carbides existed both on grain boundaries and within grains (Fig. 1b, d). The γ' average area fraction is as high as 48.3% with the average size of 0.55 ± 45 μm . In addition, it is evidently found that the size of γ' particles in interdendritic regions is larger than that in the dendrite cores (Fig. 1d). Figure 1e gives the EDS result of carbide composition in as-cast microstructures, indicating that it is Ti-rich MC carbide (where M represents metal ion and C represents carbon atom).

Table 1 Chemical composition of the alloy (wt%)

C	Cr	Co	W	Mo	Al	Ti	B	Zr	Ni
0.15	13.98	9.62	4.17	4.02	2.87	4.8	0.013	0.061	Bal.

Table 2 Heat treatment schedule

Sample code	Solution	Aging
WC	1200 °C/2 h/WC	1095 °C/4 h/AC + 1080 °C/4 h/AC + 870 °C/12 h/AC
AC	1200 °C/2 h/AC	
FC	1200 °C/2 h/FC	

3.2 Microstructural Characteristics After Solution Treatment

Figure 2 shows a typical full solution treatment after solution treatment at 1200 °C for 2 h with different cooling methods of WC, AC and FC. As shown in Fig. 2a–c, there are some common microstructural changes in WC, AC and FC samples described as follows: Firstly, original coarse γ' precipitates in as-cast state completely dissolve since the solvus temperature for γ' is about 1150 °C in this alloy [15]; secondly, composition segregation is almost eliminated and thus microstructure uniformity is improved; thirdly, the average size of carbides slightly decreases from 6.7 to 4.9 μm without change in the crystallographic structure. A purpose of the solution treatment is to fully dissolve coarse γ' precipitates and large carbides into the matrix and obtain supersaturated solution matrix for fine strengthening phase precipitating during subsequent aging treatment. Figure 2d exhibits that there are a few minimal γ' particles in WC sample with the highest cooling rate. However, Fig. 2e and f shows that abundant and fine γ' particles (secondary γ') re-precipitate in AC and FC samples. The average size of secondary γ' in AC and FC samples is about 70 and 226 nm, respectively.

3.3 Microstructural Characteristics at Different Aging Stages

To study the potential effects of the cooling rate after solution treatment on the microstructural evolution in subsequent heat treatment, aging treatment at 1095 °C/4 h/AC + 1080 °C/4 h/AC + 870 °C/12 h/AC was performed on WC, AC and FC samples. Figure 3a–i shows the microstructural evolution in three samples after each aging step. It seems that the micrograph of secondary γ' transforms from approximately spherical (ellipsoidal) into cubic; moreover, the γ' size significantly increases. Higher magnification figures show that the microstructure consists of secondary γ' particles and a mass of hyperfine tertiary γ' particles (about 58 nm) which are believed to form during cooling period, as marked by the arrows in Fig. 3. Nevertheless, the amount of tertiary γ' particles

reduces after aging treatment at 870 °C/12 h/AC, especially those nearby secondary γ' particles, as shown in Fig. 3g–i. This is due to the secondary γ' coarsening in a mode of absorbing tertiary γ' particles [12]. Table 3 provides the statistical data of secondary γ' precipitates at different aging treatment stages. It indicates that the γ' average size in WC, AC and FC samples increases gradually from 295 to 388 nm, from 290 to 402 nm and from 316 to 395 nm, respectively. The γ' area fraction increases, respectively, to 24.7, 29.2 and 26.5%, showing that three samples almost achieve a close value after the whole heat treatment.

3.4 Evolution of the Carbides and Grain Boundaries

Carbides and grain boundaries in three samples have similar changes during the heat treatment. As an example, the changes of AC sample are exhibited here. Figure 4a, b reveals that after the high-temperature aging of 1095 °C/4 h/AC + 1080 °C/4 h/AC, the morphology and size of carbides do not obviously change compared with solid solution sample. These carbides distribute on grain boundaries or in grain interiors. In addition, according to EDS analysis, it is suggested that the type of carbides does not change and it is identified to be the Ti-rich MC carbides. However, it is noted that carbides are decomposed during low-temperature aging at 870 °C/12 h/AC. Furthermore, two kinds of bright precipitates around primary MC carbides in the grain as marked in Fig. 4c are significantly observed by backscattered scanning electron morphology (BS-SEM). Moreover, the EDS results (Table 4) reveal that the phases are rich in Mo, W and Cr. Furthermore, according to the selected area electron diffraction pattern (SAED) (Fig. 5a), it is determined that the lattice parameter, a_0 , of Mo-rich carbides is 1.13 nm coinciding with that of M_6C carbides [26]. This is also confirmed by TEM observation (Fig. 5a). It is clear to observe another fine and black precipitates around MC carbide on the grain boundary (Fig. 5b), and EDS data (Table 4) reveal that the phase is rich in Cr. In addition, the SAED confirms that the lattice parameter ($a_0 = 1.06$ nm) of Cr-rich carbides coincides with that of $M_{23}C_6$ carbides [27]. Thus, from the above results, the primary MC starts to be decomposed into M_6C or $M_{23}C_6$ during the low-temperature aging.

Figure 6a together with Fig. 4a–c show that the morphology of grain boundaries almost keeps serrated and decorated with carbides and γ' precipitates in the whole process of heat treatment [28, 29]. Additionally, it is interesting to find that the γ' precipitates on the local grain boundary aggregate and coarsen with γ' depleting on both sides of grain boundary (Fig. 6b).

Fig. 1 Microstructures of the as-cast experimental alloy and EDS analysis: **a** OM image showing the dendritic microstructure; **b** SEM micrograph showing MC carbides morphology in as-cast microstructure; **c** morphology of γ' precipitates; **d** SEM micrograph showing γ - γ' eutectic in detail; **e** EDS spectrum of Ti-rich MC carbide

Fig. 2 Effects of the cooling methods after solution treatment on the microstructures: **a–c** microstructures of WC, AC and FC samples, respectively, **d–f** corresponding γ' image of WC, AC and FC samples, respectively

4 Discussion

4.1 As-Cast Microstructure and Solution Microstructure

Rene 80 is a γ' strengthening-type alloy with a high alloying degree. The addition of alloying elements causes serious dendritic segregation in as-cast microstructure (Fig. 1a). In the process of solidification, refractory elements segregate to the dendritic core, such as W and Mo, while some elements with low melting point segregate to the interdendritic region, such as Al and Ti which are γ' formation elements [30]. The enrichment of Al and Ti in the interdendritic region promotes γ' growth and $\gamma'-\gamma$ eutectic formation. This is the reason why the size of γ' particles in the interdendritic region is significantly larger than that in the dendritic cores (Fig. 1d). In addition, $\gamma'-\gamma$ eutectic, which forms at the end of solidification, is enriched in low melting point alloy elements. The existence of $\gamma'-\gamma$ eutectic is detrimental to high-temperature properties of nickel-based superalloys.

After solution treatment at 1200 °C/2 h, the alloy has a typical full-solution-treatment microstructure with fine secondary γ' particles and small carbides. During the cooling process after solution treatment, the cooling rate has noticeable influence on secondary γ' size in

supersaturated microstructure, as shown in Fig. 2. It reveals that the γ' size increases with the decrease in cooling rate. It is well known that the driving force of phase transformation is the free energy change, ΔG , before and after phase transformation. Additionally, ΔG can be expressed as a function of the supercooling degree, ΔT , which has a directly proportional relationship with cooling rate, following the formula $\Delta G = \Delta H \cdot \Delta T / T$ [31], where ΔH is the latent heat of phase transformation which is a positive value for γ' phase transformation. It indicates that there is a directly proportional relationship between ΔG and ΔT . In consequence, a fast cooling rate contributes to rapid nucleating of γ' particles, but inhibits growth of γ' particles, as shown in Fig. 2b, which shows denser and smaller γ' particles in AC sample compared with that in FC sample. However, when cooling rate is dramatically rapid, solute atoms diffusion capacity is limited to form the γ' particles during cooling process, as illustrated in WC sample which consists of rare γ' particles (Fig. 2a). When cooling rate is excessively slow, a small supercooling degree leads to a low γ' nucleation rate, but promotes the growth of γ' particles since there is enough time for solute atoms to diffuse sufficiently during the cooling process. As shown in Fig. 2c, the average size of secondary γ' particles in FC sample is larger, but the density is lower.

Fig. 3 SEM micrographs of γ' precipitates in three samples at different aging heat treatment stages: **a–c** at 1095 °C/4 h/AC; **d–f** at 1080 °C/4 h/AC; **g–i** at 870 °C/12 h/AC. *Insets in upper right show hyperfine tertiary γ' particles in the matrix as marked by the arrows*

Table 3 Statistical size and area fraction of secondary γ' precipitates at different aging heat treatment stages

Heat treatment stage	WC		AC		FC	
	d (nm)	Area fraction (%)	d (nm)	Area fraction (%)	d (nm)	Area fraction (%)
1	295 ± 24	15.7	290 ± 34	16.5	316 ± 54	14.7
2	345 ± 46	19.8	355 ± 51	17.8	372 ± 35	15.9
3	391 ± 36	24.7	402 ± 21	29.2	415 ± 24	26.5

1—Aging heat treatment at 1095 °C/4 h/AC; 2—aging heat treatment at 1080 °C/4 h/AC; 3—aging heat treatment at 870 °C/12 h/AC

4.2 Microstructural Evolution

During the aging treatment below γ' solvus temperature, secondary γ' precipitates will nucleate and grow up. The

morphology and size of γ' precipitates change significantly with extending holding time. As shown in Fig. 3, the morphology of secondary γ' precipitates in three specimens transforms from circular into cubic. This indicates that the

Fig. 4 SEM microstructures of the carbides and grain boundaries (marked by *dotted lines*) of AC sample in different heat treatment stages of **a** 1095 °C/4 h/AC, **b** 1080 °C/4 h/AC, **c** BS-SEM microstructure after aging of 870 °C/12 h/AC

Table 4 EDS analysis of different carbides

	<i>MC</i>		<i>M₆C</i>		<i>M₂₃C₆</i>	
	wt%	at.%	wt%	at.%	wt%	at.%
Al	0.7	1.7	1.9	6.2	0.9	2.1
Mo	26.0	19.3	44.5	39.9	9.9	6.3
Ti	42.7	63.5	4.2	7.5	2.5	3.2
Cr	1.6	2.2	14.0	23.4	51.9	60.4
Co	0.3	0.3	1.0	1.4	2.8	2.8
Ni	2.2	2.7	5.7	8.3	21.1	21.7
W	26.5	10.2	28.8	13.5	10.9	3.6

cooling rate after solution treatment has no effects on the morphology change in secondary γ' precipitates. In addition, the secondary γ' precipitates noticeably coarsen. It is generally accepted that the process of secondary γ' coarsening obeys the Lifshitz–Slyozov–Wagner theory (LSW), which means that the coarsening is controlled by diffusion [32]. Therefore, the coarsening of secondary γ' is described by the cube rate law [33]:

$$r^3 - r_0^3 = Kt, \quad (1)$$

where r is the secondary γ' average size after aging time t . r_0 is the initial average size at the beginning of coarsening process and, K is defined as the secondary γ' growth rate.

The value of K is measured by Formula (1) on the basis of statistical data of secondary γ' precipitates at different aging stages (Table 3). In order to analyze the impact of cooling rate after solution treatment, the relationship between secondary γ' growth rate and cooling rate after solution treatment is shown in Fig. 7. It is obvious to find that cooling rate has potential influences on K during different aging stages. The secondary γ' of WC and AC sample has a similar γ' growth rate which is higher than that in FC sample at aging of 1095 °C/4 h/AC. This is attributed to the formation of larger initial γ' precipitates in

FC sample which leads to the consumption of massive formation elements and the decrease in supersaturation degree in solution microstructure. In most of superalloys, Al and Ti are the main formation elements of γ' , which exist as substitution atoms in γ channel. As shown in schematic diagram (Fig. 8) without the consideration of lattice distortion, the amount of Al and Ti around smaller γ' precipitates is more than that around larger γ' precipitates. Accordingly, the amount of formation elements diffusing into larger γ' precipitates is less than that into smaller γ' during the same time. As a result, the growth rate of large secondary γ' during aging treatment is lower than that of small secondary γ' . However, the growth rates of secondary γ' in different samples are very close during aging treatment at 1080 °C/4 h/AC and 870 °C/12 h/AC, which is because the saturation of formation elements in all specimens reduces to a similar degree after previous aging treatment. And the statistical data in Table 3 also demonstrate that microstructural features have no significant difference after aging of 1080 °C/4 h/AC and 870 °C/12 h/AC. In addition, lower temperature contributes to the decrease in atom diffusing capability in the matrix. As a result, the γ' growth rate at 870 °C/12 h/AC is lower compared to that at 1095 °C/4 h/AC. This indicates that K value is primarily dependent on temperature.

It is interesting to find from Table 4 that the γ' area fraction is increased by 63.5% during the aging at 870 °C/12 h/AC, but the γ' average size is merely enhanced by 6.2% in FC sample. As shown in Fig. 3, fine γ' obviously grows up while large γ' precipitates almost remain unchanged during low-temperature aging. This indicates that the growth of fine γ' precipitates predominantly leads to the increase in γ' area fraction. The result is consistent with the report [21], which illustrates a conclusion that low aging temperature can suppress the growth of coarse γ' precipitates, but promote the growth of fine γ' precipitates. Furthermore, it is found that both the size and the area fraction of γ' particles in AC sample increase slightly

Fig. 5 TEM microstructures and SAED patterns of the carbides on grain boundaries of AC sample at 870 °C/12 h/AC aging: **a** M_6C topography, M_6C SAED pattern (inset in upper right corner) and MC SAED pattern (inset in lower left corner); **b** $M_{23}C_6$ topography and $M_{23}C_6$ SAED pattern (inset)

Fig. 6 SEM images: **a** γ' coarsening on grain boundary and γ' depleting on both sides of grain boundary after 870 °C/12 h/AC, **b** corresponding higher magnification image

comparing to those of WC and FC samples after long-term aging treatment (Table 3). Because the holding time of low-temperature aging is long enough, the size and area fraction of γ' phases will reach the constant values to keep concentration stable at a given aging temperature. Therefore, it is concluded that the cooling rate after solution treatment has striking influences on the microstructural evolution in subsequent heat treatment, but has no

noticeable effects on microstructural features after the overall heat treatment.

4.3 Evolution of Carbides and Grain Boundaries

The as-cast Rene 80 microstructures consist of a lot of primary MC carbides (Fig. 1b) which are the most stable carbides [21]. During solution treatment and high-

Fig. 7 Growth rate (K) of γ' particles of three samples during the aging treatment

temperature aging treatment, the MC size slightly reduces without decomposition mainly because the holding time remained relatively short. But during low and long temperature aging, the MC carbides start to decompose into relatively stable M_6C or $M_{23}C_6$ carbides (Figs. 4c, 5) because MC is chronically kept in thermodynamic instability state. Due to a long holding time, C atoms have enough time to diffuse and react with other metal elements, such as Cr or Mo. The carbide reaction is as follows: $MC + \gamma \rightarrow M_6C \text{ or } M_{23}C_6 + \gamma'$ [8, 15]. Discrete carbides distributed within grains hinder dislocation moving, and those on grain boundaries prevent the grain boundary moving during high-temperature creep deformation. Therefore, both help to improve high-temperature creep properties.

During low-temperature aging, it is found that γ' phases on local grain boundaries aggregate and coarsen with γ' phases depleting on both sides of grain boundary (Fig. 6). In general, it is considered that γ' depletion is a result of carbides or TCP phases formation at the cost of γ' formation elements [9, 20], but Fig. 6 shows that no carbides and TCP phases exist near γ' depletion zone, while it shows that

γ' particles on the grain boundaries coarsen obviously. Hence, during aging treatment, the fine γ' particles on both sides of grain boundaries have the tendency to diffuse to grain boundaries resulting in the coarsening of γ' particles on grain boundaries. It is well known that grain boundaries are important planar defects with high energy and stay in an unstable state. It indicates that the γ' particles movement contributes to decrease the energy of grain boundary, which keeps the alloy system at a low energy status and ensures the alloy system more stable.

4.4 Mechanical Properties of Finally Gained Microstructure

In order to examine the effect of the cooling rate after solution treatment on the mechanical properties, double-specimen hard tests and tensile tests were performed at 25 and 870 °C, respectively. Table 5 gives the data of yield strength (YS), ultimate tensile strength (UTS), elongation (El) and reduction of area percentage (RA) of WC, AC and FC samples. Each value represents the average value of two repetitions. Generally, the material properties are closely related to microstructural features. The comparison of the hardness and tensile results at 25 °C of three samples indicates that the cooling rate has no influences on the room temperature properties of Rene 80 after the whole heat treatment. These results are coincident with the analysis of the above microstructural evolution. However, the comparison of YS and UTS at 870 °C between AC and FC samples shows that YS and UTS of AC sample is, respectively, elevated by 7.3 and 14.8%, because the size and area fraction of γ' particles in AC sample are a little larger than those in FC sample (Table 3). The El and RA values of AC sample slightly decrease compared with FC sample. Nevertheless, there is an unexpected trouble that the tensile data of WC sample are not gained even by three tensile repetitions. These samples straightway rupture before yield stage when they are tested at 25 and 870 °C. With regard to the SEM observation of fractured surfaces

Fig. 8 Schematic model of solute atom diffusing with different supersaturation degrees: **a** high supersaturation degree, **b** low supersaturation degree

Table 5 Results of mechanical tests for three kinds of samples

Sample code	Tensile test at 25 °C				Tensile test at 870 °C				Hardness (HRC)
	YS (MPa)	UTS (MPa)	El (%)	RA (%)	YS (MPa)	UTS (MPa)	El (%)	RA (%)	
WC	–	520.5	–	–	–	337.5	–	–	41.39
AC	938	1108	6.1	8.2	585	795	18.7	29	41.49
FC	937.5	1092	6.3	8	545	692.5	21	32.25	41.51

Fig. 9 Intergranular cracks as marked by the arrows in SEM micrographs of fractured surface: **a** longitudinal section, **b** WC sample at 25 °C

and longitudinal sections (Fig. 9a, b), it reveals that a brittle intergranular fracture mode with an evidently straight crack initiated from specimen surface, as marked by the arrows in Fig. 9. This is identical to the quench crack under similar conditions [34, 35]. Kane et al. [34] regarded the fracture crack as a result of oxygen-induced dynamic embrittlement and deemed that the reason for this was probably oxygen which was readily available in water used for quenching, either from dissolved air or from a reduction reaction. When samples were cooled rapidly from high temperature, a sharp cooling rate generated the stress concentration to induce the penetration of oxygen into the core of a sharp crack and thus a high enough stress for decohesion to occur. The tiny quench cracks still existed during subsequent heat treatment and finally became a crucial factor of WC sample failure in tensile tests. Therefore, the WC method following solution treatment is not appropriate for Rene 80 alloy while the AC is an optimum option method.

5 Conclusions

1. The high cooling rate decreased the size of secondary γ' particles in the supersaturated matrix.

2. During aging at 1095 °C/4 h/AC, the growth rates of γ' particles in WC and AC sample are higher. During aging at 1080 °C/4 h/AC and 870 °C/12 h/AC, the γ' growth rates in three kinds of samples achieved to a close value.
3. The growth of fine γ' particles is more evident than that of coarse γ' particles during the low-temperature aging of 870 °C/12 h/AC.
4. The aggregation of γ' particles of the grain boundaries contributed to the depletion of γ' particles on both sides.
5. The AC method following solution treatment was an optimal option for gaining relatively excellent elevated temperature mechanical properties.

Acknowledgment This work was supported by the National Natural Science Foundation of China (No. 51571191).

References

- [1] A. Szcotok, Mater. Sci. Eng. **22**, 012007 (2011)
- [2] H. Yang, Z. Li, M. Huang, Modell. Simul. Mater. Sci. Eng. **22**, 85009 (2014)
- [3] Q. Feng, J.Y. Tong, Y.R. Zheng, M.L. Wang, W.J. Wei, H.L. Zhao, X.F. Yuan, X.F. Ding, Mater. China **31**, 9 (2012)
- [4] L. Xu, C. Cui, X. Sun, Mater. Sci. Eng. A **528**, 7851 (2011)

- [5] J. Zhang, J. Li, T. Jin, X. Sun, Z. Hu, *Mater. Sci. Eng. A* **527**, 3051 (2010)
- [6] D.J. Kim, G.G. Lee, D.J. Kim, S.J. Jeong, *J. Mater. Sci. Technol.* **29**, 1184 (2013)
- [7] W.T. Loomis, J.W. Freeman, D.L. Sponseller, *Metall. Trans.* **3**, 989 (1972)
- [8] G. Lvov, V.I. Levit, M.J. Kaufman, *Metall. Mater. Trans. A* **35**, 1669 (2004)
- [9] J.P. Shingledecker, G.M. Pharr, *Metall. Mater. Trans. A* **43**, 1902 (2012)
- [10] X. Yang, S. Li, H. Qi, *Int. J. Fatigue* **75**, 126 (2015)
- [11] H.Y. Li, J.F. Sun, M.C. Hardy, H.E. Evans, S.J. Williams, T.J.A. Doel, P. Bowen, *Acta Mater.* **90**, 355 (2015)
- [12] J. Safari, S. Nategh, *J. Mater. Process. Technol.* **176**, 240 (2006)
- [13] G.A. Rao, M. Kumar, M. Srinivas, D.S. Sarma, *Mater. Sci. Eng. A* **355**, 114 (2003)
- [14] J. Li, R.P. Wahi, *Acta Metall.* **43**, 507 (1995)
- [15] T.J. Garosshen, G.P. McCarthy, *Metall. Trans. A* **16**, 1213 (1985)
- [16] J.S. Hou, J.T. Guo, L.Z. Zhou, Z.J. Li, *Materials for Advanced Power Engineering 2006: Proceedings of the 8th Liège Conference*, Germany, 2006, p. 435
- [17] S.A. Sajjadi, H.R. Elahifar, H. Farhangi, *J. Alloys Compd.* **455**, 215 (2008)
- [18] J.L. Liu, T. Jin, J.J. Yu, X.F. Sun, H.R. Guan, Z.Q. Hu, *Mater. Sci. Eng. A* **527**, 890 (2010)
- [19] K.A. Unocic, J.P. Shingledecker, P.F. Tortorelli, *JOM* **66**, 2535 (2014)
- [20] J.S. Hou, L.Z. Zhou, C. Yuan, Z. Tang, J.T. Guo, X.Z. Qin, P.K. Liaw, *Mater. Sci. Eng. A* **560**, 25 (2013)
- [21] C. Yang, Y. Xu, H. Nie, X. Xiao, G. Jia, Z. Shen, *Mater. Des.* **43**, 66 (2013)
- [22] C. Wang, Y. Guo, J. Guo, L. Zhou, *Mater. Sci. Eng. A* **670**, 178 (2016)
- [23] P. Mao, Y. Xin, K. Han, W. Jiang, *Acta Metall. Sin. (Engl. Lett.)* **22**, 365 (2009)
- [24] Z.X. Shi, S.Z. Liu, J.R. Li, *Acta Metall. Sin. (Engl. Lett.)* **28**, 1278 (2015)
- [25] P.R. Bhowal, E.F. Wright, E.L. Raymond, *Metall. Trans. A* **21**, 1709 (1990)
- [26] Q. Wu, H. Song, R.W. Swindeman, J.P. Shingledecker, V.K. Vasudevan, *Metall. Mater. Trans. A* **39**, 2569 (2008)
- [27] M. Cabibbo, E. Gariboldi, S. Spigarelli, D. Ripamonti, *J. Mater. Sci.* **43**, 2912 (2008)
- [28] M. Huang, L. Zhuo, Z. Liu, X. Lu, Z. Shi, J. Li, J. Zhu, *Mater. Sci. Eng. A* **640**, 394 (2015)
- [29] H.U. Hong, I.S. Kim, B.G. Choi, M.Y. Kim, C.Y. Jo, *Mater. Sci. Eng. A* **517**, 125 (2009)
- [30] F. Wang, D. Ma, A. Bührig-Polaczek, *Sci. Technol. Adv. Mater.* **16**, 025004 (2015)
- [31] M. Shamsuddin, *Physical Chemistry of Metallurgical Processes* (Wiley, New York, 2016), p. 624
- [32] L. Tan, Y. Yang, R.K. Nanstad, J.T. Busby, *J. Phase Equilibria Diffus.* **35**, 524 (2014)
- [33] D. Hadjiapostolidou, B.A. Shollock, *11th International Symposium on Superalloys*, Warrendale, USA, 2008, p. 733
- [34] W.M. Kane, U. Krupp, T. Jacobs, C.J. McMahon, *Mater. Sci. Eng. A* **402**, 42 (2005)
- [35] J. Gayda, P. Kantzos, J. Miller, *Pract. Fail. Anal.* **3**, 55 (2003)