Effect of carbon on the microstructure evolution and mechanical properties of low Si-containing centrifugal casting 20Cr32Ni1Nb alloy

Jingbo Yan^{1,2}, Yuefeng Gu², Yingying Dang¹, Xinbao Zhao¹, Jintao Lu¹, Yong Yuan¹, Zhen Yang¹, Hongfei Yin¹

1. Xi'an Thermal Power Research Institute, Xingqing Road 136, Xi'an, Shaanxi 710032, China

2. High Temperature Materials Unit, National Institute for Materials Science (NIMS), 1-2-1 Sengen, Tsukuba, Ibaraki 305-0047, Japan

Correspondence address:

Dr. Jingbo Yan

Address:

Xi'an Thermal Power Research Institute

Xingqing Road 136

Xi'an, Shaanxi 710032

China

E-mail: yf625oscar@163.com

Tel: +81-080-7961-7170

Fax: +81-029-859-2501

Abstract: The dendrite structure, which is formed during centrifugal casting process, plays an important role in the fracture behavior of 20Cr32Ni1Nb alloy. The large primary carbide facilitates the crack nucleation and propagation along the dendrite interface during high temperature deformation. The secondary NbC precipitates near the dendrite interface during aging at 1120°C, while the major precipitates change to secondary Cr₂₃C₆ when aging at lower temperature. These secondary carbides effectively strengthen the dendrite interface, which is probably due to inhibiting the crack propagation and decreasing the stress concentration. However, the benefit of the secondary carbide reduce when the precipitation free zone form aging at high temperature. Increasing C content enhances the precipitation tendency of Cr₂₃C₆, and facilitates the coalescing of the carbide at elevated temperature. As a result, the cracks can easily nucleate at the interface between NbC and Cr₂₃C₆, and then propagate along the large chromium carbide during deformation. However, the small chromium carbide precipitates form at lower aging temperature, and thus strengthens the alloy effectively.

Keywords: Fracture; Precipitation; Carbide; Crack; Mechanical properties

1. Introduction

For recent years, 20Cr32Ni1Nb alloy has been considered as the most potential candidate material to be used as the reform tube in the petrochemical industry^[1]. It is usually processed by centrifugal casting method, possessing columnar austenitic grain with MX carbide (NbC) distributing at grain boundary and dendrite interface^[2]. Generally, the NbC has been considered as the major strengthening phase in the alloy, and its strengthening effect achieves maximum when the Nb/C is in stoichiometry ratio^[3]. However, the NbC was reported to convert into G phase^[4] (Ni₁₆Nb₆Si₇) during short aging at 700-850 °C^[5-9], leading to the fracture of 20Cr32Ni1Nb alloy at the early stage during service^[10-14]. In recent years, some works proved that the G phase formation can be effectively inhibited when Si content has been restricted within low level^[15, 16]. However, the service performance of low Si-containing 20Cr32Ni1Nb alloy is still unclear.

On the other hand, there is a large difference on the microstructure characterization between the 20Cr32Ni1Nb alloy and the traditional reform tube materials (HP40 and HK40 alloy). The major difference is the low carbon content in 20Cr32Ni1Nb alloy, which leads to the much lower volume fraction of the carbide^[17]. It has been extensively reported that the network chromium carbide develops at the grain boundary of HP40 and HK40 alloy, strengthening the alloy by inhibiting the grain boundary sliding during alloy deforming at elevated temperature^[18, 19]. By comparing, the Cr₂₃C₆ carbide in the 20Cr32Ni1Nb alloy usually form during service, precipitating at both the grain interior and boundary in fewer amount^[20]. The absence of chromium carbide in the cast 20Cr32Ni1Nb alloy is mainly because of the Nb/C ratio satisfies as below^[21]:

$$[Nb]/([C]+[N]) \geq 7 \quad (1)$$

Here, $[X]$ is the weight percentage of the X element in the alloy. Nevertheless, this material possesses competitive strength at elevated temperature, and its superior ductility ensures the higher fatigue fracture resistance^[22]. This indicates that the carbide in the 20Cr32Ni1Nb alloy strengthening in a more effective way, probably because carbide distribution or morphology plays a more important role in their strengthening effect. However, the literature referring to the strengthening mechanism of this material is yet unavailable, especially associating with the centrifugal structure.

In the present work, we investigated microstructure evolution of low Si-containing 20Cr32Ni1Nb alloy after aging at various temperatures, and discussed the effect of aging temperature on the mechanical properties of the materials. Some of the solution strengthening elements, such as tungsten and molybdenum, were induced for further improving the microstructure stability and alloy properties^[23-25]. In addition, a higher carbon modified alloy, which has the lower Nb/C ratio, was also involved in this research. By comparing the precipitation behavior of carbide during solidification and aging duration, the strengthening mechanism of carbide in the 20Cr32Ni1Nb alloy was discussed in this work.

2. Experimental Procedure

Two grades of 20Cr32Ni1Nb alloys were involved in this investigation, and both of them were prepared by centrifugal casting method. The raw materials were melted in a 10-kg capacity medium frequency induction furnace, and poured into a horizontal centrifugal casting furnace when the melted alloy was super-heated to 1650 °C. The ICP-OES was used for the chemical composition analyzing, and the results are shown in Table. 1. Sample 1 is the low Si-containing 20Cr32Ni1Nb alloy, with the Nb/C ratio (in weight percentage) close to 7.7. Sample 2 possesses higher carbon content, which was designed to facilitate the M23C6 precipitation due to the low Nb/C ratio.

The solution and aging treatment has been conducted for analyzing the microstructure evolution, and the detail parameters are shown in Table. 2. All the aged specimens were first homogenization at 1200 °C for 2 hours, and then aging at different temperature. The specimens for microstructure analyzing were taken from the same position of the casting ingot, selected at the position of 7 mm from the tube outer side, and sectioned into 12 mm×12 mm×3 mm. After the aging procedure, the specimens were mounted in conductive resin and grounded on SiC sand paper up to 1000 #, then polished by flannelette for less than 1 min and cleaned in acetone for 10 min. XRD and EDX analyzing were conducted for investigating the phase constitution, and the back scattering electron microscopy (BSE) were used to investigate the microstructure evolution and element distribution. In addition, TEM specimen were prepared for analyzing the precipitates phase morphology and their orientation. They were mechanically thinned down to 50-100 μm in thickness, followed by electro-polished in a solution of 45% acetic acid, 45% 2n-butoxyethanol, and 10% perchloric acid at about -10 °C/30 V.

The tensile test was conducted on both the cast and the heat treated materials, following the China national test standard GB/T 4338-2006. All the specimens were tested at the 750 °C, and they were heated in the rate of 15 °C/min from room temperature, held at the test temperature for 10 min, and then started the test with the strain rate of $3 \times 10^{-4} \text{ s}^{-1}$. SEM has been conducted for analyzing the detail of the fracture surface. The cross section of the fracture surface was prepared for analyzing the deformation mechanism of the alloy.

3. Result and Discussion

3.1 20Cr32Ni1Nb

The microstructure of the centrifugal casting 20Cr32Ni1Nb alloy is shown in Fig. 1. The alloy is composed of austenitic matrix with so-called “Chinese Script” morphology carbide precipitates at dendrite interface and grain boundary (Fig. 1a). XRD analyzing confirms this carbide as NbC, as shown in Fig. 1b. It's usually been considered as the major strengthening phase in the 20Cr32Ni1Nb alloy, primarily because of their good stability during creep at elevated temperature.

The carbide precipitation behavior in various aging temperature is shown in Fig. 2. After aging at 1120 °C for 2 hours (HT1-1 alloy), some fine precipitates are observed close to the primary NbC, with the precipitation size rarely larger than 500 nm (Fig. 2a). According to the XRD analyzing result, these fine

precipitates are deduced to be secondary NbC (Fig. 2d). This has been confirmed by and EDX elemental map analyzing, which show the small Nb rich particle precipitates within matrix (Fig. 3). The nucleation of NbC has frequently been observed in 20Cr32Ni1Nb alloy, and its precipitation orientation was reported as $[001]_{\text{NbC}} \parallel [001]_{\text{g}}$ ^[22]. The alloy aging at 850 °C (HT2-1 alloy) exhibits some gray particles precipitate adjacent to the primary NbC (Fig. 2b). The precipitates show rectangle morphologies, and commonly possess the size within 100 nm in length. Because the particle is smaller than the EDX analyzing spot size, the results will be affected by the signal from the matrix. However, it can be deduced that the precipitate phase is Cr_{23}C_6 according to the XRD analyzing results. For further confirms the precipitate phase, TEM analyzing has been conducted in this work, as shown in Fig. 4a. Small precipitates carbide can be observed within the matrix, with the precipitation orientation of $[011]_{\text{Cr}_{23}\text{C}_6} \parallel [011]_{\text{r}}$. This is coincidence to others' work, which has reported the same orientation of the Cr_{23}C_6 during aging at 760°C^[20] and 900°C^[5, 16, 24], respectively. It can also be inferred that the precipitation orientation of carbide is not affected by the aging temperature. However, the precipitation of Cr_{23}C_6 shows fewer fractions when aging at 750 °C (HT3-1 alloy). On the other hand, quite fewer amount of secondary NbC can be observed in both HT2-1 and HT3-1 alloys, further confirming that secondary NbC primarily precipitates during aging at 1120 °C. The precipitation orientation of NbC has also been confirmed as $[001]_{\text{NbC}} \parallel [001]_{\text{r}}$, as shown in Fig. 4b. All of these results show strong dependence of the carbide precipitation behavior on the aging temperature of 20Cr32Ni1Nb alloy. Moreover, no G phase has been detected in the specimen aging at various temperatures, probably because of the low Si content in the alloy. This is quite different from other's work, which shows G phase precipitates within several hours, especially in the temperature range between 700-850 °C^[26].

Tensile test at 750 °C has been conducted for investigating the mechanical properties of the materials. Compared with the commercial materials (CT15C), the higher strength and slightly lower ductility of 20Cr32Ni1Nb alloy probably associate with the addition of tungsten and molybdenum. On the other hand, the results show that the aging temperature has great effect on both the strength and the ductility (Fig. 5). Short aging at 1120 °C doesn't seem to show obvious effect on the alloy strength, while the yield strength and area reduction increase significantly after aging at lower temperatures. The HT2-1 alloy (aged at 850 °C) gains the highest mechanical properties at 750 °C, including both the yield strength and the area reduction. In addition, the decrease of the ductility, usually induced by the G phase formation during aging at 700-850 °C^[27], is not observed in this work.

The comparison on the rupture surface has been made among the samples, and the dendrite structure can be observed in all the test specimens (Fig. 6). This indicates that the sliding at the dendrite interface take place during deformation, and probably plays an important role in the rupture of the material. This is in good agreement with Luis's work, who observed that the dendrite interface is the main crack path in 20Cr32Ni1Nb alloy^[28]. However, the specimens with various aging methods show some different characteristics at the fracture surface. For the HT2-1 and HT3-1 alloys, heterogeneous deformation occurs and the largest deformation direction is approximately 30-45° to the dendrite growth direction. Moreover, the dimple can be easily identified by the higher magnification observation of the secondary crack, which

indicates that a larger sliding at dendrite interface is allowed before the fracture of the HT2-1 and HT3-1 alloy. This is quite different from the cast and HT1-1 alloys, which did not experience large deformation before rupture took place.

These phenomena have been confirmed by the cross-section of the fracture specimen, as shown in Fig. 7. The void can be found at the interface between the matrix and NbC, as well as some microcrack in the large NbC particle. This demonstrates that the interface between primary NbC and matrix is the weak point which acts as the cracks nucleation site, and thus facilitates the dendrite sliding during deformation. Similar phenomenon has been observed by Kargarnejad, who found that the Nb-rich particle in the matrix is the preferred sites to generate the voids^[29]. However, by compare to the cast and HT1-1 alloy, the subsurface of the HT2-1 and HT3-1 alloy experiences the significant deformation before the cracks propagation. This confirms a larger sliding at dendrite interface took place before crack propagation. According to the classical fracture theory, the threshold of crack propagation can be described as below^[30, 31].

$$G_I > G_{IC} \quad (2)$$

Here, G_I is the driving force energy of the crack propagation, which depends on the applied stress condition; G_{IC} is the critical energy required for the crack propagation, which is associated with the inherent properties of the material. G_I and the G_{IC} can be rough estimated as below:

$$G_I = s^2 \rho c / E \quad (3)$$

$$G_{IC} = 2(g + \gamma) \quad (4)$$

Here, s could be considered as the stress condition nearby the crack, E is the modulus of elasticity, c is the size of the micro-void, g is the surface energy of the material, γ is the energy required for the crack to penetrate the strain field at the crack tip.

Thus, it can be inferred that the crack propagation primarily depends on the stress level and the strain field condition at the dendrite interface. On the other hand, extensive rod like void has been observed adjacent to the primary NbC in the HT2-1 alloy, and all of them grow toward the deformed direction. This indicates that the secondary $Cr_{23}C_6$ adjacent to the primary NbC has pinning the matrix during alloy deformation, and thus relieves the stress concentration at the NbC/matrix interface. Similar phenomenon has been observed in the HT3-1 alloy, but the amount of the rod like void is in the less extent. In addition, $Cr_{23}C_6$ precipitates adjacent to primary NbC may have an effect on the strain field generation at the dendrite interface, which further inhibits the crack propagation during deformation. Although the small precipitate has also been detected in HT1-1 alloy, its positive effect can be eliminated for the formation of the precipitation free zone (PFZ) between the primary and secondary NbC.

3.2 20Cr32Ni1Nb with high C content

Fig. 8 shows the microstructure analyzing results on the 20Cr32Ni1Nb alloy with higher carbon content. Compared with the 20Cr32Ni1Nb alloy, the carbon content increase leads to the directly

formation of the primary Cr_{23}C_6 during solidification. This is primarily induced by the low Nb/(C+6/7N) ratio, which facilitates the chromium carbide formation when the value is below 7.7^[32].

Increasing C content in 20Cr32Ni1Nb alloy will lead to the different precipitation behaviors, and it is closely related to the aging temperature (Fig. 9). The material aging at 1120 °C (HT1-2 alloy) shows the carbide precipitates at grain boundary and dendrite interface, whose size can sometimes reach up to 2 μm in equivalent diameter (Fig. 9a). This is because the excess carbon was ejected and segregated at dendrite interface during the solidification^[19, 33], leading to the supersaturated solution of carbon. As a comparison, large amounts of secondary carbide can be observed in the materials aged at 850 °C (HT2-2 alloy) and 750 °C (HT3-2 alloy), mainly precipitates at dendrite interface with the relatively smaller size (Fig. 9b, c). This carbide has been confirmed as the Cr_{23}C_6 by EDX elemental analysis, as shown in Fig. 10. It is quite different from the precipitation behavior of 20Cr32Ni1Nb alloy, which exhibits the nose temperature of Cr_{23}C_6 precipitation at approximately 850 °C. As the carbon content increases, larger amount of Cr_{23}C_6 can be observed after aging at 750 °C, indicating the nose temperature decreased accompanied with the composition modification. This phenomenon can be explained from the aspect of thermodynamics, and the nucleation tendency of the carbide can be decided as following:

$$\Delta G = \frac{4}{3}\pi r^3(\Delta G_u + \Delta G_e) + 4\pi r^2 g_{ab} < 0 \quad (5)$$

Here, ΔG is the Gibbs energy change for the carbide precipitation; ΔG_u and ΔG_e are unit chemical energy and strain energy change during the precipitation, respectively; r is the radius of the carbide particle; g_{ab} is the interface energy between the carbide and matrix. Thus, the carbide will precipitate when the $(\Delta G_u + \Delta G_e) < 0$. In Eq. 5, all the ΔG_u , ΔG_e and g_{ab} are determined by the aging temperature. Thus, it could be inferred that, as for the 20Cr32Ni1Nb alloy, the carbide precipitation reached the maximum at about 850 °C, as well as the $\dot{\epsilon}_{\text{Cr}} \approx 0$. However, the driving force of the precipitation, ΔG_u , is strongly affected by the super cooling degree, which can be calculated as:

$$\Delta G_u = (T_c - T) \Delta S = \Delta T \Delta S \quad (6)$$

Here, T_c is the starting temperature of the precipitation; T is the aging temperature; ΔS is the entropy change during precipitation. Thus, the ΔG_u greatly increased in the high C-adopted 20Cr32Ni1Nb alloy, mainly because the starting precipitation temperature (T_c) significantly increased. Although the high carbon content may also increases the ΔG_e and g_{ab} , but it is relatively small compared with the increase of ΔG_u at the very beginning of the precipitation. Thus, the temperature should be lower for the higher ΔG_e and g_{ab} in the high C-adopted 20Cr32Ni1Nb system, and thus the nose temperature should decrease. As a comparison, when the alloy aging at higher temperature, the ΔG_u is too small that the chromium carbide cannot precipitate from the matrix directly. Thus, the carbide precipitates at the interface between NbC and matrix, which needs much lower ΔG_e and g_{ab} during the carbide nucleation. At the same time, the rapidly diffusion of the atom at elevated temperature permits the fast coalesce of the chromium carbide, and thus results in the formation of large Cr_{23}C_6 adjacent to the primary NbC in HT1-1 alloy. In addition, secondary NbC particles have been observed nearby the dendrite interface, and keep their size in various aging temperatures.

The tensile test result is shown in Fig. 11. It has been confirmed again that the aging temperature has great effect on the yield strength and alloy ductility, and the maximum ductility was achieved after aging at 850 °C. Compared with 20Cr32Ni1Nb alloy, the yield strength of high C-adopted alloy is relatively lower after aging at 850 °C, probably because of its **relatively** lower tungsten content^[34]. However, the strength of high C-adopted alloy shows kind of different tendency, and the highest value was obtained after 750 °C aging. This is probably **because of** the excess precipitation of carbide, which is accompanied with the decrease of alloy ductility.

The observation of the fracture surface confirmed the difference rupture mechanism of the materials (Fig. 12). All high C content 20Cr32Ni1Nb alloys show the dendrite structure characteristics on the fracture surface, indicating the weakness at the dendrite interface. However, the HT2-2 and HT3-2 alloys gain higher fraction of transgranular fracture, as well as the relatively larger deformation along the dendrite growth direction. Moreover, dimple can easily be found at the dendrite interface of the alloy aged at lower **temperatures**.

The cross-section of the fracture subsurface has been observed by BSE, and the results show **great effect** of the carbide size on the crack propagation behavior (Fig. 13). It can be seen that the void is prone to nucleate at the interface between large Cr_{23}C_6 particle and primary NbC, and usually develops into **relatively** large size. The micro-cracks also have been observed within the large Cr_{23}C_6 particle in HT1-1 alloy, showing a convenient path for the cracks propagation. According to Eq. 4, the threshold of the crack prolongation depends on both the σ and the σ_c . Due to the poor ductility of both Cr_{23}C_6 and NbC, the size of the strain field (σ_c) should be very small, **which** demonstrates that σ could be **neglected** in this case, and thus the crack propagation resistance becomes smaller. This probably responds **to** the low properties of HT1-2 alloy at the test temperature. The void between two kinds of carbide was **also found** in HT2-2 and HT3-2 alloys, but such kind of void rarely forms beside small Cr_{23}C_6 particle near the primary NbC.

4. Conclusion

1. The aging temperature **greatly affects** the precipitation behavior of the secondary carbide. NbC will precipitate near the dendrite interface after the 2 hours of aging at 1120 °C, as well as the PFZ developed between the primary and the secondary NbC. On the other hand, small Cr_{23}C_6 achieved the greatest precipitation tendency at 850 °C, adjacent to the primary NbC carbide.
2. The distribution of the secondary carbide affects the mechanical properties of 20Cr32Ni1Nb alloy at elevated **temperatures**. Small secondary carbide precipitates at dendrite interface **significantly improve** the yield strength and the ductility of the material, mainly because the crack propagation at the interface between primary NbC and matrix was **effectively** inhibited. However, this positive effect will be eliminated when the PFZ developed between the primary and the secondary carbide.
3. Increasing the carbon content in the 20Cr32Ni1Nb **affects** the microstructure and the mechanical properties **significantly**. The primary Cr_{23}C_6 **forms** at the dendrite interface of the cast high C-

20Cr32Ni1Nb alloy. Aging at 1120 °C facilitates the precipitation of large Cr_{23}C_6 adjacent to the NbC, which acts as the crack propagation path during alloy deformation. Aging at lower temperature facilitates the disperse distribution of small secondary Cr_{23}C_6 at dendrite interface. However, once the Cr_{23}C_6 growth into large size, it plays an important role in the crack nucleation during deformation, and thus brings negative effect on the ductility of the materials.

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Reference

- [1] A. Energy, in, API, 2011, pp. 942-A.
- [2] D.M. Knowles, C.W. Thomas, D.J. Keen, Q.Z. Chen, International Journal of Pressure Vessels and Piping, 81 (2004) 499-506.
- [3] J. Wadsworth, J.H. Woodhead, S.R. Keown, Metal Science, 10 (1976) 342-348.
- [4] H.J.B. Jun, F.L. Versnyder, Nature, 178 (1956) 208-209.
- [5] G.D. Barbabala, L.H. de Almeida, T.L. da Silveira, I. Le May, Materials Characterization, 26 (1991) 1-7.
- [6] G.D. Barbabala, L.H. Dealmeida, T.L. Dasilveira, I. Lemay, Materials Characterization, 26 (1991) 193-197.
- [7] G.D.D. Soares, L.H. Dealmeida, T.L. Dasilveira, I. Lemay, Materials Characterization, 29 (1992) 387-396.
- [8] D.J. Powell, R. Pilkington, D.A. Miller, Acta Metallurgica, 36 (1988) 713-724.
- [9] J. Yan, Y. Gu, J. Lu, Materials Science and Technology, (2014).
- [10] J. Hoffman, in: NACE International, Corrosion 2000, 2000, pp. no. 512.
- [11] R. Colwell, J. Hoffman, in: NACE International, 1998, pp. no. 423.
- [12] L.S. Monobe, C.G. Schön, International Journal of Pressure Vessels and Piping, 86 (2009) 207-210.
- [13] J.C. Guimarães Teixeira, G.V. Peixoto Donato, L.N. Júnior, C.B. Eckstein, in: World Petroleum Congress Proceedings, 2006.
- [14] S. Shi, J.C. Lippold, J. Ramirez, Welding Journal (Miami, Fla), 89 (2010) 210S-217S.
- [15] S. Zhao, X. Xie, G.D. Smith, S.J. Patel, Materials & Design, 27 (2006) 1120-1127.
- [16] R.A. Pedro Ibañez, G.D. de Almeida Soares, L.H. de Almeida, I. Le May, Materials Characterization, 30 (1993) 243-249.
- [17] S. Shi, J.C. Lippold, Materials Characterization, 59 (2008) 1029-1040.
- [18] I.A. Sustaita-Torres, S. Haro-Rodríguez, M.P. Guerrero-Mata, M. de la Garza, E. Valdés, F. Deschaux-Beaume, R. Colás, Materials Chemistry and Physics, 133 (2012) 1018-1023.

- [19] X.Q. Wu, H.M. Jing, Y.G. Zheng, Z.M. Yao, W. Ke, Z.Q. Hu, *Materials Science and Engineering: A*, 293 (2000) 252-260.
- [20] Q.Z. Chen, C.W. Thomas, D.M. Knowles, *Materials Science and Engineering: A*, 374 (2004) 398-408.
- [21] M.P. Dewar, A.P. Gerlich, *Metall. Mater. Trans. A-Phys. Metall. Mater. Sci.*, 44A (2013) 627-639.
- [22] M. Dewar, in: *Department of Chemical and Materials Engineering, University of Alberta, Edmonton, Alberta*, 2012.
- [23] B. Piekarski, *Materials Characterization*, 47 (2001) 181-186.
- [24] L.H. de Almeida, A.F. Ribeiro, I. Le May, *Materials Characterization*, 49 (2003) 219-229.
- [25] J. Yan, Y. Gao, F. Yang, C. Yao, Z. Ye, D. Yi, S. Ma, *Materials Science and Engineering: A*, 529 (2011) 361-369.
- [26] R.C. Ecob, R.C. Lobb, V.L. Kohler, *J. Mater. Sci.*, 22 (1987) 2867-2880.
- [27] J. Hoffman, J. Magnan, in: *NACE International, Corrosion 2003*, 2003, pp. no. 3469.
- [28] L.S. Monobe, C.G. Schön, *Journal of Materials Research and Technology*, 2 (2013) 195-201.
- [29] S. Kargarnejad, V. Abbasi-Chianeha, *Case Studies in Engineering Failure Analysis*.
- [30] G.R. Irwin, in: *Fracturing of Metals, Proceedings of the ASM symposium on Fracturing of Metals*, Cleveland, 1948, pp. 147-166.
- [31] E. Orowan, in: *Reports on Progress in Physics*, 1949, pp. 185-232.
- [32] S.R. Keown, F.B. Pickering, in: *Creep Strength in Steel and High Temperature Alloys*, Metals Society London, London, 1974, pp. 134-143.
- [33] F.C. Nunes, J. Dille, J.L. Delplancke, L.H. de Almeida, *Scripta Materialia*, 54 (2006) 1553-1556.
- [34] D.J. Crudden, A. Mottura, N. Warnken, B. Raeisinia, R.C. Reed, *Acta Materialia*, 75 (2014) 356-370.

Table. 1 Chemical composition of the materials

Alloy	Weight percentage of the element /wt.%								
	Fe	Cr	Ni	C	W	Mo	Nb	Si	Mn
1. 20Cr32Ni1Nb	Bal.	22.51	33.28	0.10	2.86	0.95	0.77	0.14	0.99
2. High C-20Cr32Ni1Nb	Bal.	23.94	36.61	0.19	1.83	0.95	0.79	0.16	0.96
20Cr32Ni1Nb Alloy*	Bal.	19-22	31-35	0.07-0.16			0.6-1.6	1.2≤	1.5≤

*: This composition refers to the ASTM: A 351 (Grade CT 15C)

Table. 2 Aging parameters of alloy

Aging parameters	
HT 1	1200 °C /2.0 h (AC)+1120 °C /2.0 h (AC)
HT 2	1200 °C /2.0 h (AC)+850 °C /18.0 h (AC)
HT 3	1200 °C /2.0 h (AC)+750 °C /30.0 h (AC)

Table Captions

Table. 1 Chemical composition of the materials

Table. 2 Aging parameters of alloy

Figure.1a

Figure.1b

Figure.2a

Figure.2b

Figure.2c

Figure.2d

Figure. 3

Figure. 4a

Figure. 4b

Figure. 5a

Figure. 5b

Figure. 6a

Figure. 6b

Figure. 6c

Figure. 6d

Figure. 7a

Figure. 7b

Figure. 7c

Figure. 7d

Figure. 8a

Figure. 8b

Figure. 9a

Figure. 9b

Figure. 9c

Figure. 9d

Figure. 10

Figure. 11a

Figure. 11b

Figure. 12a

Figure. 12b

Figure. 12c

Figure. 12d

Figure. 13a

Figure. 13b

Figure. 13c

Figure. 13d

Figure Captions

Fig. 1 Microstructure of as-cast 20Cr32Ni1Nb alloy: (a) Microstructure, (b) Phase constitution

Fig. 2 Microstructure evolution of 20Cr32Ni1Nb alloy during different aging temperature: (a) aged at 1120 °C, (b) aged at 850 °C, (c) aged at 750 °C, (d) phase constitution after aging

Fig. 3 EDX map analyzing result of 20Cr32Ni1Nb alloy after aging at 1120°C for 2 hours

Fig. 4 Morphology and its precipitation orientation of the secondary (a) Cr_{23}C_6 and (b) NbC in the HT2-1 and HT1-1 alloy, respectively.

Fig. 5 The 750 °C tensile test results of the 20Cr32Ni1Nb alloy: (a) Strength, (b) Ductility (The data of CT 15C alloy refer to the ASTM: A351 Grade CT15C)

Fig. 6 Morphology of the alloy fracture surface: (a) cast alloy, (b) 1120 °C aged alloy, (c) 850 °C aged alloy, (d) 750 °C aged alloy

Fig. 7 Cross-section microstructure of the fracture subsurface: (a) cast alloy, (b) 1120 °C aged alloy, (c) 850 °C aged alloy, (d) 750 °C aged alloy

Fig. 8 Microstructure of as-cast high C-adopted 20Cr32Ni1Nb alloy: (a) Microstructure, (b) Phase constitution

Fig. 9 Microstructure evolution of high C-adopted 20Cr32Ni1Nb alloy during different aging temperature: (a) aged at 1120 °C, (b) aged at 850 °C, (c) aged at 750 °C, (d) phase constitution after aging

Fig. 10 EDX map analyzing result of high C 20Cr32Ni1Nb alloy after aging at 850°C for 18 hours

Fig. 11 The 750 °C tensile test results of the high C-adopted 20Cr32Ni1Nb alloy: (a) Strength, (b) Ductility

Fig. 12 Morphology of the alloy fracture surface: (a) cast alloy, (b) 1120 °C aged alloy, (c) 850 °C aged alloy, (d) 750 °C aged alloy

Fig. 13 Cross-section microstructure of the fracture subsurface: (a) cast alloy, (b) 1120 °C aged alloy, (c) 850 °C aged alloy, (d) 750 °C aged alloy