AN INFLUENCE OF MICROSTRUCTURE ON THE MECHANICAL PROPERTIES OF THE CORROSION RESISTANT SUPERALLOY CHS88U

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Abstract

Nickel based corrosion resistant superalloy ChS88U, containing 16 mass % Cr, is widely used as a material of gas turbine components. At present, the ChS88U equiaxed blades have more than 80 000h operating time in industrial gas turbine. An influence of microstructure on the mechanical properties of the ChS88U alloy is investigated, using the samples after cast, heat treatment (HT) and hot isostatic pressing (HIP) + HT. Typical creep-rupture lives of cast + HT and HIP + HT samples at 900°C/274MPa are 271h and 316h, respectively. Mechanical properties of ChS88U are also compared with those of the ChS70 and In738 alloys. The microstructure of the blade with ChS88U post 30 000h operating time in industrial gas turbine investigated by SEM is given in the paper.

Introduction

Nickel based superalloy ChS88U was developed specifically for the marine and industrial gas turbine applications requiring superior hot corrosion resistance and also a low cost. It is widely used as a material of gas turbine components, including blades and vanes, manufactured by conventional investment casting technology. At present, the ChS88U equiaxed blades have more than 80 000h operating time in industrial gas turbine.

The ChS88U alloy has a good corrosion resistance, first of all, due to 16 mass % chromium (Cr) concentration. It is accepted that the temperature capability decreases with increasing of Cr concentration in superalloys [1], therefore the superalloys for aero-engines – single crystals of third generation contain 2...4% (wt.) Cr [2, 3]. The target temperature capability for the superalloys is set at 1100°C for 1000h creep rupture life. Operating conditions of industrial gas turbines are different from those of aero-engines and require the strength and stability for time 100.000h at temperature 900...950°C with hot corrosion resistance. To provide the high corrosion resistance, superalloys for land-based turbines contain more than 10% Cr and the ratio Al/Ti $\geq$ 0,7...1,0 [3]. To provide the long operating time the alloys must have very good phase and structure stability.

The objective of the paper is to describe the results of the study of microstructure, phase stability and mechanical behaviors in view of the use of the alloy in gas turbine of the 100.000 hours operating time.

Experimental Procedure

Material

The corrosion resistant superalloy ChS88U contains 16 mass % Cr, the compositional ratio of Al/Ti=0,7. The total concentration of the refractory elements: tungsten (W) and molybdenum (Mo) is 7,2 wt.%. The alloy contains trace elements: carbon (C), boron (B), zirconium (Zr), yttrium (Y) and Hf as the grain and phase boundary strengthening elements. A feature of the composition is of a high concentration of boron - 0,09 mass %, whereas a number of new superalloys have been developed based on the principle of trace element control [4, 5]. The chemical composition of the superalloy ChS88U is given in Table I with the superalloy ChS70 containing no boron as reference.

The ChS88U alloy composition is analyzed with the aid of the d-electrons concept, so that any undesirable phases do not appear in the alloy [6, 7]. It is known that high strength nickel based superalloys have special value $\overline{M}_d$: 0,975...0,995. The $\overline{M}_d = 0,98$ of the alloy ChS88U is set on the upper bound of this range by the given composition of the refractory elements in the alloy.

The alloy was melted by Firth Rixson Superalloys Ltd. (Great Britain) and cast to the equiaxed bars of 15 mm diameter and 70 mm in length.

DTA

The temperatures of phase equilibrium are obtained by differential thermal analysis (DTA) in conditions of the heating and cooling rate 20 degree/minute, using cylindrical as cast samples of 7 mm diameter and 7 mm in length and nickel reference pattern.

Heat Treatment (HT)

HTs for investigation of the influence of microstructure on mechanical characteristics were performed under air atmosphere in a laboratory furnace. The conditions of solution treatment and precipitation heat treatment of γ' -phase are shown in Table II.

Table I: Chemical composition in wt% of ChS88U and ChS70 alloys (balance: Ni)											
Alloy	Co	Cr	W	Mo	Nb	Al	Ti	C	Hf	B	Zr
ChS88U	11,7	15,8	5,3	1,96	0,15	3,0	4,6	0,07	0,3	0,09	0,03
ChS70	11,0	15,9	5,5	2,0	0,2	2,8	4,5	0,09	-	-	0,05

Figure 1. Typical DTA curves for ChS88U (A) and ChS70 (B)

Heat Treatment (HT)

HTs for investigation of the influence of microstructure on mechanical characteristics were performed under air atmosphere in a laboratory furnace. The conditions of solution treatment and precipitation heat treatment of γ' -phase are shown in Table II.

Table II: Standard heat treatment of ChS88U and ChS70 alloy

Conditions	T, °C	time, h
solution	1170	4
aging 1	1050	4
aging 2	850	16

Microstructure

The microstructure of the alloys ChS88U and ChS70 are observed, using scanning electron microscopy (SEM) JSM-5600 equipped with EDX analyzer, after cast, solution treatment, solution and one-step aging treatment, solution and two-step aging treatment, hot isostatic pressing (HIP) and two-step aging treatment in creep ruptured equiaxed specimens and blades after 30 000 hrs of operating time in industrial gas turbine.

Density

To acquire data about microporosity, the density of the ChS88U samples is measured by the hydrostatic weighing standard procedure, using high-accuracy weighing machine with the accuracy 0,0001 g.

The densities of the samples after cast, solution treatment and HIP are measured, using the same sample after and before solution and after and before HIP.

Mechanical Properties

As cast, HT and HIP + HT samples are machined mechanically for the creep rupture and the tensile tests. The gauge length of the specimen is 5 mm in diameter and 35 mm in length. The creep rupture tests carried out under temperature/stress conditions of 900°C/274MPa; 800°C/382MPa. The tensile tests were conducted under temperature conditions of 20; 725; 800; 900°C.

Experimental Results and Analysis

The DTA curves for the alloys ChS88U and ChS70 show (Figure 1) liquidus, solidus, carbide, carboboride formation and γ' -solution temperature. The most significant difference in the shape of the DTA curve for ChS88U is the endothermic effect at 1190°C that is thought to be associated with the melt of carboboride eutectic. The values of the phase transformation temperatures are given in Table III.

Table III: DTA data of ChS88U and ChS70 alloys (°C)

Alloy	T _L	T _S	T _{carbobor eut}	T _{BP} ^{BP} _{solv}
ChS88U	1363	1250	1193	1162
ChS70	1357	1251	—	1166

Figures 2–5 show the microstructures of the ChS88U. Carbides are observed mostly in the interdendritic regions and also at grain boundaries. The microstructure of the ChS88U alloy includes two types of carbide phases. The main carbide is a script form (MC-I); the bright contrast suggests that the phase contains elements having higher atomic numbers. The second type of carbide (MC-II) is discrete shaped carbide of more dark contrast. The chemical analyses of the phases are given in Table IV. The MC-I phase is (Cr, Mo, W)-enriched. Their morphology «script» (as cast sample) is evidence of the eutectic crystallization (Figure 2). The second type of carbides MC-II are (Ti, W, Hf)-enriched.

The segregation areas contain small discrete borides and Hf enrichment.

It is estimated by the alloy design program that the amount of γ' eutectic or primary is about 3,5%. It is shown by SEM that the size of γ' eutectic is rather small 1,0...5,0 μ m. The chemical analyses of γ/γ' and γ B_{eu}-phases are given in Table IV.

During solution treatment at 1170°C the γ' phase is not totally dissolved (Figure 3). Almost all the γ' precipitates are dissolved except for the γ' particles in interdendritic regions and also at the grain boundaries (eutectic areas). The main γ' phase is due to the solution and re-precipitation of. The average size of the γ' particles is $\sim$ 0,1 μ m.

After solution and one-step aging treatment the average size of γ' particles is about 0,3 μ m due to the re-precipitation of at 1050 °C. As a result of two-step aging treatment at 1050°C for 4 hrs and at 850°C for 16 hrs the γ' size $\sim$ 0,5 μ m and cubic shape are optimal for creep resistance (Figure 4).

The main (Cr, Mo, W)-enriched carbides have discrete shaped form due partial dissolving at 1170°C.

For alloys by a large solidification interval, the solidification of the interdendritic eutectic is accompanied by shrinkage. In these closed regions some microporosity appears as a result of bad feeding. One more cause of the microporosity appearance in the ChS88U alloy is a small amount of the incipient melting of carboboride eutectic at the temperature of solution treatment. HIP is an opportunity to get less microporosity and as a result to increase mechanical behaviors especially fatigue properties [10].

The density data of two ChS88U samples: №1 as cast and after solution treatment, №2 as cast and after HIP are given in Table V.

The density of the sample after solution has decreased by 0,03%. The cause of the effect can be the pores aroused after

solution of phases, mainly of γ' , over the difference of γ and γ' molar volumes [9]. After HIP the density of the sample has increased by 0,05% in comparison with the as cast sample. It has to point, that the initial as cast samples chosen for the investigation have the sufficiently density. Therefore, obtained effect of HIP is not very sizeable.

Figure 5 demonstrates the microstructure in the sample after HIP and two-step aging treatment. It differs from the microstructure in sample after standard HT of lesser liquating heterogeneity accompanied by partial MC-I carbides solution and their fragmentation.

The measured mechanical properties of the ChS88U samples are listed in Tables VI and VII. The samples after HIP + HT have higher stress rupture life and slightly higher tensile properties than the samples after standard heat treatment.

Figure 6 shows the typical tensile stress-temperature curves with the stress maximum in the temperature range 600 to 800°C for the standard heat treated ChS88U samples. The creep rupture data at 700, 800, 850 and 900°C for ChS88U, ChS70, IN738 are plotted in Figure 7. As shown, there is a 40 and 60 MPa loss in creep strength for the ChS70 and In738 accordingly at 900°C, 850°C and 800°C.

The creep-rupture results of ChS88U including long-term data are compared with the ChS70 and In738 data with Larson-Miller Parameter in Figure 8.

SEM microstructures of the blade post 30 000h operating time in the industrial gas turbine M-70 of the rated capacity 8Mwt/h are shown in Figures 9 and 10. Some precipitations of acicular looking TCP phases (Cr & W rich) are apparent, though not of sufficient volume fraction to de-alloy the material or nucleate creep cracking (Figure 9).

Figure 2. As cast ChS88U samples: dendritic structure (A); interdendritic region (B); two types of carbide phases (C)

A

B

Figure 3. Standard heat treated ChS88U sample: grain boundary (A); interdendritic regions (B)

Figure 4. The strengthening γ' - phase in heat treated ChS88U sample

Figure 5. Microstructure in ChS88U sample after HIP + HT

Figure 5. Microstructure in ChS88U sample after HIP + HT

Table IV: Content of M in MC, γ/γ' and γ'_{eu} - phases (ChS88U alloy)
(mass fraction in %)

phase	W	Mo	Cr	Ti	Al	Hf	Ni	Co
MC-I	37,62	22,06	31,47	3,31	-	-	6,4	3,04
MC-II	32,11	5,11	1,85	39,91	-	2,35	3,27	-
γ/γ'	4,18	1,84	15,54	4,27	2,23	-	63,03	11,25
γ/γ' *)	6,3	1,5	16,1	3,0	2,1	-	59,4	11,3
γ/γ' **)	3,8	1,4	14,8	6,0	2,7	-	60,0	11,0
γ'_{eu}	0,5	1,6	5,8	11,2	7,4	-	68,2	6,8

*) dendrite area; **) interdendritic area

Table V: Density of ChS88U samples as cast, after standard HT and after HIP + HT

Sample		Density, g/cm ³	Δ , %
№1	as cast	8,2213	-0,03
	solution	8,2192	
№2	as cast	8,2272	+0,05
	HIP	8,2309	

Table VI: Tensile properties of ChS88U samples (mean values)

Sample	Temperature (°C)	0,2%YS (MPa)	UTS (MPa)	EL (%)	RA, (%)
as cast	20	919	1193	10,5	12,5
standard HT	20	897	1113	5,7	9,7
standard HT	600	897	1113	6,0	10,0
standard HT	725	837	1147	10,3	10,45
standard HT	800	912	985	13,4	26,18
standard HT	900	657	717	24,3	37,7
HIP + HT	20	924	1129	7,3	14,7

Table VII: Stress rupture life of ChS88U samples

Alloy	Sample	Temperature (°C)	Stress (MPa)	Life, (hrs)
ChS88U	as-cast	900	274	152
	standard HT	900	274	271
	standard HT	800	382	1000
	HIP + HT	900	274	316
ChS70	standard HT	900	245	200

Figure 6. Tensile properties-temperature curves of ChS88U

Figure 7. Creep rupture lives of different superalloys

Figure 8. Larson-Miller Plot for comparison between ChS88U, ChS70 and In738 alloys

Airfoil

Root

Figure 9. Microstructure in airfoil and root of blade having 30 000h operating time in industrial gas turbine

Figure 10. Strengthening γ' -phase in the airfoil of blade having 30 000h operating time in industrial gas turbine

Conclusions

An influence of microstructure including density of the samples on the mechanical properties of the corrosion resistant ChS88U alloy is investigated. The following results have been obtained:

1. The ChS88U alloy has an advantage over the ChS70 alloy due to boron and hafnium alloying and over IN738.
2. The higher stress rupture life and slightly higher tensile properties are obtained after HIP, due to the increase of density of the alloy.
3. The microstructure investigation of the blade post 30 000h operating time in industrial gas turbine shows a good phase and structure stability of the alloy.

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