MICROSTRUCTURE AND PROPERTIES ANALYSIS OF THE ESR-CAST-TO-SHAPE GAS TURBINE DISCS

J.F. Chang

Daye Steel Works

G.P. Hu

Shanghai Steel Works No.5

L. Gao

The Central Iron and Steel Research
Institute, Beijing, China

ABSTRACT

The ESR-cast-to-shape process can be used for the manufacture of gas turbine discs. The structure and properties of large ESR-cast-to-shape discs made of Fe-27Ni-15Cr-1.6 Mo-2.8Ti were studied in this work. These discs have successfully passed 800 h engine test on certain aircraft gas turbine. The results of experiments showed that the macrostructure of the disk is radially unsymmetrical, but that the properties are not appreciably dependent on the orientation of the columnar grains except for the high stress low cycle fatigue properties. However, the LCF properties still meet the service demands, although the value along the disc thickness is rather low. Therefore, these discs are suitable for use in aircraft engines and also have significance for large ground gas turbines.

INTRODUCTION

Up to now gas turbine discs are usually manufactured by the die forging processes. These processes are rather expensive for large discs, with diameters over say 800 mm. Besides, they require a long production cycle and powerful die forging presses.

In recent years the ESR-cast-to-shape process has been developed rapidly in China. It is used to produce large turbine discs for aircraft engine. Results of two pilot runs of the discs for 800 hr and other experiments showed that the structure and the properties of the ESR-cast-to shape turbine discs are satisfactory for service in aircraft engines.

Because of the simple technology, the ESR-cast-to-shape process gives a metal saving of about 20% and decreases considerably the manufacturing cost. Therefore, ESR cast discs are promising for use in large ground and Naval gas turbines.

The present paper describes the influence of the structure of ESR-cast turbine discs on their properties, and the viability of their commercial adoption.

EXPERIMENTAL

The alloy selected was Fe-27Ni-15Cr-1.6Mo-2.8Ti, the diameter of the ESR-cast-to-shape discs was 850 mm. Studies were made on the macrostructure and microstructure of the ESR cast disc, and their influence on the mechanical properties at room and high temperature.

Specimens were taken directly from the discs. The etchant for macrostructure was $\text{CuSO}_4 \cdot 5\text{H}_2\text{O}$ 150g., H_2SO_4 35 ml., HCl 500 ml. and that for microstructure a 10% H_2SO_4 , 3.5% NaCl , 3% FeSO_4 water solution (electrolytic etching). The various phases present in the alloy were identified and their quantities determined by means of electrochemical separation, chemical analysis, X-ray diffraction, electroprobe, electron diffraction, etc.

Normal mechanical properties were examined by conventional methods. Fracture toughness (K_{IC}) at room temperature was determined with compact tensile specimens $B = 24$ mm., according to ASTM 399-72. The high stress LCF properties at 650°C were determined with specimens of $\varnothing 5 \times 56$ mm size, on axial tensile testing (frequency 0.583 Hz, the stress ratio 0.1) in terms of the number of cycles (N_f) up to fracture. The hot fatigue properties $650^\circ\text{C} \rightleftharpoons 20^\circ\text{C}$

were examined by measuring the crack length on notched plate specimens 2 x 35 mm after 150 cycles according to HCS 106-62.

RESULTS OF EXPERIMENTS

1. Macrostructure of ESR-cast-to-shape Discs and Its Influence on the Properties

These discs had a radially unsymmetrical macrostructure, as shown in Fig. 1. On longitudinal sections there are two rows of coarse columnar grains existing in a herringbone pattern with an included angle of about 90° . These columnar crystals are 40-60 nm long and have grown along the direction $[001]$.

Because of the radical unsymmetry in the macrostructure it was thought necessary to examine the mechanical properties in various directions of the disc. These were examined in 3 directions, i.e. parallel to the direction of columnar grain growth, perpendicular to the columnar grain growth, and the direction along the disc thickness. The mechanical properties at room temperature and at 650°C , the stress rupture properties at 650°C , and the fracture toughness at room temperature were found to be practically the same, as shown in table 1. Only the high stress LCF properties were strongly

affected by the sampling direction. It can be seen from Fig. 2 that the LCF values are highest along the direction of columnar grain growth, lowest along the disc thickness, and medium in the direction perpendicular to the direction of columnar grain

Fig. 1 General view and longitudinal macrostructure of disc

Table 1. Properties of specimens determined in different directions
(Specimens: 1050°C, 4 hr, O.C., 700°C, 32 hr, A.C.)

Direction of sampling	R.T. tensile		R.T. impact K_Q	650°C tensile		Stress-rupture properties	
	UTS MN/m ²	YS MN/m ²		UTS MN/m ²	YS MN/m ²	650°C MN/m ²	490.3 MN/m ²
columnar grains	1016.0	843.4	16.4	22.7	5.9	717.8	583.5
└ columnar grains	1051.3	885.5	13.6	20.8	5.3	758.1	578.6
along the disc thickness	1011.1	872.8	14.0	19.4	4.6	661.9	526.6
columnar grains	1016.0	843.4	16.4	22.7	5.9	717.8	583.5
└ columnar grains	1051.3	885.5	13.6	20.8	5.3	758.1	578.6
along the disc thickness	1011.1	872.8	14.0	19.4	4.6	661.9	526.6

* Notch consists with groove direction.

** Notch consists with the intersecting surface of columnar grains.

growth. The highest value was 3-4 times the medium, and 7-8 times the lowest one.

2. Microstructure of ESR-cast alloy and its effect on properties

8 phases were found in ESR-cast GH136 alloy, as shown in table 2. Among them, the minor phases TiC, Ti(C, N), Y, Laves, $M_{23}B_{12}$, primary γ , σ etc. occurred in the interdendritic regions and along grain boundaries. They were probably products of segregation in the ESR cast process and amounted altogether to about 1%. γ' phase dispersed in the matrix; needle-like γ phase precipitated around the minor phases. They were secondary phases precipitated during the consequent cooling process after solidification of the castings. The typical as-cast structure is shown in Fig.3.

Fig.4 shows change of phases in the alloy after solution treatment at different tempera-

Table 2. Morphology, structure and composition of the phases in the alloy

Phase	Metallographic appearance	Composition	Crystal system
TiC	Grey, blocky	(Ti _{0.96} Mo _{0.04})C	F.C.C. a=0.436nm
Ti(C,N)	Yellow, blocky	(Ti _{0.96} Mo _{0.04})(C _{0.87} N _{0.13})	F.C.C.
Y	Red-brown, lath-like	Ti ₄ C ₂ S ₂	H.C.P. a=0.318nm c=1.112nm
Laves	White, blocky	(Fe _{0.50} Ni _{0.21} Cr _{0.15} Si _{0.14}) ₂ (Ti _{0.93} Mo _{0.07})	H.C.P. a=0.495nm c=0.815nm
M ₃ B ₂	Light blue-grey, rib-like	(Mo,Ti) ₃ B ₂	T. a=0.580nm c=0.310nm
σ			T. a=0.880nm c=0.454nm
primary	white, blocky	(Ni _{0.53} Fe _{0.45} Cr _{0.02}) ₃ (Ti _{0.61} Cr _{0.30} Mo _{0.09})	H.C.P. a=0.500nm c=0.815nm
Cellular	cellular		H.C.P. a=0.505nm c=0.823nm
Needle-like	Needle-like	(Ni _{0.85} Fe _{0.13} Cr _{0.02}) ₃ (Ti _{0.99} Mo _{tr.} Al _{tr.})	H.C.P. a=0.505nm c=0.822nm
γ'	Globular	(Ni _{0.89} Fe _{0.05} Cr _{0.06}) ₃ (Ti _{0.91} Al _{0.09} Mo _{tr.})	F.C.C. a=C.361nm C.362nm

Fig. 2 Effect of sampling direction on high stress ($\frac{\sigma_{min}}{\sigma_{max}} = 0.1$) LCF properties. Specimens: 1050°C, 4h. O.C., 700°C, 32h., A.C.

Fig. 3 As-cast microstructure of the alloy

ment below 1000°C, larger amounts of the minor phases and secondary η phase remained in the alloy. Their microstructure is shown in Fig. 5. It appears that the best solution treatment is at 1050°C.

The effect of aging at different temperatures after solution treatment at 1050°C for 4 hr has also been examined. The amounts of minor phases retained in the interdendritic region did not change in the range of the aging temperatures, but γ' and secondary η greatly changed, as is shown in Fig. 6. The precipitation peak is at about 750°C, it amounts 7.60%, 30nm. At the temperatures 750-1020°C η phase precipitates considerably, the peak is at about 900°C, amounts 1.66%. η phase has two morphologies, similar to that observed by B.R. Clark & F.B. Pickering. The η precipitated at grain boundaries is cellular. The space between Ni_3Ti plates in the cellu-

Fig. 4 Change in amount of phases in the alloy solution treated at different temperatures (held at temperature for 4 hr., oil cooled)

tures. It is seen that the higher the solution temperature, the more complete the resolution of the various phases back into the matrix. An eutectic occurs above 1100°C. With solution treat-

Fig. 5 Microstructure of the alloy solution treated at a. 980°C, b. 1050°C, c. 1150°C for 4 hr and oil cooled.

Fig. 6 Change in amount of phases in the alloy solution treated at 1050°C 4 hr., O.C., aged at the given temperature for 32h. and A.C.

lar η widens with increasing aging temperature. Needle-like η is precipitated around the minor phases in the interdendritic region. Fig. 7 shows the γ' phase and the two secondary η phases (cellular and needle-like) in the alloy after aging.

The effects of the η phase on the properties of the alloy can be seen from table 3. Precipitation of the η phase (no matter cellular or needle-like) results in a considerable improve-

Fig. 7 Microstructure of the alloy solution treated at 1050°C 4 h., O.C., aged at a. 700°C, b. c. 850°C, 32hr. and then A.C.

Table 3. Properties of alloy heat treated by different regimes

Heat treat- ment	Micro- struc- ture	R.T. tensile UTS MN/m ²	0.2% YS MN/m ²	%El. %RA	R.T. impact K _Q kg ^{1/2} M/ cm	R.T. Stress rupture prop. at 650°C 706.1 MN/m	650°C 706.1	650°C 706.1
						Notched h:min	N _f	fati- gue mm
980°C, 4h, OC, Needle- like η 700°C, 32h. A.C.		902.2	808.1	7.8 14.0	2.5 83.4	31:04 5.2 185:22	576	0.63
	phase and other primary phases							
1050°C, 4h. OC, 700°C, 32h.A.C.	no η phase	970.9	820.8	14.4 22.9	3.5 101.4	48:45 2.6 166:20	3187	0.94
1050°C, 4h. OC, 830°C, 4h.A.C., 700°C, 32h.A.C.	cellular η phase	951.2	807.1	9.8 17.0	2.4 97.1	26:00 11.0 ₁ 72:25	1077	0.32

* The specimens are perpendicular to the direction of columnar grain growth.

ment of thermal fatigue properties and stress rupture elongation, but also a deterioration of other properties.

In order to obtain the best combination of properties, the solution treatment should be at a temperature below the eutectic temperature but above the solution temperature for the secondary η phases. Aging should be conducted at the temperature below the precipitation temperature of secondary η phases. The experiments show that the best results are obtained by a solution treatment at 1050°C for 4 hr., oil cooling, aging at $700\text{--}750^{\circ}\text{C}$ for 32 hr. and then air cooling.

3. Structure and properties of the alloy aged for long periods

In order to simulate service conditions, heat treated specimens were subjected to long period aging at 650°C under stress 294.2 MN/m^2 , and the changes in structure and properties observed. It was found that after aging for 400 hr η phase had begun to precipitate, its amount increased from 4.72% to 7.2%. After aging for 1600 hr. there was no change in the amount of γ' phase, but its size had slightly increased, from 20 nm to 25 nm. Secondary η phase was less than 0.30%. There was a slight decrease in room temperature strength, impact value and an increase in elongation and reduction of area, as is shown in table 4.

Table 4 Room temperature mechanical properties of the alloy before and after stress aging

Conditions of specimens	R.T. tensile				R.T. impact kg-M/cm ²
	UTS MN/m ²	0.2%YS MN/m ²	%EL	%RA	
1050°C , 4h.O.C.	981.6	841.4	13.7	24.9	3.9
700°C , 32h.A.C.					
1050°C , 4h.O.C.	863.0	794.3	24.0	32.7	2.8
700°C , 32h.A.C.					
+ 650°C , 294.2 MN/m^2					
1600h. A.C.					

DISCUSSION

1. The radial unsymmetry of the macrostructure of the ESR-cast discs makes their high stress LCF properties highly dependent upon the columnar grain orientations. The highest LCF value is obtained along the direction of columnar grain growth, the lowest along the thickness of the discs. Obviously, the intersecting surface of the two rows of columnar grains is the weakest part of the disc under this kind of stress condition. However, during actual service, the discs would be subjected only to a very low stress in its thickness direction, so there would be no problem with high stress LCF here. Although the low cycle fatigue properties in the direction perpendicular to the columnar grain growth are lower than that in the parallel direction, they are superior to those for conventional die forged discs. Owing to the higher thermal fatigue properties obtained in the ESR-cast disc as compared to the die forged, the orientation dependence of its properties would not have any adverse effect in actual service.

The occurrence of cellular and needle-like η phase in the microstructure causes some decrease of room temperature ductility, toughness, fracture toughness, high temperature stress rupture prop. and high stress LCF, but no appreciable effect on notched bar stress rupture properties. That differs from the results obtained by Raymond F. Decker et al. (2) The cellular η phase has no considerable effect on the notched bar stress rupture properties. The reason probably lies in the fact that cellular η phase precipitates at grain boundaries, and in ERS-cast alloy the grains are coarse, so that less cellular η could precipitate. Results of long period stress aging at 650°C have indicated that it would have no adverse effect in actual application.

CONCLUSIONS

1. The ESR-cast-to-shape process can be used

for the manufacture of gas turbine discs. It would be economically advantageous for making large discs with diameter over say 800 mm.

2. The macrostructure of ESR-cast GH136 discs is radially unsymmetrical, so that their high stress LCF is orientation dependent, being lower along the disc thickness. This would have no adverse effect on the service properties, however other mechanical properties are not dependent on the orientation.

3. The types and morphology of the phases present in the ESR-cast alloys are practically the same as in the wrought alloy. Solution treatment at 1050°C for 4 hr., oil cooling and aging at $700-730^{\circ}\text{C}$ for 32 hr., air cooling give the best combination of properties.

REFERENCES

- (1) B.R. Clark and F.B. Pickering "Precipitation Effects in Austenitic Stainless Steels Containing Titanium and Aluminium Additions", Journal of the Iron and Steel Institute, Vol.205, 1967, Part 1, p. 70.
- (2) Raymond F. Decker and Chester T. Sims, "Metallurgy of Nickel-Base Alloys", The Superalloys, 1972.

J.Y.Shie (Daye Steel Works), C.B.Li, C.C.Chang, C.F. Lu, W.C. Shyu, G.S.Chen (Central Iron and Steel Research Institute, Beijing, China) also participated in this research work.