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Directionally solidified Soviet superalloy, ZHS6-K

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Directional solidification of the Soviet superalloy, ZHS6-K, has been carried out in an argon atmosphere. Tensile and stress–rupture properties have been measured for the investment cast and directionally solidified (DS) alloy. The DS alloy shows a several fold increase in rupture life and ductility compared with the investment cast alloy. It also shows improved tensile properties. Stress–rupture and tensile fracture behaviour has been examined.

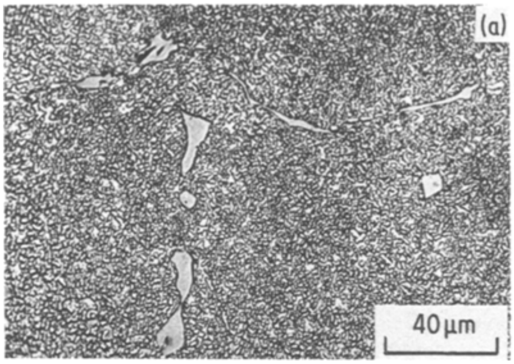
1. Introduction
Directionally solidified (DS) superalloy turbine blade and vane components are now being extensively used in most of the advanced gas turbine aeroengines. Their extensive application has come about because of higher DS component life compared with the investment cast ones, due to improved stress–rupture and thermal fatigue [1–8] resistance. Directional solidification, by eliminating the grain boundaries transverse to the centrifugal loading axis of the blade has generally been observed to result in improved stress–rupture life and ductility of the superalloys. However, in the case of IN-738 [9] such improvement was not observed. While the response of the Western superalloys to directional solidification has been extensively investigated and reported in the literature, little has been reported on the Soviet alloys.

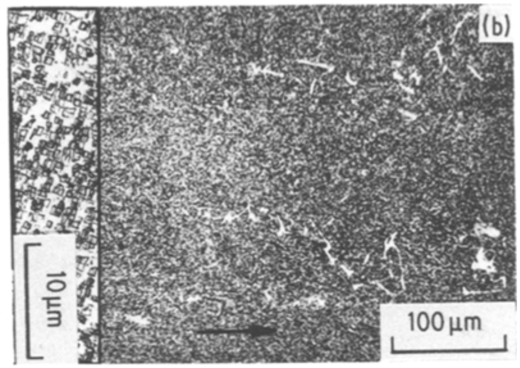
ZHS6-K, a nickel base alloy, [10], is a high-performance Soviet superalloy suitable for applications as investment cast gas turbine engine blade or vane components. The purpose of this work was to investigate the response of ZHS6-K to directional solidification.

2. Experimental procedure
1.1 cm diameter bars were vacuum investment cast from an alloy prepared with a nominal composition, Ni–11 Cr–5.0 W–5.0 Al–4.5 Co–4.0 Mo–2.5 Ti–0.15 C–0.02B (weight per cent). These investment cast bars were directionally solidified in a modified Bridgeman apparatus similar to the one reported in an earlier paper [11]. The alloy bar kept in a cylindrical 1.2 cm diameter recrystallized alumina crucible and was melted (except for the bottom 2 cm which was held in the withdrawal chuck) in a flowing argon atmosphere by radiation from a graphite susceptor which was inductively heated. Soon after melting the alumina crucible was withdrawn through a copper chill ring at 20 cm h^{-1}. The directional solidification was carried out with a temperature gradient of about 150° C cm^{-1} in the liquid at the liquid–solid interface.

Tensile (gauge diameter = 4 mm and gauge length = 26 mm), and stress–rupture (gauge diameter = 3 mm and gauge length = 15 mm) specimens were machined from the investment cast and directionally solidified alloy bars, both with and without heat treatment (1200° C for 4 h, in air followed by air cool to room temperature). Tensile tests were conducted in air at 700 and 900° C at a cross-head speed of 0.1 cm min^{-1}. Stress–rupture tests were conducted in air at 800° C/53 kg mm^{-2} and 900° C/33 kg mm^{-2}. In the case of DS specimens the loading axis was parallel to the columnar grain growth (crucible withdrawal) direction.

3. Results and discussion
3.1. Microstructure
In the investment cast version the microstructure (Fig. 1a) consists of an fcc matrix (gamma, γ) containing about 60 vol % fine intermetallic gamma prime, γ', precipitates. Primary MC-type carbides precipitate at the interdendritic and grain boundary regions, and provide strength against high-tempera-





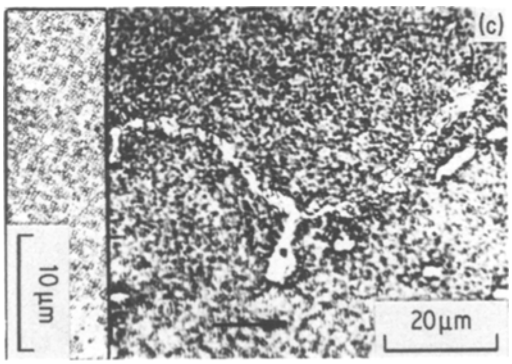


Figure 1 Microstructure of alloy ZHS6-K: (a) investment cast; (b) directionally solidified (longitudinal) and (c) directionally solidified + heat treated (longitudinal). (Insets are electron replica micrographs showing the cuboidal γ' precipitates. Arrow shows the directional solidification direction.)

ture grain-boundary sliding (Fig. 1a). These carbide particles which precipitate out during the last stages of solidification and choke the liquid flow through the interdendritic regions are believed to lead to microporosity in these regions for the investment cast components. This alloy has been observed to be more prone to the formation of shrinkage cavities compared with an equivalent superalloy IN-100 [12]. Occasionally in the microstructure fan-like eutectic "nodular" γ' precipitates [12] were also observed.

Directional solidification of the alloy resulted in a microstructure with primary dendrites aligned parallel to the alloy growth direction. Some secondary dendrite formation was also observed. Carbide particles precipitated in flake shapes in the interdendritic regions (Fig. 1b) mostly parallel to the alloy growth direction (between primary dendrites), some transverse to it (between secondary dendrites) and some along the grain boundaries running parallel to the alloy growth direction.

Heat treatment of the alloy leads to the dissolution of γ' and reprecipitation of finer γ' precipitates (compare inset in Fig. 1b with that in Fig. 1c), some blunting of the sharp edges of the MC carbide flakes and occasional formation of some very fine precipitates along the grain boundary (Fig. 1c) which may be $M_{23}C_6$ type carbide precipitates.

3.2. Stress rupture properties

Investment cast specimens, both with and without heat-treatment, generally showed a large scatter in rupture life (Table I). This may be due to the specimen porosity and is similar to the usual scatter in rupture life observed for cast superalloys. Rupture ductilities observed were generally low, 0.7 to 2.6% elongation and 1.2 to 6.7% area reduction (RA). Metallographic examination showed the fracture to be intergranular running along the grain boundaries which are located approximately perpendicular to the load axis, for both the stress–rupture test conditions, Fig. 2a. Such intergranular fracture running perpendicular to the load axis is expected for the investment cast alloys, because both the "R" and "W" type cracks nucleate and grow on these boundaries [13].

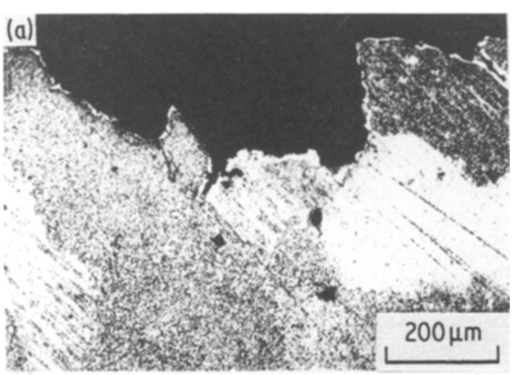
Directional solidification results in about 11 times improvement in rupture life at 800° C and about 3 times improvement at 900° C compared with the maximum life observed for the investment cast version of this alloy, Table I. Such a decrease in rupture life improvement with increasing temperature has been reported for other DS superalloys [1, 4]. It is believed to result from increasing plastic deformation at increasing temperature in the grain interior, accommodating

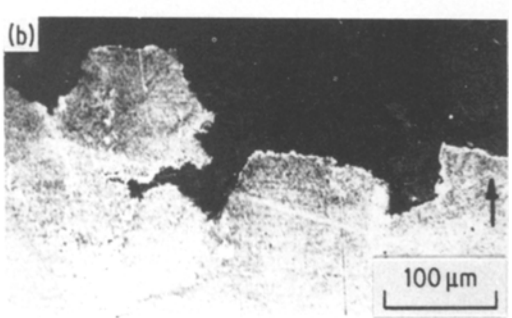
TABLE I Temperature dependence of the stress–rupture properties of ZHS6-K (IC: investment cast; HT: heat treated at 1200° C for 4 h followed by air cool to room temperature; DS: directionally solidified)

Sample number	Specimen condition	800° C/53 kg mm^{-2}			900° C/33 kg mm^{-2}		
		Life (h)	Percentage elongation	Percentage RA	Life (h)	Percentage elongation	Percentage RA
1	IC	54.7	1.8	6.7	12.6	2.2	2.2
2	IC	18.2	1.4	1.7	58.7	0.7	1.6
3	IC + HT	25.5	0.7	1.2	37.2	2.6	–
4	IC + HT	34.1	0.7	1.1	57.5	2.0	3.7
5	DS	581.2	17.7	22.6	156.4	29.6	40.8
6	DS	566.6	–	45.8	125.2	32.6	42.4
7	DS + HT	796.7	20.4	35.2	184.7	25.6	46.4
8	DS + HT	751.8	–	31.4	150	25.3	40.6

grain-boundary sliding more and more effectively. This makes the presence of transverse grain boundaries progressively less detrimental as a source of rupture cracking. Directional solidification has also resulted in a several-fold increase in the alloy rupture ductility (Table I). Heat-treatment of the DS alloy results in further significant improvement in its rupture life at 800° C (Table I). However at 900° C no such improvement with heat treatment was observed. This rupture life advantage at 800° C is possibly due to finer γ' precipitates in the DS + HT alloy compared with the DS alloy. For the higher temperature 900° C rupture test, this advantage is lost due to Ostwald ripening of γ' precipitates during testing. If the rupture lives obtained for the DS + HT ZHS6-K alloy are plotted on the stress against Larson Miller number graph and compared with a similar plot for DS MARM-200 alloy [14], the two alloys appear to be equivalent in the temperature range examined. A major difference between MARM-200 [14] and ZHS6-K alloy chemistries is the absence of zirconium in ZHS6-K. ZHS6-K is slightly richer in chromium (11% against 9%) and poorer in cobalt (4.5% against 10%). However, the contents of γ' forming elements (Ti, Al), and the refractory elements (W/Mo) are equivalent. MARM-200 also has about 1% niobium, which is absent in ZHS6-K. Carbon and boron contents of both the alloys are similar.

Fig. 2b shows the typical rupture fracture profile observed for the DS specimens. The fracture cuts through the primary dendrites. However it selects the interdendritic path between secondary





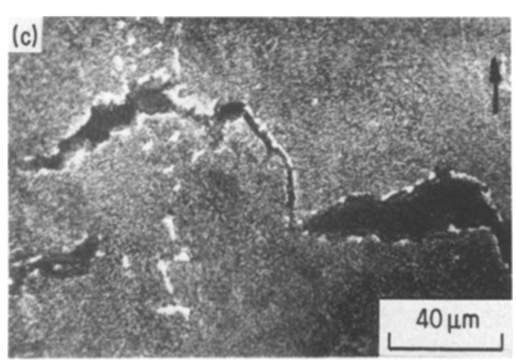


Figure 2 Section through stress–rupture fracture surface of ZHS6-K: (a) investment cast (900° C/33 kg mm^{-2}–12.6 h life); (b) directionally solidified (longitudinal section, DS + HT, 800° C/53 kg mm^{-2}–796.7 h life) and (c) secondary fracture in DS alloy (longitudinal section). (Arrow shows the load axis.)

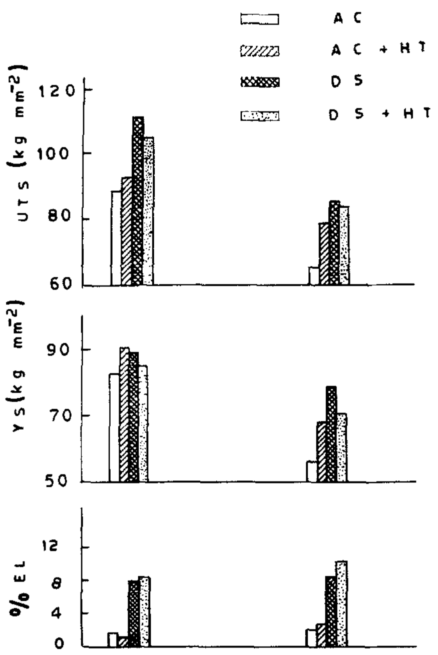


Figure 3 Temperature dependence of the tensile properties of ZHS6-K.

dendrites. It is more clearly evident in Fig. 2c, where a secondary fracture below the main fracture surface is shown. The crack appears to nucleate on the carbide–matrix interface along the interdendritic regions (between secondary dendrites) and grows in the interdendritic manner.

The other major source of rupture crack nucleation observed for the DS specimens were the interdendritic regions (between secondary dendrities) connected to the gauge surface of the specimen exposed to air at high temperatures. Preferential oxidation of the carbides occurred along these regions, thus making a crack available perpendicular to the load axis. This crack slowly spread by repeated cycles of oxidation and crack growth and resulted in rupture failure of the specimen. This mode of failure is expected to be very crucial for thin-walled DS blades and can only be avoided by providing an oxidation resistant coating on the surface.

3.3. Tensile properties

The heat treatment generally results in improvement of the tensile properties over the investment cast condition (Fig. 3). The tensile fracture for the investment cast specimens is mostly intergranular both at 900 and 700° C (e.g., Fig. 4a). Directionally solidified alloy has similar tensile property values both before and after the heat treatment. It shows considerable ductility advantage over the investment cast specimens. Directional solidification has also resulted in generally improved ultimate tensile strength (UTS) without adversely affecting the yield strength. The fracture is transdendritic (Fig. 4b), i.e. the fracture path cuts across the primary dendrites. The transdendritic portion of the fracture path was observed to be more jagged for 900° C test specimens compared with the 700° C ones. There were more secondary voids created below the fracture surface for tests conducted at 900° C compared with tests at 700° C. The tensile fracture in the DS specimens was observed to be initiated either due to the long carbide particles located in the interdendritic regions fracturing perpendicular to the load axis, or to grain boundaries pulling apart at regions where they are perpendicular to the load axis. The ductility and UTS advantages of the DS alloy can be attributed to the absence of transverse grain boundaries and reduction in the carbide precipitate size compared with the investment cast alloy, as well as the complete elimination of microporosity.

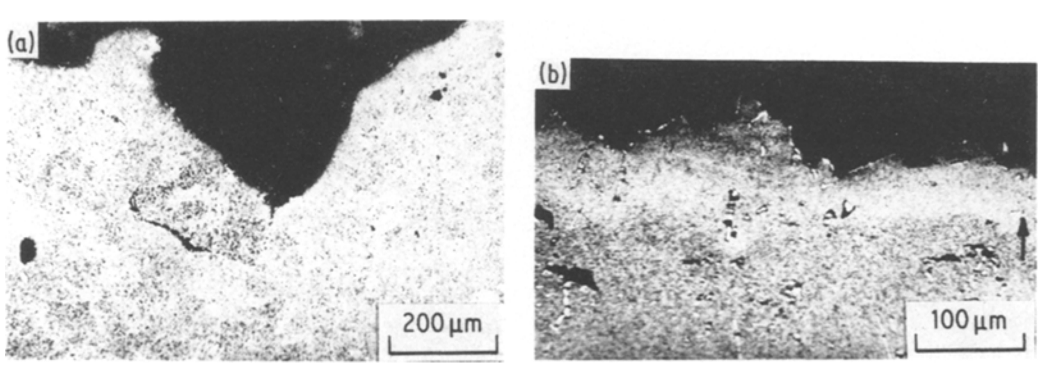


Figure 4 Elevated temperature tensile fracture profile of ZHS6-K: (a) investment cast and (b) directionally solidified (longitudinal section, 900° C test).

4. Conclusions

The following conclusions can be drawn from this investigation for the nickel base Soviet superalloy, ZHS6-K.

(1) Directional solidification of the alloy results in a several-fold increase in the stress–rupture life and ductility compared with the investment cast condition. The stress–rupture life can be further improved by heat treatment. High-temperature, oxidation resistant coatings are expected to further improve the rupture life, especially for thin sections.

(2) Directional solidification also results in improved ultimate tensile strength and tensile ductility for the alloy without adversely affecting its yield strength.

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