JMEPEG (1995) 4:48-53

Development of Design Principles for a Creep-Limited Alloy for Turbine Blades

Z. Huda

Recent advancements in turbine-blade materials engineering are reviewed in light of general superalloy research and the author's work on a new powder metallurgy IN-792 creep-limited alloy for application in blades of gas-turbine engines. The developed set of principles presented in this paper incorporates all the factors that must be taken into consideration in selecting and designing an alloy for turbine blades.

Keywords
alloy design, creep, IN-792, powder metallurgy, turbine blades

1. Introduction

MANUFACTURERS of gas-turbine engines are driven by marketplace economics to control the product/cost ratio (Ref 1-3). A significant factor in the cost of operating and maintaining aircraft gas-turbine engines is the frequent replacement of turbine blades that have failed or have reached some design limit. The failure of a turbine blade is caused by: (1) exceeding the prescribed stretch limit, (2) exceeding the prescribed twist limit, (3) exceeding service operating temperatures, or (4) erosion of the leading edge (Ref 4). Most damage can be attributed to creep, the rate of which is chiefly influenced by stress and temperature (Ref 5).

Turbine blades are usually manufactured by a net-shape investment casting process, which is economically attractive. Notable developments in blade casting include single-crystal (SX) and directionally solidified (DS) turbine blades with improved creep resistance (Ref 6-9). However, there can be porosity-related problems in cast blades, leading to fairly low fatigue properties (Ref 10, 11). The use of powder metallurgy (P/M) techniques can overcome these technical drawbacks (Ref 8, 12). Therefore, forged blades of stronger alloys are preferred for superior fatigue resistance. However, marine gas turbines still employ investment casting alloys with appropriate processing techniques. Recently, DS investment casting alloys IN-6201 and IN-6203 have been developed at INCO Alloys Ltd. to improve engine power as well as increase resistance to creep and aggressive marine environments (Ref 13).

Alloys developed for application in blades of modern gas turbines, particularly for aircraft engines, chiefly include MAR-M 200, MA 6000, IN-738, and IN-792 (Ref 7, 12, 14). Oxide-dispersion-strengthened (ODS) alloys offer higher creep resistance than any other superalloy at temperatures higher than 800 to 900 °C. The most notable advancement in aircraft engine turbine blades is ODS alloy MA 6000, developed at Garrett AiResearch (Ref 15, 16). In general, grain-coarsened P/M superalloys and the ODS alloys find chief application in turbine blades (Ref 17-19).

A notable advancement in forged blading is a powder-modified IN-792 nickel-base superalloy, APK-6, developed at INCO (Inco Alloys Ltd., Hereford, U.K.)(Ref 20). An appropriate grain-growth heat treatment has been developed by the author for this alloy, making it suitable for application in turbine blades (see section 3.6). This heat treatment involves controlled annealing at subsolvus temperature, resulting in grain growth accompanied by precipitation of a high volume fraction of fine γ' particles at grain boundaries (Ref 17, 21).

2. Analyzing Creep Problems

An aircraft turbofan engine is the best subject for analysis of a turbine-blade material, as the stress and temperature environments are most severe in this type of engine. At takeoff, the blade is subjected to a stress of approximately 250 MPa, and design specifications require that this stress be supported for 30 h at 850 °C without more than a 0.1% irreversible creep strain (Ref 22).

Due to rotation, the blade is subjected to centrifugal stresses of approximately 275 MPa in the last stages of an aircraft engine fan turbine (Ref 2). These operating conditions of stress and temperature demand that creep resistance be given the most attention in the design of a turbine-blade material. This concern becomes most serious in the high-pressure stage, where the material must be resistant to dislocation (power law) creep.

3. Solving Creep Problems

Achieving good creep resistance in a material involves analyzing the variables on which creep strength depends. Creep strength is a function of a number of variables, including crystal structure, solutes, precipitates, dispersoids, grain size (and its distribution), and grain-boundary engineering (grain-boundary structure and chemistry). The effects of each are discussed in the following sections.

3.1 *Effect of Crystal Structure*

A close-packed crystal structure that is stable at temperatures approaching the melting point favors creep resistance. The face-centered cubic (fcc) lattice of nickel serves this purpose. Furthermore, the presence of refractory metals in the alloy stabilizes the lattice at high temperatures. Thus, the

---

Z. Huda, Metallurgical Engineering Department, Dawood College of Engineering & Technology, Karachi 74800, Pakistan.

requirement of a favorable crystal structure is fulfilled by the general composition of nickel-base superalloys.

3.2 *Effect of Solutes*

Steady-state creep in crystalline single-phase solids is related to diffusivity (*D*), stacking fault energy (γ_{SFE}), elastic modulus (*E*), temperature (*T*), and stress (σ) according to (Ref 23, 24):

$$\dot{\varepsilon} = A\left(\frac{\sigma}{E}\right)^n f(\gamma_{SFE})\, e^{-Q/RT} \qquad \text{(Eq 1)}$$

where $f(\gamma_{SFE})$ is a function of γ_{SFE} and *Q* is the activation energy for creep; values of *n* lie in the range of 3 to 7. It can be seen from Eq 1 that solute additions that raise the modulus of elasticity or that lower the SFE and the diffusivity will favor high creep strength (Ref 18). Cobalt is effective in lowering the SFE of nickel-base superalloys.

3.3 *Effect of Precipitates*

Stable and hard precipitates of compounds such as Ni_3Al, Ni_3Ti, MoC, and TaC obstruct dislocations and thus improve creep strength. In particular, precipitation of γ' approximating the formula $Ni_3(Ti,Al)$ ensures that creep resistance is maintained for a long period of time. The mode of γ' precipitation has a pronounced effect on creep resistance and is discussed in the following sections.

3.3.1 Coarsening of γ'

Measures that minimize γ' coarsening will help to retain long-time creep resistance (Ref 25).The coarsening rate of γ' is retarded significantly by addition of cobalt, molybdenum, chromium, or a combination of molybdenum and tungsten in nickel-base superalloys. Increasing niobium content by 2 to 5% also markedly reduces the coarsening rate of γ' in these superalloys.

3.3.2 Heat Treatment for Fine γ'

In most SX nickel-base superalloys, a precipitate size of about 0.5 μm leads to optimum creep resistance (Ref 8). Heat treatment experiments on P/M René 95 showed that air cooling from the solution temperature resulted in γ' of approximately 0.2 μm in size (Ref 26). Similar results were obtained for the alloy NASAAIR 100 (Ref 9, 27). Therefore, it can be concluded that the more rapid the cooling, the finer the size distribution of γ' in the microstructure. Fine γ' particles have been successfully achieved in P/M superalloy APK-6 by the author (see section 3.6).

3.3.3 Volume Fraction of γ'

A high value fraction (>0.5) of γ' enhances creep resistance in nickel-base superalloys. Dreshfield (Ref 27) and Chang et al. (Ref 26) have studied the effect of subsolvus annealing on the amount of γ' in René 95. The fact established in these reports is that the solution temperature during annealing must be subsolvus and as high as possible to achieve a high volume fraction of γ'. One way of achieving this high solution temperature is to increase the melting point by extra addition of elements with high T_m (Ref 27) or by eliminating alloying elements such as carbon, boron, zirconium, and hafnium (Ref 28). The latter is achieved by eliminating grain boundaries through the manu The absence of grain boundaries in SX blades makes possible the removal of alloying elements, such as C, B, Zr and Hf that are used to strengthen the grain boundaries. Removal of these elements causes a significant increase in the melting point in most superalloys.

3.3.4 Gamma/Gamma-Prime Misfit

Alloys with low γ/γ' misfit show better creep resistance than those with high misfit (Ref 18). Partitioning of solute elements between γ and γ' is the best means of controlling misfit. Titanium and niobium partition to γ' and increase its lattice parameter; chromium, molybdenum, and iron tend to partition to γ. To approach zero misfit, secondary elements that when added partition to γ' should be balanced by those that partition preferentially to γ.

3.4 *Effect of Dispersoids*

Dispersoids such as oxides, which are incoherent with the matrix, occur in mechanically alloyed (MA) superalloys and favor creep resistance. Dispersion strengthening requires a fine, uniform distribution of the dispersoids. For example, yttrium oxide in a superalloy matrix requires an average interparticle spacing of less than 500 nm for significant strengthening (Ref 14).

An additional factor is the grain aspect ratio (GAR): the ratio of grain length to width. Improvement of creep-rupture properties due to high GAR has been confirmed in MA 753 (Ref 30) and in MA 6000 (Ref 31). The most direct means of achieving coarse, elongated grains is to use an extrusion press both to consolidate powder and to achieve an appropriate structure suitable for subsequent anomalous grain growth.

3.5 *Effect of Grain Size*

The effect of grain size on creep resistance can be numerically estimated by:

$$\text{Rate of creep} = C\left(\frac{\sigma D}{d^2}\right) \qquad \text{(Eq 2)}$$

where C is a constant, *D* is a diffusion coefficient, σ is applied stress, and *d* is grain size. Obviously, increasing the grain size will significantly retard the rate of creep. The consideration of grain size and its distribution is of great technological importance in P/M superalloys. These extremely fine-grained materials require grain growth to retard creep.

Anomalous grain growth with a large grain size has been suggested as a means for producing artifacts for resistance to high-temperature creep (Ref 32). This microstructural feature (abnormal grain growth with a grain size of 175 μm) has been observed by the author in APK-6 alloy (powder-formed IN-792) after prolonged annealing at high temperature (1270 °C/22 h) (see Fig. 1). However, a limitation is believed to have been imposed in this microstructure (total dissolution of γ'

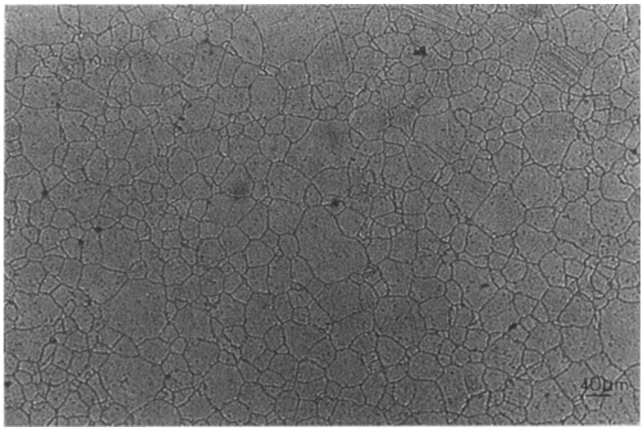

Fig. 1 Anomalous grain growth in heat-treated (1220 °C/22 h) APK-6 alloy showing total dissolution of γ′ particles due to high-temperature annealing

phase), which does not ensure suitability of this material for application in turbine blades (Ref 16).

Cast-to-shape SX turbine blades have shown high creep strength (Ref 8) (see also section 3.3.3). However, grain-coarsened P/M superalloys are preferred in turbine blades for their superior fatigue strength (Ref 1). An annealing temperature in the range of 1205 to 1232 °C has been suggested for the P/M IN-792 alloy to obtain grain size larger than 76 µm (Ref 33); the effect of grain growth on the creep strength of this alloy is discussed in the following section.

3.6 *Improving Creep Strength in P/M IN-792*

3.6.1 Mechanical Test Reports

Superalloys produced by P/M possess excellent tensile strength, fatigue strength, and toughness at low and moderate temperatures because of their fine grain size. However, grain-boundary sliding becomes a problem at high temperatures, leading to creep-related phenomena (Ref 21, 34). This requires an additional processing of grain growth to improve creep resistance by reducing grain-boundary area. Experiments on P/M IN-792 have shown that values of initial stress intensity required for failure under creep crack growth increase from 16 to 40 MN · $m^{-3/2}$ as grain size increases from 10 to 100 µm (Ref 35).

3.6.2 Metallurgical Investigation Reports

An extensive microscopic study of APK-6 alloy (powder-modified IN-792) has been made by the author. This alloy possesses fatigue strength that is two and a half times greater than that of conventionally cast alloys (Ref 20). A series of grain-growth heat treatments were carried out to render APK-6 alloy useful for application in turbine blades (Ref 1, 17, 21, 36-38). Some of the results of these metallurgical investigations, which indicated improvement in creep resistance, are presented in the following sections.

3.6.2.1 *Experimental*

Details of manufacturing, chemical composition, and metallographic practice for APK-6 alloy are given elsewhere (Ref 21, 38). Metallographic samples of this alloy were heat treated according to the scheme shown in Table 1. Microstructural characterization involved the use of an optical microscope (linked with a computerized image analyzer) and a transmission electron microscope operating at a high voltage of 200 KV.

Table 1 APK-6 alloy sample designation according to heat treatment

Sample designation	Heat treatment
A	1220 °C/20 min
B	1220 °C/2 h
C	1220 °C/22 h

3.6.2.2 *Results and Discussion*

A transmission electron micrograph of sample D (as-received alloy) is shown in Fig. 2. An optical micrograph of sample A is shown in Fig. 3. Micrographs for samples B and C are shown elsewhere (Ref 17, 21). Image analysis data for samples A to D are presented in Table 2.

Figure 2 and the grain size and particle size values for sample D given in Table 2 indicate that the as-received alloy is an extremely fine-grained material with particle sizes comparable to grain size. These microstructural features lead to the conclusion that the APK-6 alloy, in the as-received condition, is unsuitable for application in high-temperature service.

Image analysis data for samples B and C (Table 2) indicate that, compared to sample A, a significant coarsening of γ′ particles has occurred, with no appreciable grain growth. A severe inhibition in grain growth is apparent due to particle-pinning effects (Ref 39). Because the coarsening characteristics of γ′ particles are crucial (see section 3.3.1), the suitability of alloys B and C in terms of their use in turbine blades is also questionable.

Figure 3 and the data in Table 2 show that in alloy A a significant volume fraction of fine γ′ particles (0.7 µm average particle size) is precipitated, accompanied by a sufficient degree of grain growth (14 µm average grain size). This microstructural feature is thought to be favorable for imparting good creep resistance in this heat-treated (1220 °C/20 min) material.

3.6.2.3 *Conclusions*

Annealing of APK-6 alloy at a subsolvus temperature for a short duration results in a microstructure that, based on metallurgical principles, appears to be appropriate for application in turbine blades.

3.7 *Grain-Boundary Engineering*

Under creep conditions, a primary concern involves the ability of the grain boundaries to resist sliding and to avoid the accretion of voids from vacancy condensation (Ref 34). At the simplest level, these effects can be minimized or removed by obtaining specific grain geometries or by controlling the amounts of alloying elements that decorate the grain boundaries in a favorable manner. An ideal grain-boundary condition in an engineering alloy is achieved by proper control of grain-boundary structure as well as of grain-boundary chemistry.

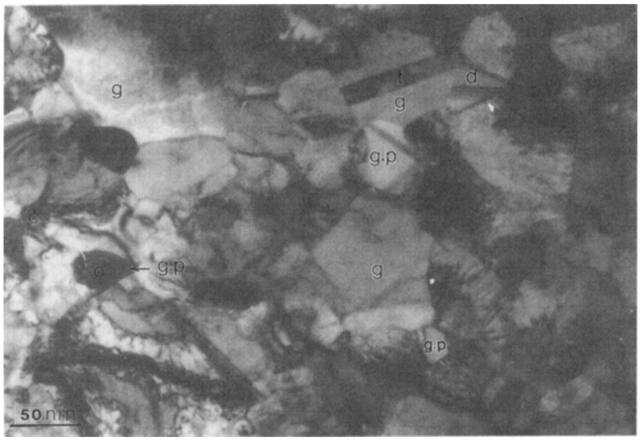

Fig. 2 Transmission electron micrograph of APK-6 alloy in the as-received condition (sample D) showing an extremely fine-grained multiphase microstructure. g, gamma; g.p., gamma prime; c, carbide; t, twin; d, dislocation

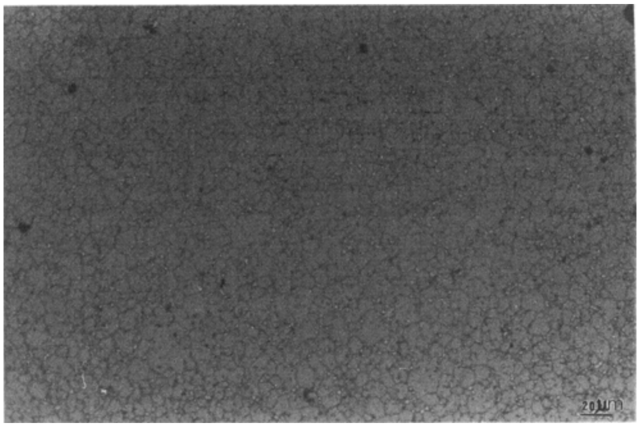

Fig. 3 Optical micrograph of APK-6 alloy heat treated at 1220 °C/20 min (sample A) showing precipitation of fine γ' particles due to short-duration annealing

Table 2 Average grain size and γ' particle size for samples A to D

Sample designation	Average grain size, μm	Average γ' particle size, μm
A	14	0.7
B	14.2	2.8
C	15.3	6.0
D	0.1	0.05

3.7.1 Grain-Boundary Structure

Achievement of creep strength by obtaining specific grain geometries is possible by directional solidification or by the single-crystal blading technique (Ref 8, 22, 34). A grain-boundary serration in heat-treated (at controlled cooling rates) P/M IN-792 has been reported to significantly improve resistance to creep crack growth (Ref 12).

An alternate approach is to design the microstructure in the grain-boundary regions–for instance, by the incorporation of grain-boundary precipitates, which help to limit sliding (see Fig. 3). Another possibility is to modify the microstructure so as to maximize the percentage of grain boundaries within the polycrystals that have low sliding and vacancy-capture characteristics. This can be accomplished by modifying texture, particularly the grain misorientation texture (GMT), which is defined as the difference in orientation between two grains (Ref 34). Recently, a detailed local geometry (crystallography) of grain boundaries has been determined for P/M-modified IN-792 (Ref 36).

3.7.2 Grain-Boundary Chemistry

For good creep resistance, grain boundaries must be stronger than the interiors of the grains. Addition of hafnium has been found to be very effective in preventing creep failure in an intergranular manner (Ref 25). Small additions of boron and zirconium enhance creep resistance in nickel-base superalloys (Ref 40). The presence of boron at grain boundaries blocks the upset of tearing under creep-rupture conditions.

Carbides play an essential role in improving the creep strength of superalloys. Carbide particles pin the grain boundaries and inhibit grain-boundary sliding or migration. The general goal for creep strength is a regular arrangement of small (0.1 to 0.3 μm), stable, and discrete carbide particles at the grain boundaries (Ref 29).

A profound beneficial effect of grain-boundary chemistry on creep resistance has been noticed. Blocky carbides ($M_{23}C_6$ and M_6C) and γ' particles are produced by the following reactions:

$$MC + \gamma \rightarrow M_{23}C_6 + \gamma' \qquad \text{(Eq 3)}$$

$$MC + \gamma \rightarrow M_6C + \gamma' \qquad \text{(Eq 4)}$$

The precipitate particles produced by Eq 3 and 4 decorate the grain boundaries; the γ' so generated engulfs the carbides and the grain boundary in a relatively ductile and creep-resistant layer (Ref 25).

3.8 *Creep-Limited Alloy Design Principles*

Having established the means to achieve a creep-limited alloy (see sections 3.1 to 3.7), a developed set of principles is now presented. Table 3 presents the design principles for a creep-limited alloy for turbine blades. It also incorporates developments in turbine-blade materials and manufacturing processes for meeting the demands of current and future engine requirements

4. Summary

An improved model of design principles for a creep-limited turbine-blade alloy has been developed. The main features are summarized below:

Table 3 Design principles for a creep-limited alloy

Factors controlling creep resistance	Alloy characteristics for good creep strength	Chemical means for creep-resistant alloy	Process design for optimum creep strength
Crystal structure	Close-packed, stable to T_M. (Choose fcc lattice to dissolve many atoms).	Nickel matrix. Add refractory metals to the alloy in appropriate amounts.	...
Solutes	High modulus, low fault energy, slow diffusivity of the matrix	Add W, Mo, and Co to the alloy in appropriate amounts.	...
Precipitate (γ')	Fine and hyperfine, large volume fraction, low misfit, high fault energy	High-nickel matrix. Add Co, Mo, Cr, or a combination of Mo and W; add Mo up to the solid-solubility limit of Mo in γ'. Strike a balance between (Ti + Nb) and (Cr + Mo + Fe).	Try to achieve 0.05 μm particle size; heat treatment solution temperature must be subsolvus, but as high as possible. For stress annealing, use crystals in the $\langle 100 \rangle$ orientation.
Grain size	Coarse	...	Induce grain growth by annealing. Achieve anomalous grain growth with a large grain size.
Dispersoids	Large volume fraction, stable, high GAR; fine, uniform distribution	For Y_2O_3, set an average interparticle spacing of 500 nm.	Use an extrusion press both to consolidate powder and to achieve an appropriate structure.
Grain-boundary structure	Eliminate grain boundaries; obtain elongated grains or serrated grain boundaries; design the microstructure in the grain-boundary region; or modify the GMT.	...	Make single-crystal blades or directionally solidify; anneal with controlled cooling rate; or incorporate grain boundaries with precipitates.
Grain-boundary chemistry	Add proper amounts of carbides, Zr and Mg (in some cases), and γ' at grain boundaries.	Achieve regular arrangement of 0.1 to 0.3 μm stable carbide particles at grain boundaries. Add proper amounts of Hf.	Heat treat to form γ' and $M_{23}C_6$ simultaneously so that γ' may engulf $M_{23}C_6$ and grain boundaries to form a creep-resistant layer.

- A nickel-base superalloy with high-melting-point alloying elements, high-modulus solutes, low fault energy, and slow diffusivity offers good thermal stability and resistance to creep.
- Creep strength is improved by any factor that increases the stability of the precipitated phase, particularly the γ' phase.
- Oxide dispersion strengthening and mechanical alloying can produce alloys with superior creep resistance if proper processing conditions are adopted.
- Excellent creep strength can be ensured by single-crystal or directional-solidification blading techniques.
- Grain coarsening in P/M superalloys significantly improves creep resistance if fine γ particles are precipitated.
- For good creep strength, grain boundaries must be stronger than the interiors of the grains. A creep-limited alloy can thus be produced by controlling the grain-boundary engineering (grain-boundary structure and chemistry) by appropriate alloying, processing, and heat treatment procedures.

Acknowledgments

The author is grateful to Dr. John R. Ogren of Helsinki University of Technology (Finland) for his suggestions regarding the general presentation of this paper and regarding the DS blading technique. The author is also indebted to Dr. R. Dreshfield of NASA Lewis Research Center for providing information about low-cost SX turbine blades. The provision of laboratory facilities by the Experimental Techniques Centre and the Department of Materials Technology (Professor Brian Ralph) at Brunel University of West London is gratefully acknowledged.

References

1. Z. Huda, paper presented at 3rd Int. Symp. on Advanced Materials (Islamabad), 20-24 Sept 1993.
2. H.E. Miller and W.L. Chambers, in *Superalloys II*, C.T. Sims et al., Ed., John Wiley & Sons, 1987
3. W. Betteridge and S.W.K. Shaw, *Mater. Sci. Technol.*, Vol 3, Sept 1987, p 643
4. M.A. Imam, K. Sadananda, and J. Feng in Abstracts of the 3rd Int. Symp. on Adv. Mat. (Islamabad), 1993
5. V.B. John, *Engineering Materials*, Macmillan, 1990
6. Z. Huda, *Eng. Rev.*, Vol 17 (No. 11), 1992
7. C.T. Sims, in *Superalloys II*, C.T. Sims et al., Ed., John Wiley & Sons, 1987
8. T. Khan, *Mater. Sci. Technol.*, Vol 2 (No. 5), 1986
9. T.E. Strangman, B. Heath, and M. Fuji, Garrett 21-4314-1, CR-168218, Vol I, Garrett Turbine Engine Co., NASA, 1983
10. W.R. Freeman, in *Superalloys II*, C.T. Sims et al., Ed., John Wiley & Sons, 1987
11. W. Hoffelener, *Metall. Trans. A*, Vol 13A, 1982, p 1245
12. G.H. Gessinger, *Powder Metallurgy of Superalloys*, Butterworths, 1984
13. S.W.K. Shaw and M.J. Fleetwood, *Mater. Sci. Technol.*, Vol 5, 1989, p 925
14. G.A.J. Hack, *Met. and Mater.;* August 1987, p.457
15. G.A.J. Hack; *Met. Powder Rep.;* Vol. 36, 1981, p.425
16. G.S. Hoppin and F.A. Schweizer, in *Proc. INCO MAP Conf.*, 1981, p.75
17. Z. Huda, Ph.D. thesis, Brunel University of West London, 1991
18. N.S. Stoloff, in *Superalloys II*, C.T. Sims et al., Ed., John Wiley & Sons, 1987
19. E. Arzt and R.F. Singer, in *Superalloys '84*, M. Gell et al., Ed., TMS-AIME, 1984
20. S.W.K. Shaw, Tech. Publ. No. P-BL-460, INCO Engineered Ltd., Birmingham, U.K., 1987
21. Z. Huda and B. Ralph, *Prakt. Metallogr.*, Vol 27, 1990, p 64

22. F.M. Ashby and D.R.H. Jones, *Engineering Materials*, Vol 1, Pergamon Press, 1980
23. O.D. Sherby and P.M. Burke, *Prog. Mater. Sci.*, Vol 13, 1967, p 32
24. T. Mukherjee, W.E. Stumph, C.M. Sellars, and W.J.McG. Tegart, *J. Iron Steel Inst.*, Vol 207, 1969, p 621
25. E.W. Ross and C.T. Sims, in *Superalloys II*, C.T. Sims et al., Ed., John Wiley & Sons, 1987
26. D.R. Chang, D.D. Kruger, and R.A. Sprague, in *Superalloys '84*, M. Gell et al., Ed., TMS-AIME, 1984
27. R.L. Dreshfield, *J. Met.*, Vol 18, 1987, p 16
28. Ver Snyder and M.E. Shank, *Mater. Sci. Eng.*, Vol 6, 1970, p 235
29. S. Floreen, in *Superalloys II*, C.T. Sims et al., Ed., John Wiley & Sons, 1987
30. J.K. Tien and S.M. Copley, *Metall. Trans.*, Vol 2, 1971, p 543
31. D.D. Pearson, B.H. Kear, and F.D. Lamkey, in *Creep Fracture of Engineering Materials*, Pineridge, 1981
32. B. Ralph and R.C. Ecob, in *Proc. 5th Int. Conf. on Metals and Materials Science*, Riso Laboratories, Denmark, 1984
33. J.M. Larson, *Metall. Trans. A*, Vol 7A, 1976, p 1497
34. V. Randle and B. Ralph, *Rev. Phys. Appl.*, Vol 23, 1988, p 501
35. J.M. Larson and S. Floreen, *Metall. Trans. A*, Vol 8A, 1977, p 51
36 .Z. Huda in Advanced Materials-93; Ed. A. ul. Haq, A.Q. Khan, 1993, p.391
37. Z. Huda, paper presented at 9th Natl. Conf. on Engineering and Fracture Mechanics (Karachi, Pakistan), 28 April 1993
38. Z. Huda and B. Ralph, *Mater. Charac.*, Vol 25, 1990, p 210
39. B. Ralph, K.B. Shim, Z. Huda, J. Furley, and M. Edirisnghe, *Mater. Sci. Forum*, Vol 94-96, 1992
40. K.E. Volk and A.W. Franklin, *Z.Mettalkd*, Vol 51, 1960, p.172