A R C H I V E S O F M E T A L L U R G Y A N D M A T E R I A L S
Volume 54 2009 Issue 3

A. ONYSZKO*, K. KUBIAK*

METHOD FOR PRODUCTION OF SINGLE CRYSTAL SUPERALLOYS TURBINE BLADES

METODA WYTWARZANIA MONOKRYSTALICZNYCH ŁOPATEK TURBIN Z NADSTOPÓW NIKLU

The superalloys are materials displaying excellent resistance to mechanical and chemical degradation particularly at temperatures beyond 800 °C. Single crystal superalloys are used nowadays more and more commonly as blade materials for aircraft turbines. Their metallurgical development is linked inextricably to the history of the jet engine, for which the first superalloy grades were designed, although use has now spread to other high-temperature applications – notably to the turbines used for electricity generation. The technological incentive for further improvements being sought in alloy capability and processing developments arises from the enhanced fuel efficiency and reduced emissions expected of modern turbine engines; consistent with thermodynamic consideration, higher operating temperatures are then required. The risk of failure by fracture or fatigue, so that necessitates that components fabricated from the superalloys are of the highest integrity, so that inspection and lifetime estimation procedures are critical to safe ongoing operation. A survey of the creep performance of available engineering materials confirm the superiority of the superalloys for high-temperature applications. The use of nickel as the solvent for these materials can be justified on account of its face-centered cubic (FCC) crystal structure, its moderate cost and low rates of thermally activated processes.

Nadstopy niklu są materiałami odznaczającymi się wysoką żarowytrzymałością i żaroodpornością w temperaturze przekraczającej 800 °C. Monokrystaliczne nadstopy niklu używane są do produkcji łopatek turbin silników lotniczych. Ich rozwój metalurgiczny jest nierozłącznie połączony z historią silników lotniczych dla których to skonstruowano pierwsze łopatki z nadstopów niklu. Ponadto na przestrzeni lat łopatek zaczęto używać do innych wysokotemperaturowych zastosowań - szczególnie w turbinach prądotwórczych. Zwiększona wydajność paliw oraz potrzeba zmniejszenia emisji, jakiej oczekuje się po nowoczesnych silnikach, stanowią dodatkową zachętę do dalszego ulepszania technologii jak i procesu wytwarzania łopatek z nadstopów niklu. Ryzyko uszkodzenia poprzez ułamanie lub zmęczenie materiału wymaga, aby części wyprodukowane z nadstopów niklu były najwyższej jakości. Badania wytrzymałościowe takie jak pomiar prędkości pełzania i odkształcania potwierdzają przewagę nadstopów niklu jako materiałów pracujących w wysokich temperaturach. Użycie niklu w tych materiałach może być uzasadnione jego strukturą krystaliczną, przystępnymi kosztami produkcji, a także wolnym tempem aktywacji cieplnej.

Laboratorium Badań Materiałów dla Przemysłu Lotniczego Politechniki Rzeszowskiej jest jednostką produkującą monokrystaliczne nadstopy niklu i kobaltu. W pracy przedstawiono zagadnienia związane z wytwarzaniem z nadstopów niklu łopatek turbin silników lotniczych a także perspektywy rozwoju tego przedsięwzięcia.

1. Introduction

The superalloy CMSX-6 was used to produce the single crystal turbine blades, becouse it is first generation superalloy who has good properties of molten metals. First generation Ni-based single crystal superalloys are mainly Cr, Co, Mo, W, Al, Ti, Ta and sometimes Nb or V. Cr, Co and Mo partition preferentially to the austenitic face-centred cubic nickel-based γ matrix where they act mainly as solution strengthening elements. Cr also plays an essential role in the hot corrosion resistancesince it promotes the formation of a protective Cr_2O_3 oxide scale. These alloys contain a high volume fraction of strengthening ordered Ni_3Al-based γ' phase particles homogeneously distributed in the matrix as near-cubical precipitates [1].

Several alloy designers showed that a significant improvement of the creep strength of the second generation single crystal superalloys may be obtained by the addition of rhenium at the expense of other refractory elements such as Mo or W. Thus, a study carried out on modified Mar-M200 SC alloys showed that additions

* DEPARTMENT OF MATERIALS SCIENCE, RZESZÓW UNIVERSITY OF TECHNOLOGY, 35-959 RZESZÓW, 2 W. POLA STR., POLAND

of rhenium substantially lower the γ' coarsening kinetics and result in large negative $\gamma - \gamma'$ misfits. Atom-probe studies performed on CMSX-2 and PWA-1480 alloys modified by additions of rhenium also showed the existence of Re atom clusters within the γ matrix of these alloys, which is a more potent source of strengthening than the conventional solid solution effect [2].

More recently, alloy designers tried to improve again the high temperature capability of the SC blade alloys by increasing the content of rhenium up to about 6 wt.%. The challenge was to achieve improved creep strength, without increasing the density and by keeping the alloy not too much prone to the precipitation TCP phases. Two typical third generation alloys are CMSX-10 developed by Cannon-Muskegon and René N6 developed by General Electric. More recent development work conducted by GE was devoted to third generation SC alloys containing also some additions of ruthenium. A new generation of SC alloys, a typical example of which is the MC-NG alloy, is developed in France by ONERA and the alloys TMS75 and TMS80 were developed in Japan [3,4].

Most of nickel based alloys contain 10-20% Cr, up to 8% Al and Ti, 5-10% Co, and small amounts of B, Zr, and C. Other common additions are Mo, W, Ta, Hf, and Nb. The major phases present in most nickel superalloys are as follows gamma (γ) and gamma prime (γ'). The continuous matrix (called gamma) is an face-centered-cubic (FCC) nickel-based austenitic phase that usually contains a high percentage of solid-solution elements such as Co, Cr, Mo, and W. The primary strengthening phase in nickel-based superalloys is Ni_3 (Al,Ti), and is called gamma prime (γ'). It is a coherently precipitating phase with an ordered $L1_2$ (fcc) crystal structure. The close match in matrix lattice parameter (~0-1%) combined with the chemical compatibility allows the γ' to precipitate homogeneously throughout the matrix and have long-time stability. Interestingly, the flow stress of the γ' increases with increasing temperature up to about 650°C. In addition, γ' is quite ductile and thus imparts strength to the matrix without lowering the fracture toughness of the alloy. Aluminum and titanium are the major constituents and are added in amounts and mutual proportions to precipitate a high volume fraction in the matrix. In some modern alloys the volume fraction of the γ' precipitate is around 70%. There are many factors that contribute to the hardening imparted by the γ' and include γ' fault energy, γ' strength, coherency strains, volume fraction of γ', and γ' particle size [4,5].

The three main advantages of single crystal over the conventionally cast and directionally solidified components are [6]:

a) Elimination of grain boundaries made strengthening elements, such as carbon and hafnium redundant. This has facilitated heat treatment and allowed for the further optimization of the alloy chemistry to increase of the high temperature capability,
b) Elimination of grain boundaries transverse to the principal tensile stress axis has reduced grain boundary cavitations and cracking, resulting in greatly enhanced creep ductility,
c) The preferred <001> crystallographic solidification direction, which coincides with the minimum in Young's modulus and is oriented parallel to the component axis minimizes the thermal stresses developed on engine start-up and shut-down, this has dramatically improved the thermal fatigue resistance of the turbine hot gas path components.

2. Experimental procedure

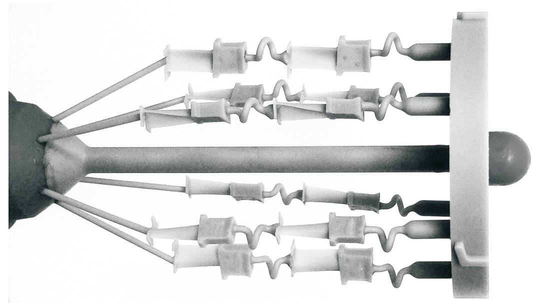

Fig. 1. Wax models of turbine blades

Figure 1 is schematic illustration of the casting furnace used for the production of single crystal turbine blades. An important recent contribution has come from the alignment of the alloy grain in the single crystal blade, which has allowed the elastic properties of the material to be controlled more closely. These properties in turn control the natural vibration frequencies of the blade.

Fig. 2. Ceramic modul for production the single crystal turbine baldes

In Research and Development Laboratory for Aerospace Materials is used Vacuum Induction Melting furnace ALD for investment casting of equiaxed superalloy materials, directionally solidified and single crystal materials for making aircraft engine turbine blades (Fig 2). This furnace has a conventional Bridgman crystal-growing method. A speed of a few inches per hour is typical - so that the solid/liquid interface progresses gradually along the casting, beginning at its base. This has the effect of producing large, columnar grains which are elongated in the direction of withdrawal, so that transverse grain boundaries are absent [10]. Prior to casting, the chamber is evacuated to a partial pressure of approximately 10^{-3} bar and the moulds raised into the furnace chamber, which is pre-heated to above the melting point of the charge using graphite resistance elements. The whole assembly is then allowed to equilibrate for a number of hours prior to casting.

Fig. 3. Vacuum Induction Melting furnace ALD for investment casting of single crystal materials

The first step in the production of single crystal turbine blade is to make its wax model (Fig 3). A wax model of the casting is prepared by injecting molten wax into a metallic mould – if necessary by allowing wax to set around a ceramic core, which is a replica of the cooling passages required. These are arranged in clusters connected by wax replicas of runners and risers; this enables several blades to be produced in a single casting. Wax replicas consisting of a feed arrangement, platen, starter, selector, and test bar sections were assembled. The test bars were rectilinear (of dimensions $12 \times 12 \times 165$ mm^3) and these were mounted on double helical grain selectors (6 mm diameter with a 30° pitch, total height of 45 mm) which were in turn placed above cylindrical starter blocks (25 mm height and 20 mm diameter). A single casting consisted of eight such test bars arranged on a circular carousel. This was bottom fed via a central riser which was connected to the starter blocks by a series of horizontal runners [11].

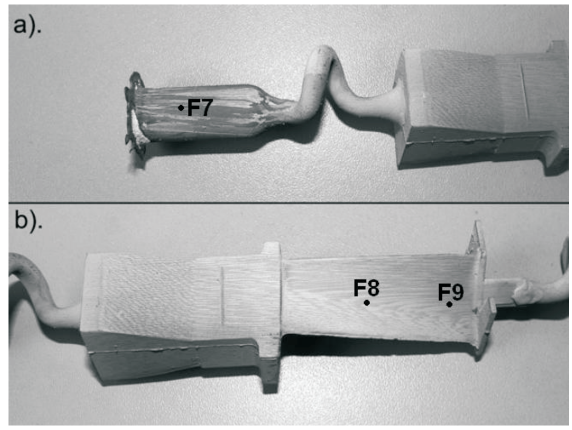

Fig. 4. a). grain selector – pig-tail-shaped spiral, b). turbine blade after removal of the investment shell

Next, an investment shell is produced by dipping the model into ceramic slurries consisting of binding agents and mixtures of zircon silicate ($ZrSiO_4$), alumina (Al_2O_3) and silica (SiO_2), followed by stuccoing with larger particles of these same materials (Fig 4). This operation is usually repeated three or four times until the shell thickness is adequate. Finally, the mould is baked to build its strength. The first step involves a temperature just sufficient to melt out the wax – usually a steam autoclave is used. Further steps at higher temperatures are employed to fire the ceramic mould. After preheating and degassing, the mould is ready to receive the molten superalloy (Tab 1), which is poured under vacuum at a temperature of ~1550 °. After solidification is complete, the investment shell is removed and the internal ceramic core leached out by chemical means, using a high-pressure autoclave. It is clear from this description that many steps are required. Fortunately, in most modern foundries considerable amounts of automation have been introduced [8-9].

TABLE 1

Nominal composition of alloy CMSX-6 (%mass.)

Element	wt. %
Ni	Balance
Cr	10.0
Co	5.0
Mo	3.0
Al	4.8
Ti	4.7
Ta	6.0
Hf	0.1

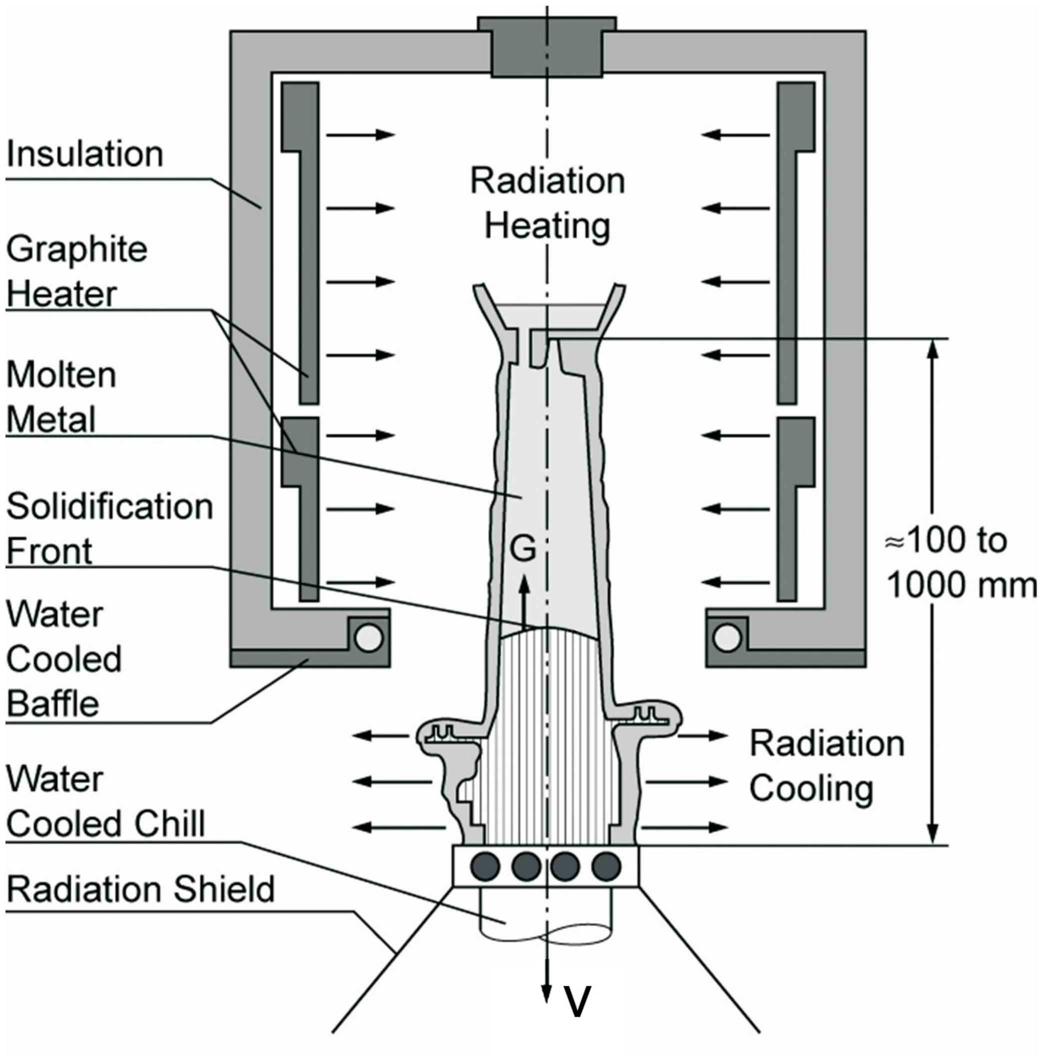

Fig. 5. Schematic ilustration of Vacuum Induction Melting Furnace ALD. G – temperature gradient, v – velocity of draw out

Most typically, this is achieved by adding a "grain selector" to the very base of the wax mould, typically in the form of a pig-tail-shaped spiral (Fig 5a). Since this is not significantly larger in cross-section than the grain size, only a single grain enters the cavity of the casting, which is then monocrystalline form. Alternatively, a seed can be introduced at the base of the casting; provided the processing conditions are chosen such that this is not entirely, growth occurs with an orientation consistent with that of the seed. Figure 5b shows turbine blade after removal of the investment shell, with pig-tail grain selector still in place [12-15].

3. Experimental results and their discussion

The removal of the grain boundaries using a grain selector, or else by seeding, has ramifractions for the crystallographic orientation of the casting. The preferred grown direction for nickel and its alloys, in common with all known FCC alloys, is <001>. Thus turbine blade aerofoils have a <001> direction aligned along, or close to, the axis of the casting. The dendrites grow at a rate which is largely controlled by solute diffusion, since the solid phase grows from the liquid with a very different composition from it, thus the local dendrite tip undercooling, scales monotonically with the velocity (v) of the dendrite, measured along the temperature gradient (G). It follows that dendrites which are misaligned by an angle Θ with respect to perfectly aligned ones must grow at a greater undercooling and hence at the rear of the grown front (Fig. 6).

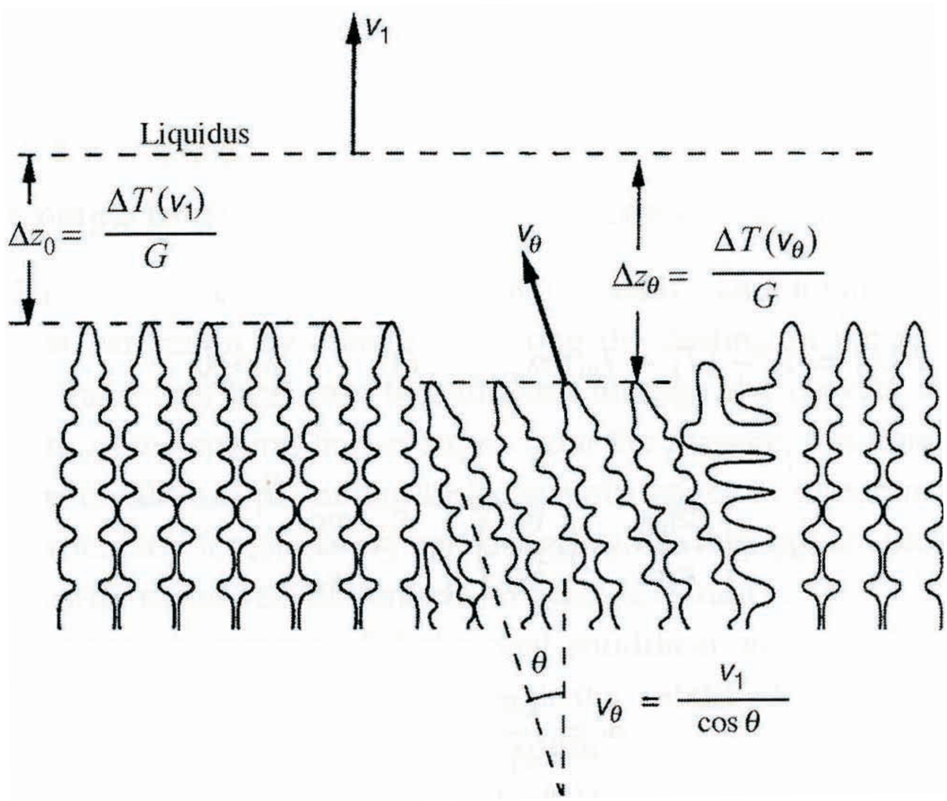

Fig. 6. Illustration of the competitive grain growth process, in which misaligned dendrites are suppressed by the secondary arms of well aligned ones, R¿C¿ Reed¿ The Superalloys. Cambridge 2006, page 133

The perform's microstructure observed on the metallographic microscope is presented in Figure 7 – 9. In Figure 7 the microstructure of CMSX-6 single crystal turbine blade is different than the microstructure in Figure 9. This is the result of the grain selector use, where only a single grain enters the cavity of the casting, which is then in monocrystalline form.

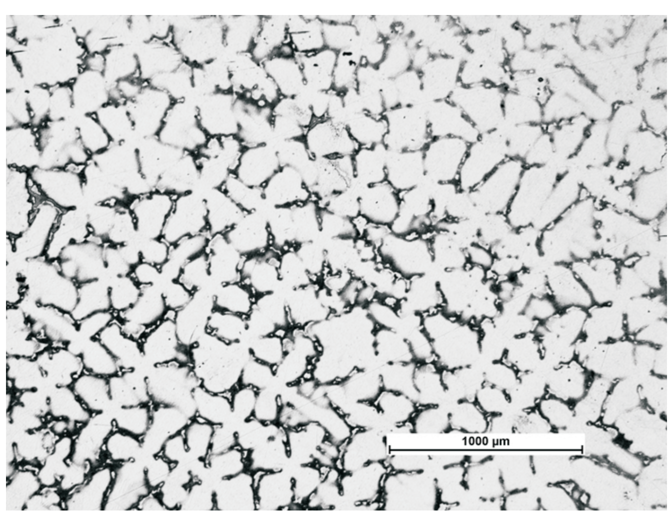

Fig. 7. The microstructure of CMSX-6 single crystal turbine blade (center area of the turbine blade)

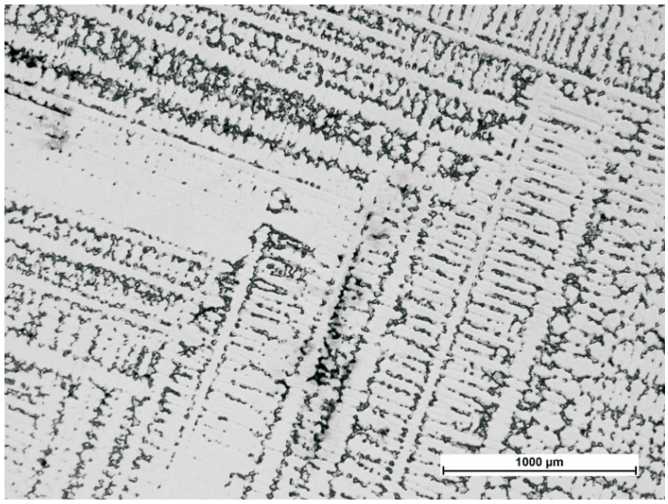

Fig. 8. The microstructure of CMSX-6 single crystal turbine blade (last area of the turbine blade)

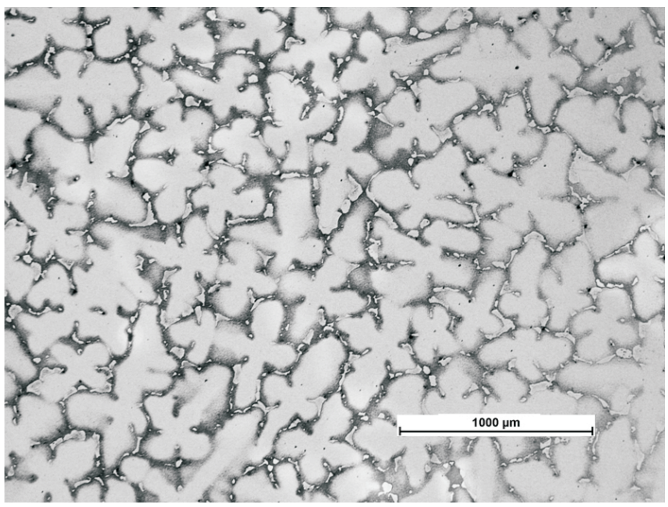

Fig. 9. The microstructure of CMSX-6 single crystal turbine blade (area before grain selector of the turbine blade)

Microstructure in the F7 area (polycrystalline zone in front of the selector) is characterized by disordered dendrite distribution that is feature for polycrystalline material. Monocrystal F8 area exhibits symmetrical distribution of dendrites within all directions in space. The arms of monocrystal dendrites from the F9 area are much narrower than arms formed in the F8 area that is in accordance with the theory of monocrystal solidification.

4. Conclusions

The attention was paid in the work to the critical stages of the process of the production of ceramic forms and single crystal turbine blades with CMSX-6. At present in Research and Development Laboratory for Aerospace Materials in Rzeszów University of Technology are led experimental works with settlement of the correct conditions of the process of the production of the forms and casting single crystal turbine blades with alloys CMSX-4 and PWA 1426.

Future work will examine crystallographic and metallographic response to various solidification parameters of nickel-based superalloys of II and III generation. It is also planned production and detailed analysis of monocrystal blades with internal cooling circuits of various configurations.

REFERENCES

[1] W. Betteridge, S. W. S. Shaw, Development of superalloys, Materials Science and Technology **3**, 682-694 (1987).

[2] R. C. Reed, The Superalloys, Fundamentals and Application, Cambridge University Press, Cambridge, 2006.

[3] J. C. Williams, E. A. Strake, Progress in struc-

tural materials for aerospace systems, Acta Materialia **51**, 5775-5799 (2003).
[4] R. Scharfrik, R. Sprague, The saga of gas turbine materials, Advanced Materials and Processes, 162 (2004) 3:33-35.
[5] G. A. Whittaker, Precision casting of aero gas turbine components, Materials Science and Technology **2**, 436-441 (1986).
[6] J. D. Livingston, H. E. Cline, E. F. Koch, R. R. Russell, High speed solidification of several eutectic alloys, Acta Metallurgica **18**, 399-403 (1970).
[7] A. D. Cetel, D. N. Duhl, Second generation nickel-base single crystal superalloy, in Reichman S., Duhl D.N., Maurer G., Antolovich S., Lund C, eds, Superalloys 1988, Warrendale, PA: The Metallurgical Society, 1988, 236-243.
[8] F. R. N. Nabarro, H. L. de Villiers, The Physics of Creep, Taylor and Francis, 1995.
[9] S. Suresh, Fatigue of Materials, 2nd edn, Cambridge University Press, Cambridge, 1998.
[10] T. I. Barry, A. T. Dinsdale, Thermodynamics of metal gas liquid reactions, Materials Science and Technolody **3**, 503-510 (1987).
[11] P. Carter, D. C. Cox, C. A. Gandin, R. C. Reed, Process modelling of grain selection during the solidification of single crystal superalloy castings, Materials Science and Engineering **A280**, 233-246 (2000).
[12] A. G. Dodd, Mechanical design of gas turbine blanding in cast superalloys, Materials Science and Technology **2**, 476-480 (1986).
[13] L. A. Jackman, G. A. Maurer, S. Widge, New knowledge about white spots in superalloys, Journal of Metals **43**, 19-25 (1993).
[14] D. Furrer, H. Fecht, Ni-based superalloys for turbine discs, Journal of Metals **51**, 15-18 (1999).
[15] E. Volterra, J. H. Gaines, Advanced Strength of Materials, Engelwood Cliffs, NJ: Prentice-Hall, 1971.

Received: 10 January 2009.