Hot strength of creep resistant ferritic steels and relationship to creep rupture data

R. C. Dimitriu and H. K. D. H. Bhadeshia*

Experimental data on the tensile strength of ferritic steels designed for prolonged service at elevated temperatures have been assessed as a function of many variables, including the testing temperature. The resulting model has been combined with other data on the intrinsic strength of pure ferritic iron and substitutional solute strengthening to show that there is a regime in the temperature range 780–845 K beyond which there is a rapid decline in the microstructural contribution to strength. This decline cannot be attributed to changes in microstructure, but possibly to the ability of dislocations to overcome obstacles with the help of thermal activation. There is evidence of an approximate relationship between the temperature dependence of hot tensile strength and creep rupture stress.

Keywords: Hot strength, Ferritic steel, Creep rupture, Power plant, Energy

Introduction

There have been many studies in which the hot strength of austenite has been modelled, primarily as an aid to the simulation of the hot rolling process or that of the bending of steel during the continuous casting process.[1–5] There do not appear to be similar studies for creep resistant ferritic steels, where hot strength is a parameter in the design of power plant components. In recent work, the creep rupture life of such steels was factorised using non-linear methods into a variety of components including those due to precipitates, solid solution strengthening and pure iron.[6] One result is illustrated in Fig. 1, which shows that the role of precipitates becomes smaller as the test temperature is increased. The original purpose of the present work was to make a similar assessment for the hot tensile strength but certain trends that emerged from the analysis proved interesting in the context of creep, so a comparison is also made against creep data.

Method

A general method for treating complex data is the neural network in a Bayesian framework. This has been documented thoroughly[7–10] and applied extensively in the study and design of steels.[11–18] For this reason, only specific points of relevance are introduced here.

The network is a non-linear regression method which, because of its flexibility, is able to capture enormous complexity in the data, while at the same time avoiding overfitting. There are a number of interesting outputs other than the coefficients which help recognise the significance of each input. First, there is the noise in the output, associated with the fact the input set is unlikely to be comprehensive, i.e. a different result is obtained from identical experiments. Second, there is the uncertainty of modelling because many mathematical functions may be able to adequately indicate known data but which behave differently when extrapolated. A knowledge of this uncertainty helps make the method less dangerous in extrapolation.

Variables

The analysis is based on published data[19] on the hot strength (0·2% proof strength) of ferritic creep resistant steels including the classical 2·25Cr–Mo, 5%Cr, 9Cr–Mo and 12Cr–Mo type steels. The hot strength is a function of the microstructure and solid solution strengthening, both of which depend on chemical composition and heat treatment: the relevant variables are listed in Table 1. The measures taken to avoid overfitting, the training procedures and the use of optimised committees of models have been described elsewhere.[7–11,13]

Figure 2 shows the reasonable agreement between the calculated and measured values of hot strength. The error bars indicate $\pm 1\sigma$ modelling uncertainties; the noise in the output was estimated at $\pm 3\%$.

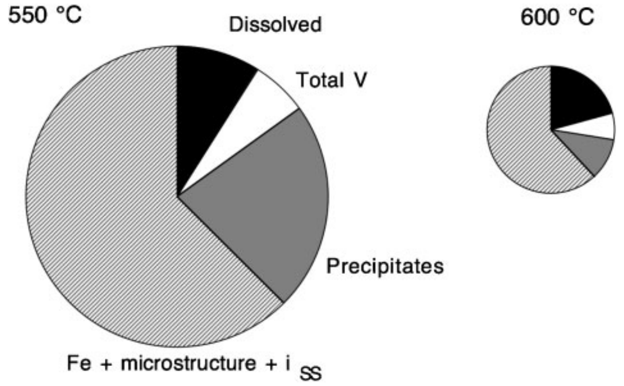
Interpretation of hot strength

Three examples are considered here, using steels in the 2·25Cr–Mo, 5%Cr and 9Cr–Mo categories. In each case, phase diagrams were calculated using MTDATA[20] and the 'solution plus' thermodynamic database. The phases allowed in the calculation were cementite, M_3C_2, M_7C_3, $M_{23}C_6$, M_6C ('M' stands for metal atoms) and ferrite, including Fe, C, Si, Mn, Ni, Cr, Mo, Cu, Al and N as the components. The composition used to estimate the solid solution strengthening of ferrite was that calculated for the tempering temperature, since this is higher than the tensile test temperature and because the test itself is

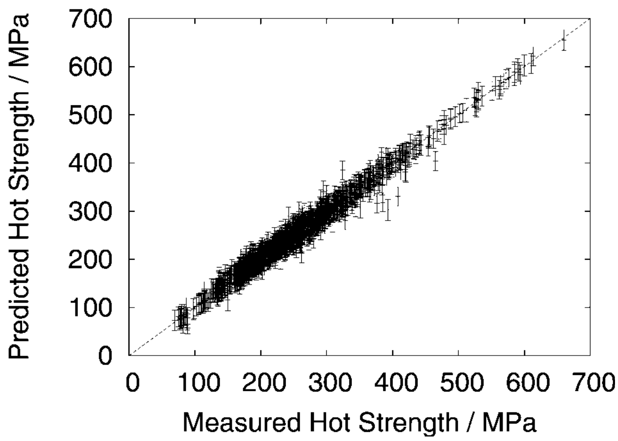
Department of Materials Science and Metallurgy, University of Cambridge, Pembroke Street, Cambridge CB2 3QZ, UK

*Corresponding author, email hkdb@cam.ac.uk

Published by Maney on behalf of the Institute
Received 24 April 2007; accepted 26 May 2007
DOI 10.1179/174328407X213332



1 Pie charts showing factorisation of 10^5 h creep strength of 2·25Cr–Mo steel: diameters of pie charts are scaled to reflect 10^5 h strength at appropriate temperature; terms i_{SS} indicates contributions due to dissolved solutes other than Mo and V (Ref. 6)



2 Comparison of predicted and measured hot strength

of a short duration (the hot tensile tests are reported to have been carried out to Japanese standard JIS G0567: after the specified temperature is reached, the sample is held there for 15 min). Given the concentrations of substitutional solutes in the ferrite, their contributions were estimated as a function of test temperature using the model due to.[21] The small concentrations of interstitials in equilibrium with carbides were neglected because they are likely to be located at defects[22–25] and hence do not contribute to solid solution effects. The strength of pure iron with a coarse microstructure, as a function of temperature, is from Leslie.[26] The microstructural contribution from carbides and the tempered martensite or bainite plates, is then the difference between the neural network estimate and the SS+Fe curve.

The results are presented in Fig. 3*a*, *c* and *e* for the three steels. The first observation is that hot strength can be categorised into two regimes. The first is an almost linear decrease in strength with temperature approximately over the range 200–700 K. This occurs at a rate consistent with the loss of strengthening due to substitutional solutes and iron (cf. SS+Fe curve). The second part of the hot strength is also approximately linear over the range 800–950 K, but the decrease in strength with temperature is much more dramatic. The transition occurs for all the steels at approximately T_T=780–845 K (510–570°C). By extrapolating the low and high temperature behaviour, T_T is found to be 793, 845 and 780 K for the 2·5Cr–Mo, 5%Cr and 9Cr–Mo steels respectively.

Table 1 Variables used to develop model

Variable	Minimum	Maximum
Aluminium, wt-%	0·001	0·04
Carbon, wt-%	0·09	0·48
Copper, wt-%	0·0001	0·25
Chromium, wt-%	0·0001	12·38
Manganese, wt-%	0·38	1·44
Molybdenum, wt-%	0·01	1·05
Nickel, wt-%	0·0001	0·6
Nitrogen, wt-%	0·001	0·04
Silicon, wt-%	0·18	0·86
Austenitising time, min	10	5400
Tempering time, min	30	660
Austenitising temperature, K	1143·15	1243·15
Tempering temperature, K	898·15	1023·15
Test temperature, K	293·15	973·15
Hot strength, MPa	69	660

The microstructural component of strength is maintained below T_T but decreases dramatically above that temperature. Because this happens in short duration hot tensile tests, there should exist a similar phenomenon in creep rupture testing.

The accelerated decrease in hot strength when $T>T_T$ cannot be attributed to coarsening phenomena or microstructural changes, because in all the cases illustrated, the samples have been tempered at temperatures in excess of 990 K, which is much higher than the tensile test temperatures. Neither can it be associated with any similar behaviour in the solute strengthening or the strength of pure iron, both of which are almost monotonic straight lines.

The phase diagrams (Fig. 3*b*, *d* and *f*) show that there is no dramatic or consistent change in equilibrium phase fractions at T_T. The remaining possibility is that it becomes easier for dislocations to overcome obstacles by a thermally activated mechanism beyond T_T.

Relevance to creep rupture data

Figure 4*a–c* shows creep rupture data (10^3, 10^4 and 10^5 h)[19] plotted on the same graphs as the hot strength data. It is striking that the temperature sensitivity of the rupture stress is similar to that of the proof strength for $T>T_T$. Unfortunately, low temperature creep data are not available to make a similar comparison for $T<T_T$, but Fig. 4*d* shows that the allowable stress in creep design varies in a manner strikingly similar to the behaviour of hot strength as a function of temperature.

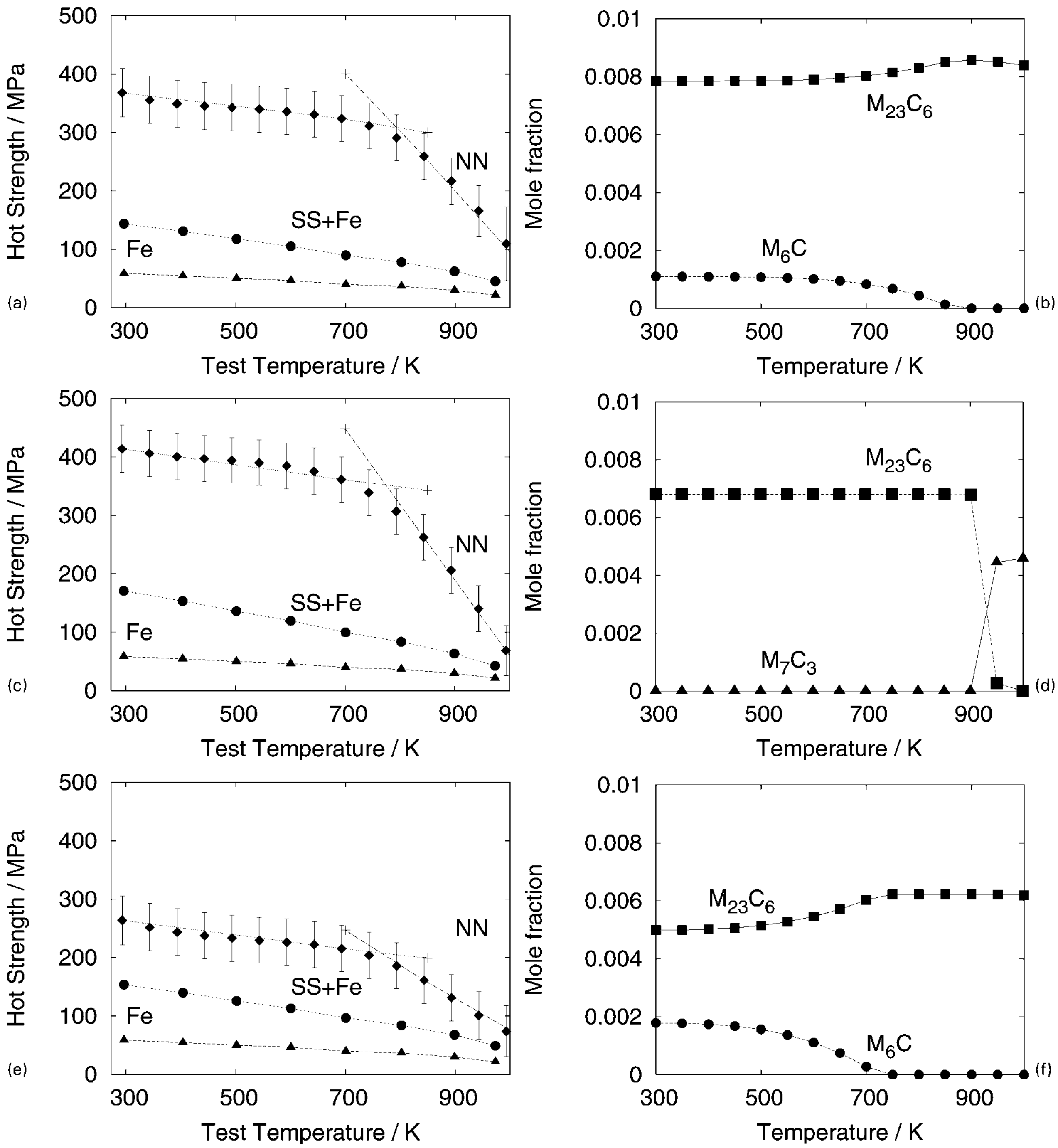
Following Dorn[27] and Weertman,[28] the steady state creep rate $\dot{\varepsilon}$ at low stress may be written as

$$\dot{\varepsilon}=A\sigma^{n}\exp\left(-\frac{Q}{kT}\right) \tag{1}$$

where σ is the stress, T is the absolute temperature, A and n are constants, Q is an activation energy and k is the Boltzmann's constant. Alternatively, the flow stress corresponding to a given creep rate may be written as

$$\sigma=\left(\frac{\dot{\varepsilon}}{A}\right)^{1/n}\exp\left(\frac{Q}{nkT}\right) \tag{2}$$

Assuming that the creep failure time t_f is related to the creep rate via $n\dot{\varepsilon}t_f\cong 1$, it follows that the failure stress σ_f



a 2·5%Cr steel: 0·15C–0·18Si–0·63Mn–0·024Ni–2·23Cr–0·97Mo–0·2Cu–0·01Al–0·0083N, 1193 K for 8 h, air cooled, tempered at 993 K for 6 h; *b* 2·5%Cr steel; *c* 5%Cr steel: 0·12C–0·33Si–0·56Mn–0·046Ni–5Cr–0·049Mo–0·05Cu–0·066Al–0·017N, 1173 K for 10 min, air cooled, tempered at 1023 K for 120 min; *d* 5%Cr; *e* 9Cr–Mo steel: 0·11C–0·59Si–0·41Mn–0·1Ni–9·15Cr–1·05Mo–0·02Cu–0·011Al–0·018N, 1133 K for 30 min, air cooled, tempered at 1033 K for 90 min; *f* 9Cr–Mo

3 Hot strength and equilibrium phase diagrams

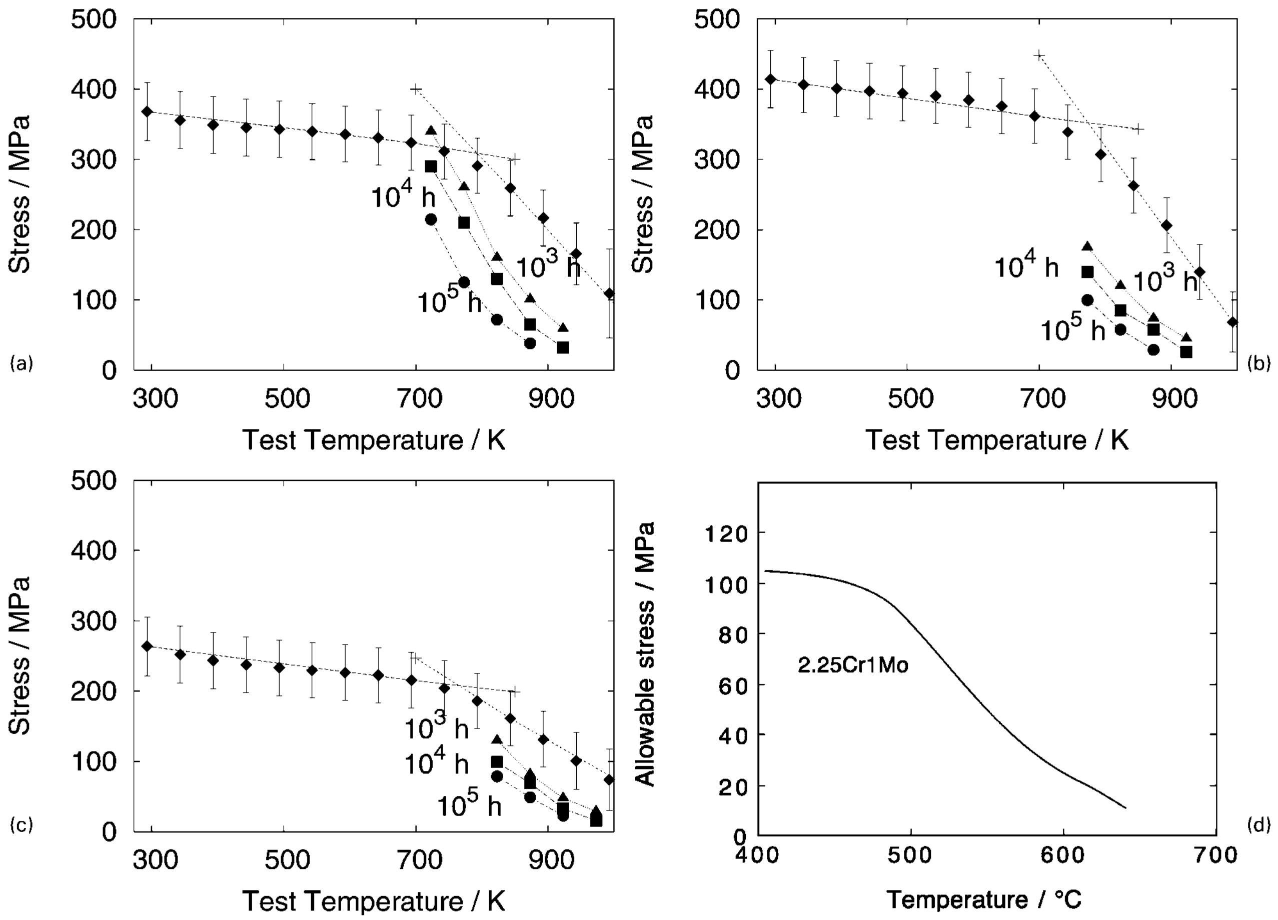
is given by

$$\sigma_f \cong \left(\frac{1}{Ant_f}\right)^{1/n} \exp\left(\frac{Q}{nkT}\right) \tag{3}$$

If it is assumed that equation (3) can now be used to indicate a hot tensile test by setting t_f to a sufficiently small value, then the activation energy for a hot tensile test becomes equivalent to that for creep.

This analysis using creep models assumes that the mechanism operating in the hot tensile test is identical to that in a creep test for $T>T_T$. However, we have not using these equations justified the relationship between a hot tensile test and a creep test.

But the overwhelming observation from the neural network model is clear that there exists a clear relationship between creep rupture data and hot tensile strength for $T>T_T$. This suggests that hot strength tests could in research programmes be used as rough indicators of the temperature sensitivity of creep rupture data. This may not be too far fetched if for $T>T_T$, the mechanism remains thermally activated dislocation motion for both tensile and creep deformation.



a 2·5Cr–Mo; *b* 5%Cr; *c* 9Cr–Mo; *d* allowable stress for 2·5Cr–Mo steel[29]

4 Comparison of temperature sensitivity of creep rupture and proof strength

Summary

A neural network model has been developed to enable the estimation of the 0·2% proof strength of creep resistant ferritic steels as a function of chemical composition and heat treatment parameters. The model has been combined with other observations to show that there is a regime in the range 780–840 K beyond which there is a steep decline in the microstructural contribution to strength. This decline cannot be attributed to changes in microstructure, but perhaps to an increased ability of dislocations to overcome obstacles with the help of thermal activation.

The temperature sensitivity of hot tensile tests at high temperatures seems to be replicated in creep rupture data. This observation could be exploited in research and development programmes.

The computer program developed can be downloaded freely from www.msm.cq.uk/map/mapmain.html.

Acknowledgement

R. C. Dimitriu would like to thank the Marie Curie Early Stage Research Training Programme, European Commission.

References

1. V. Narayan, R. Abad, B. Lopez, H. K. D. H. Bhadeshia and D. J. C. MacKay: *ISIJ Int.*, 1999, **39**, 999–1005.
2. P. D. Hodgson, L. X. Kong and C. H. J. Davies: *J. Mater. Process. Technol.*, 1999, **87**, 131–138.
3. L. X. Kong and P. D. Hodgson: *Adv. Eng. Softw.*, 2000, **31**, 945–954.
4. L. X. Kong, P. D. Hodgson and D. C. Collinson: *J. Mater. Process. Technol.*, 2000, **102**, 84–89.
5. Z. Sterjovski, D. Nolan, K. Carpenter, D. Dunne and J. Norrish: *J. Mater. Process. Technol.*, 2005, **170**, 536–544.
6. M. Murugananth and H. K. D. H. Bhadeshia: in 'Mathematical modelling of weld phenomena 6', (ed. H. Cerjak and H. K. D. H. Bhadeshia), 243–260; 2002, London, Maney.
7. D. J. C. MacKay: *Neural Comput.*, 1992, **4**, 448–472.
8. D. J. C. MacKay: *Neural Comput.*, 1992, **4**, 415–447.
9. H. K. D. H. Bhadeshia: *ISIJ Int.*, 1999, **39**, 966–979.
10. D. J. C. MacKay: 'Information theory, inference, and learning algorithms'; 2003, Cambridge, Cambridge University Press.
11. H. K. D. H. Bhadeshia, D. J. C. MacKay and L.-E. Svensson: *Mater. Sci. Technol.*, 1995, **11**, 1046–1051.
12. T. Goswami: *ISIJ Int.*, 1996, **36**, 354–360.
13. S. B. Singh and H. K. D. H. Bhadeshia: *Mater. Sci. Eng. A*, 1998, **A245**, 72–79.
14. J. M. Vitek, Y. S. Iskander and E. M. Oblow: *Weld. J.*, 2000, **79**, 33s–50s.
15. J. Tenner, D. A. Linken, P. F. Morris and T. J. Bailey: *Ironmaking Steelmaking*, 2001, **28**, 15–22.
16. D. Dunne, H. Tsuei and Z. Sterjovski: *ISIJ Int.*, 2004, **44**, 1599–1607.
17. Z. Guo and W. Sha: *Comput. Mater. Sci.*, 2004, **29**, 12–28.
18. M. Mukherjee, S. B. Singh and O. N. Mohanty: *Mater. Sci. Eng. A*, 2006, **A434**, 237–245.
19. Y. Kojchi, I. Hiroshi, T. Hideo, Y. Masayoshi, K. Osamu, K. Kiyoshi and K. Kazuhio: 'NRIM data sheets 1b, 3b, 8b, 11b, 12b, 17b, 18b, 19b, 20b and 21b'; 1994, Tokyo, National Research Institute for Metals.
20. NPL: 'MTDATA: software'; 2006, Teddington, National Physical Laboratory.
21. A. A. B. Sugden and H. K. D. H. Bhadeshia: *Metall. Trans. A*, 1988, **19A**, 1597–1602.

22. D. Kalish, S. A. Kulin and M. Cohen: *J. Met.*, 1965, **17**, 157–164.
23. H. K. D. H. Bhadeshia and A. R. Waugh: *Acta Metall.*, 1982, **30**, 775–784.
24. M. Peet, S. S. Babu, M. K. Miller and H. K. D. H. Bhadeshia: *Scr. Mater.*, 2004, **50**, 1277–1281.
25. F. G. Caballero, M. K. Miller, S. S. Babu and C. Garcia-Mateo: *Acta Mater.*, 2007, **55**, 381–390.
26. W. C. Leslie: *Metall. Trans. A*, 1972, **3A**, 5–23.
27. J. E. Dorn: *J. Mech. Phys. Solids*, 1955, **3**, 85–88.
28. J. Weertman: *J. Appl. Phys.*, 1955, **26**, 1213–1217.
29. N. Komai, F. Masuyama, I. Ishihara, T. Yokoyama, Y. Yamadera, H. Okada, K. Miyata and Y. Sawaragi: in 'Advanced heat resistant steels for power generation', 96–109; 1999, London, IOM Communications.