$$\frac{\mathrm{d}a}{\mathrm{d}N} \propto \left(\frac{K}{K_c}\right)^m$$

and that changes in $\mathrm{d}a/\mathrm{d}N$ with frequency are really reflections of changes in K_c. This is not a new concept and, indeed, it has been noted before that changes in K_c are reflected in $\mathrm{d}a/\mathrm{d}N$. It is then proposed that the changes in K_c due to frequency are due to the visco-elasticity of the β transition in most cases (other transitions could be relevant) and that this can be represented approximately by:

$$K_c \propto w^{\tan\Delta}$$

where $\tan\Delta$ is the loss factor appropriate to the transition. This derivation is based on a constant crack opening displacement and assumes a modulus appropriate to the highly strained crack-tip region (hence, the use of a relatively high strain value). The high frequency sensitivity of $\mathrm{d}a/\mathrm{d}N$ is then seen to be the product $-m \tan\Delta$ and reflects the β transition through $\tan\Delta$. In their comments, the authors show that they can measure a small strain modulus which does not reflect the β transition, which is probably correct but the value is not relevant since the fatigue crack growth is influenced by the β transition. It is considered that the argument given here, while not complete, does provide a physical basis for the effect with some quantitative support. The author apologises for the lack of clarity on this point in the original presentation.

Received 11 October
and accepted 23 November 1978

J. G. WILLIAMS
Department of Mechanical Engineering,
Imperial College,
London, UK

The Larson–Miller C constant as applied to a cobalt-based directionally solidified eutectic alloy

The mechanical properties of the eutectic composite C73 [1] have attracted much interest over the past years with particular attention being given to the improvement of the ductility and creep strength of the alloy [2–6].

To carry out experiments within a reasonable time, a method has to be chosen which allows the extrapolation of long-time rupture data from short-time results. From the many methods available [7], the Larson–Miller parameter [8] is commonly used, with

$$(\log F_\sigma) = P_{\mathrm{LM}} = T(C + \log t_{\mathrm{B}})$$

where T is the absolute temperature, t_{B} is the rupture life and C is a constant widely assumed as being 20.

However, the use of this approach with data for directionally solidified high-temperature eutectic composites has been shown to be questionable. Buchanan and Tarshis [9], for example, found that the Larson–Miller parameter was dependent on stress level for the Co–15Cr–13TaC* eutectic alloy.

Recently, Woodford [10], working on the system Ni–13Ta found it necessary to vary the C constant in order to obtain a reliable correlation, the parameter becoming $P = T\,[(2000/\sigma) + \log t]$, where σ denotes the stress.

The purpose of this letter is to show that similar deviations are also to be found with a modified C73 eutectic alloy of composition 56.8% Co, 39% Cr, 2.2% C and 2% Al. A des-

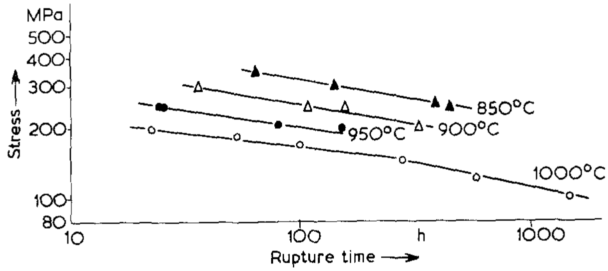

Figure 1 Logarithmic plot of stress versus rupture life for air data.

*wt %.

0022–2461/79/071759–03 $2.30/0

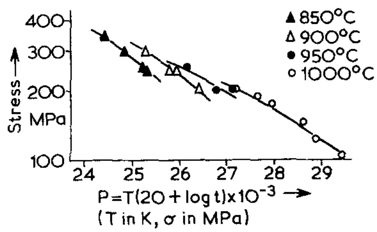

Figure 2 Larson–Miller plot for air showing systematic deviations with increasing time.

cription of this alloy and its relevant properties is given elsewhere [11]. Stress-rupture testing in air was carried out on specimens of 2 mm nominal diameter and 20 mm length. The test temperatures were 1000, 950, 900 and 850° C and the loads varied between 100 and 350 MPa. Metallographic examination of the broken specimens showed no fibre cracking indicating the same mode of failure over the entire stress range.

The results are shown in Table I and are plotted in a conventional log/log form in Fig. 1. Fig. 2 shows a Larson–Miller plot for all the data points. It is clear from this that $C = 20$ is not the optimal value. Figs. 3 and 4 show the effect of varying C from 5 to 30. For $C = 5$, parallel curves are obtained which are temperature-dependent. As the Larson–Miller form of representation is supposed to remove temperature-dependence, the curves have no meaning. With $C = 30$, the spacing between the curves becomes larger, and they become concave.

TABLE I Summary of rupture data

Temperature (° C)	Stress (MPa)	Time to rupture (h)
850	350	65, 6
	300	143, 2
	248	442
	251	391
900	300	37
	250	109, 160
	245	190
	200	337
950	250	25, 26
	205	80, 5
	200	150, 3
1000	200	23
	185	54
	170	102
	145	283
	120	586
	100	1469

Fig. 4 shows the curves for C between 10 and 15, which indicate that the constant for this modified alloy lies between 12 and 13. In order to test this graphical result, a multiple linear regressional analysis was carried out. This was adopted from [12].

From

$$\log t_B = \log a + bT(C + \log t)$$

it follows that

$$\log t_B = \log a + bCT + bT \log t$$

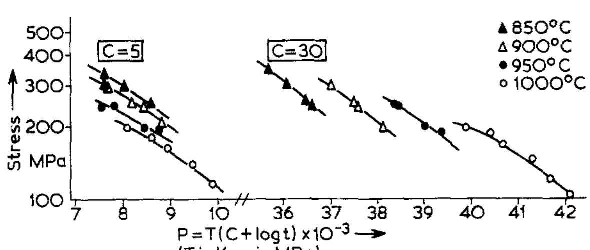

Figure 3 Larson–Miller plot for $C = 5$ and $C = 30$.

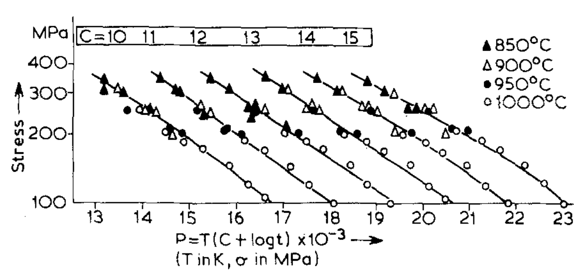

Figure 4 Larson–Miller plot for C varied between 10 and 15.

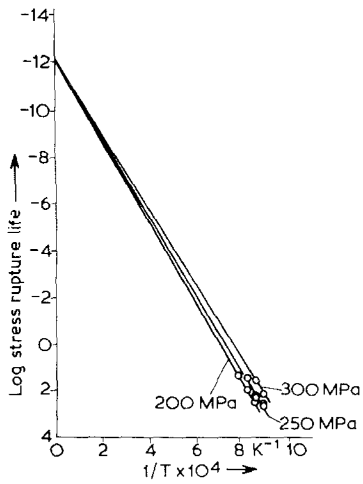

Figure 5 Log time–inverse temperature plot.

with

$$bC = c \quad \text{and} \quad T \log t = z$$

and one obtains

$$\log t_B = \log a + cT + bz.$$

The constant $\log a$, is found from the regression analysis as well as the partial coefficients c and b, where the Larson–Miller constant C is given by

$$C = \frac{c}{b}.$$

Using the values given in Table I, a C value of 12.3 was obtained using a Wang 2200 Series Program which correlates extremely well with the graphical result.

The constant C can also be determined using a technique proposed by Larson and Miller [8], namely a log time–inverse temperature plot. This method has been used by Buchanan and Tarshis [9]. Fig. 5 shows such a plot which, due to the lack of data, can only poorly indicate that C lies between 12 and 13. An alternative method proposed by Woodford [10] of solving simultaneous equations to give a constant parameter value at stresses where results at two or more test temperatures are available, gives variable values for C.

In summary, this study shows that the C constant has to be adjusted to the material in question and that in this case a modification of $C \sim 12$ reasonably correlates the data.

Acknowledgement

Part of this work was supported by the German Ministry of Research and Technology (BMFT).

References

1. E.R. THOMPSON and F.D. LEMKEY, *Met. Trans.* **1** (1970) 2799.
2. E.R. THOMPSON, D.A. KOSS and J.C. CHESNUTT, *ibid* **1** (1970) 2807.
3. A.R. NICOLL and P.R. SAHM, Proceedings of the International Conference on Composite Materials, Vol. 1 (1975) pp. 904–917.
4. K. FRITSCHER and G. WIRTH, *ibid*, Vol. 2 pp. 597–603.
5. R.F. VANDERMONSEN *et al.*, *Cobalt* **2** (1973) 33.
6. P.R. SAHM, *High Temp. High Press.* **7** (1975) 241.
7. D.A. WOODFORD, *Mat. Sci. Eng.* **15** (1974) 169.
8. F.R. LARSON and J. MILLER, *Trans. AIME* **6** (1975) 766.
9. E.R. BUCHANAN and L.A. TARSHIS, *Met. Trans.* **5** (1974) 1413.
10. D.A. WOODFORD, *ibid* **8A** (1977) 639.
11. P.R. SAHM and U.W. HILDEBRANDT, BMFT-NTS 37 Final Report (1978).
12. G. ZWINGMAN and K.F. ZIMMERMAN, *Metall.* **31** (1977) 746.

Received 18 October and accepted 20 November 1978

A. R. NICOLL
Brown Boveri & Cie AG Research Centre, Heidelberg, West Germany

Hot-pressing diagrams for f c c metals

Ashby [1] has constructed sintering diagrams (log (a/R) versus T/T_m) which identify, at a given particle size R, neck size a, and reduced temperature T/T_m, the dominant mechanism. As various mechanisms contribute simultaneously to the neck growth, and as the different rates are functions of a/R, the boundaries of these fields are obtained by equating pairs of rate equations and solving for neck size as a function of temperature. The aim of this communication is to extend such diagrams to hot-pressing.

Let us first return to sintering diagrams. The

0022–2461/79/071761–05 $2.50/0