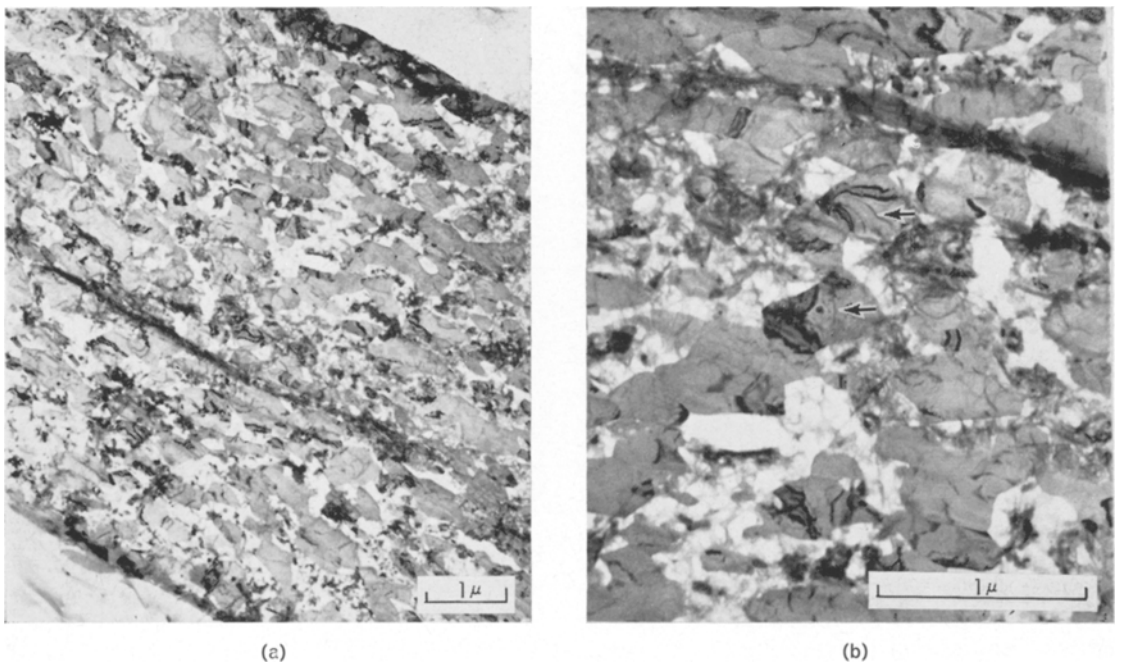

Figure 3 Al–78 wt. % Zn, quenched and cold-rolled (75%). Arrows on (b) show typical thickness contours.

Fig. 3 is an electron micrograph of a sample which was cold-rolled (75%) after quenching. An elongation of the grains in the rolling direction can be observed; however, no dependence of grain elongation on the percentage of rolling is seen. Since even the sample which had only been quenched (fig. 2) shows this grain distortion, it was attributed to a pre-annealing treatment.

The sharp contrast lines, especially noticeable in the zinc-rich phase, are caused by diffraction effects due to bending of the foils and local thickness variations. Some folding of the sections parallel to the knife edge can also be observed in the electron micrographs.

While ultramicrotomy has been used successfully in the present study to reveal the distribution of phases, the same technique would not prove as valuable in obtaining information on the actual deformation mechanism in this alloy, since the cutting operation introduces unavoidable strains in the foils.

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Comments on the Letter "A Method for Determining the Thermal and Athermal Components of Flow Stress from Stress Relaxation" by P. Rodriguez (J. Materials Sci. **3** *(1968) 98)*

Rodriguez [1] has recently suggested that the athermal component of the flow stress can be obtained from stress-relaxation in the following manner. A sample is deformed, and at a given strain, the crosshead movement is stopped and load relaxation is observed. The load is then removed in small decrements until the relaxation rate is too small to be detected. The stress at which no relaxation rate is observed was designated the *athermal stress*, σ_μ. The purpose of this communication is to point out that the Rodriguez method can lead to incorrect values of the athermal stress.

The most common method of obtaining the athermal stress for a bcc metal or alloy is to observe the temperature dependence of the flow

stress and to designate the temperature-independent portion observed at high temperature as the athermal stress, τ_μ. Another method of obtaining the athermal stress is Li's [2] analysis of the dislocation velocity-stress relation originally suggested by Johnston [3]:

$$v = B(\tau - \tau_i)^{m*} \quad (1)$$

where v is the dislocation velocity, B is the dislocation mobility at unit shear stress, τ is the applied shear stress, τ_i is the long range internal stress, and m^* is the dislocation velocity exponent. Assuming a constant mobile dislocation density, ρ, and a constant τ_i during stress-relaxation, Li [2] obtains:

$$\tau - \tau_i = k(t + a)^{-n} \quad (2)$$

where $n = 1/(m^* - 1)$, t is the time, a is an integration constant and k is a constant depending on mobile dislocation density, elastic modulus, and other material constants. One can differentiate and rearrange equation 2 such that a linear relationship between $\log(d\tau/dt)$ versus $\log(t + a)$ is predicted with a slope of $(-n - 1)$. If $t \gg a$, then $\log(d\tau/dt)$ will be linear with $\log t$ but will deviate at short times by an amount equal to a. Such behaviour for Nb–6 at % W single crystals is shown in fig. 1. Knowing n and a, τ_i can be obtained directly by utilising equation 2 at two different times during a single relaxation. Also knowing n, the value of m^* can be obtained. Gupta [4] has applied the Li analysis to many materials and observes all the materials to obey equation 2 with values of m^* comparable with those values obtained independently by etch pitting. Koss [5] has also found good agreement between Li's analysis and room temperature relaxation of pure Nb and Nb–W single crystals. Furthermore, the long range internal stress obtained by Li's analysis and the athermal stress obtained by the temperature dependence of the flow stress method should be equal, barring any metastable obstacles or change in thermally activated mechanism. Koss [5] has shown that both methods give similar results for the magnitude of the athermal stress for pure Nb and four Nb–W alloys (1, 3, 6 and 15 at. %).

Consider now the Rodriguez method. The stress at which no stress-relaxation could be observed is very much a function of the apparatus. The author, avoiding thermal fluctuation by working only at room temperature with an Instrom machine, could barely detect relaxation

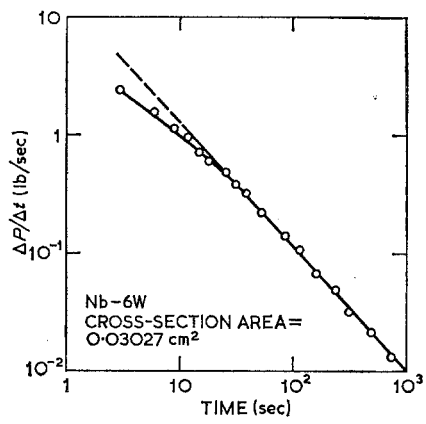

Figure 1 The dependence of load relaxation on time and for a Nb–6 at. % W single crystal. $m^* = 23$; $n = 0.45$; $a = 2.5$ sec.

of 180 g in a 20 min period or 0.15 g/sec. The implications of a relaxation rate of less than 0.15 g/sec will now be examined. For example, if we consider the relaxation rate of Nb–6 at. % W (see fig. 1), the relaxation rate of 0.15 g/sec can be attained in 9.3 h by extrapolation of data in fig. 1. The shear stress at this time would be about 9.5 kg/mm², which according to Rodriguez should be the athermal stress. This extrapolated value agrees well with that value obtained if one applies the Rodriguez technique to Nb–6 at. % W single crystals (i.e. $\tau_\mu = 8.6$ to 10.5 kg/mm). However, the room temperature athermal stress as obtained by Li's method and by the temperature dependence of the flow stress (τ–T) is shown below.

TABLE I

Method	$\tau^{\gamma=0.02}$ for Nb–6 at. % W
τ at 0.15 g/sec	~ 9.5 kg/mm²
Rodriguez	8.6–10.2
Li	3.3–4.5
τ–T*	4.4–5.0

*Corrected for temperature-dependence of shear modulus

As shown above, there is considerable discrepancy between the Rodriguez technique and the other two methods. Since the temperature-dependence of the flow stress method is generally accepted, and there is ample evidence for the validity of Li's method [2–5], one must conclude that the Rodriguez method results in erroneous values of τ_i. Examination of Li's analysis also shows that the magnitude of the error depends on

n (and hence m^*) and a in equation 2. This means that even comparative values of Rodriguez's "τ_i" are misleading unless one is certain that a and especially m^* are constant. Koss [5] has shown that m^* can vary by a factor of about four, between pure Nb and Nb–15 at.% W. Also Guberman [6] finds that m^* is sensitive to interstitial impurity content, while in certain cases, Gupta [4] has shown m^* to be a function of strain as well. We may conclude that the Rodriguez technique of determining τ_i is valid only as a comparative method in the very special case where m^* is constant at points of comparison.

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Book Reviews

High Pressure Methods in Solid State Research

C. C. Bradley

Pp 176 (Butterworths, 1968) 60*s*

On the flyleaf of this book it is stated that over 300 labs are engaged in making high pressure measurements. It adds that consequently it is fitting that an up-to-date publication which concerns itself with the practical problems associated with high pressure generation is now available. It is also claimed that the volume will be of great value not only to high pressure researchers but also technologists and engineers working in this field.

There is no doubt that the interest in high pressures has grown rapidly in the last few decades. The effect of pressure on the properties of matter has steadily increased since the early work of Perkins (*Trans. Roy. Soc.* (1819-1820) 324) on the compressibility of water at pressure up to 2 kbar. The classical researches of the late Professor P. W. Bridgman carried out from 1909 up until a decade ago greatly accelerated this interest. Commercially the discovery of polythene and its commercial exploitation thirty years ago was the beginning of high pressure engineering. Perhaps the synthesis of diamonds achieved in 1955 was one of the most important developments of high pressure research. There is also a growing industrial interest in high pressure isostatic compaction and hydrostatic extrusion of metals.

One of the regions of great interest is in the pressure range below about 30 kbar where fluid transmission of pressure can be employed. In this region conventional cylinder construction can be employed. Consequently it is a disappointment to the reviewer to read what can only fairly be described as an inadequate treatment of the strength and design of a simple thick-walled cylinder, which is probably still the most widely used vessel in this pressure range. There is also no adequate treatment of the design of more conventional compound vessels, or fluid support designs, or sectored vessel design. What is perhaps even more regrettable is that the bibliography at the end of Chapter 3 is totally inadequate and it does not do justice to the published work in this country or elsewhere.

Similarly the treatment of the selection of materials for construction in Chapter 2 is dangerously inadequate. The problem of fatigue of vessels is dismissed in a paragraph including a statement that "it is not relevant enough to cause a great deal of concern". Fatigue is of importance from the first cycle onwards and there is a considerable amount of data on both long-term and short-term fatigue of vessels yet no references are given. Again the whole problem of fracture toughness in relation to the failure of