17. R. J. JACCODINE, *Appl. Phys. Letters* **4** (1964) 114.
18. E. LEVINE, J. WASHBURN, and G. THOMAS, *J. Appl. Phys.* **38** (1967) 87.
19. H. J. QUEISSER, K. HUBNER, and W. SHOCKLEY, *Phys. Rev.* **123** (1961) 1245.
20. F. KARSTENSEN, *J. Electronics and Control* **3** (1957) 305.
21. H. WIDMER, *Phys. Rev.* **125** (1962) 30.
22. L. A. HELDT and J. N. HOBSTETTER, *Acta. Met.* **11** (1963) 1165.
23. C. D. CALHOUN and L. A. HELDT, *ibid* **13** (1965) 932.
24. R. F. PEART, *Phys. Stat. Solidi* **15** (1966) K119.
25. R. N. GHOSTAGORE, *Phys. Rev. Letters* **16** (1966) 890.
26. J. M. FAIRFIELD and B. J. MASTERS, *Bull. Amer. Phys. Soc.* July 1966, and *Appl. Phys. Letters* **8** (1966) 280.
27. E. TANNENBAUM, *Solid State Electronics* **2** (1961) 123.
28. M. O. THURSTON and J. TSAI, Ohio State University Foundation Report No. 1233-49 (1962).
29. I. M. MACKINTOSH, *J. Electrochem. Soc.* **109** (1962) 392.
30. E. L. WILLIAMS, *ibid* **108** (1961) 795.
31. L. E. MILLER, "Properties of Elemental and Compound Semiconductors", edited by H. Gatos (Interscience, New York, 1960).
32. Y. SATO and H. ARATA, *J. Appl. Phys. Japan* **3** (1964) 511.
33. P. BARUCH, C. CONSTANTIN, J. C. PFISTER, and R. SAINTSPRIT, *Discuss. Faraday Soc.* **31** (1961) 76.
34. G. E. MOORE, "Microelectronics", edited by E. Keonjian (McGraw-Hill, New York, 1963) p. 262.
35. R. GERETH, P. G. G. VAN LOON, and W. WILLIAMS, *J. Electrochem. Soc.* **112** (1965) 323.
36. K. H. NICHOLAS, *Solid State Electronics* **9** (1966) 35.
37. H. J. QUEISSER and P. G. G. VAN LOON, *J. Appl. Phys.* **35** (1964) 3066.
38. R. GERETH and G. H. SCHWUTTKE, *Appl. Phys. Letters* **8** (1966) 55.
39. P. SCHMIDT and R. STICKLER, *J. Electrochem. Soc.* **111** (1964) 1188.
40. J. E. LAWRENCE, *J. Appl. Phys.* **37** (1966) 4106.

Letter

A Method for Determining the Thermal and Athermal Components of Flow Stress from Stress-Relaxation

It is now generally accepted that plastic deformation of metals can be thermally activated if strain rate is determined by the overcoming of short-range obstacles by dislocations [1, 2]. This behaviour can be satisfactorily accounted for by an Arrhenius type of equation

$$\dot{\gamma} = \dot{\gamma}_0 \exp\left[-\frac{H(\tau^*)}{kT}\right] \qquad (1)$$

where $\dot{\gamma}$ is the shear strain rate; $\dot{\gamma}_0$ contains geometrical factors and the frequency with which a dislocation pressed against an obstacle attempts to overcome it; H is the activation energy; k and T have their usual significance.

It was first suggested by Seeger [3] that the shear stress for yield or flow of a metal crystal, τ, can be considered to consist of two components; one, an athermal component τ_μ, depends on temperature only through the shear modulus μ, and the other, the thermal component τ^*, depends on temperature T and strain rate $\dot{\gamma}$;

$$\tau = \tau^*(T, \dot{\gamma}) + \tau_\mu \qquad (2)$$

The thermodynamic variables associated with thermal activation can be determined by evaluating τ^* at various temperatures and also by determining deformation partials such as $(\partial\tau^*/\partial T)\dot{\gamma}$ and $(\partial\tau^*/\partial \ln\dot{\gamma})_T$ [1, 2]. By evaluating these variables, it is then possible to speculate on the operating short-range hardening mechanism.

The starting point of any such analysis is the separation of τ^* from the total flow stress τ. This is usually accomplished by measuring τ at various temperatures up to a sufficiently high temperature T_0 above which all short-range obstacles are transparent to dislocations (i.e. $\tau^* = 0$). The total applied stress is then applied solely to overcoming the long-range stress-field; τ^* for other temperatures can then be determined by knowing the shear modulus variations with temperature

$$\tau^*_T = \tau_T - \frac{\tau_\mu(T_0).\mu_T}{\mu_{T_0}} \approx \tau_T - \tau_{T_0} \qquad (3)$$

From tensile tests on polycrystals, the tensile flow stress σ^*, is similarly determined and τ^* is taken to be equal to $\frac{1}{2}\sigma^*$.

The above method is reasonably accurate, and has been successfully employed in several investigations (e.g. in studies on the variation of τ^* with temperature in bcc metals). The question arises as to whether this method can be extended indiscriminately for studying thermally activated deformation in metals and alloys containing "metastable" obstacles. The method would obviously fail if the "metastable" structure changes to a stabler structure below T_0 where τ_μ is usually evaluated. A case in point is the effect of irradiation or quenching on the flow stresses of metals. If the defects introduced by irradiation or quenching anneal out at temperatures lower than T_0, it will be difficult to assess the thermal and athermal components of the radiation- or quench-hardening by the method described above. Hence there is a strong case for evaluating τ_μ and τ_0 at each temperature of the experiment. The purpose of this investigation is to suggest a possible method for such an evaluation from stress-relaxation experiments. This technique has been applied to niobium and the results obtained are compared with values determined by the conventional method.

The theory of stress relaxation has been discussed in detail in the literature [4-7], and will therefore be outlined only briefly. When a tensile test is interrupted, stress-relaxation occurs since plastic deformation continues to take place as long as the applied stress is sufficiently high for the dislocations to move. Since the total strain is constant, the plastic strain ϵ_p that occurs during relaxation is matched by an elastic strain ϵ_e i.e.

$$\frac{d\epsilon_p}{dt} = -\frac{d\epsilon_e}{dt} = -\frac{1}{E}\frac{d\sigma}{dt} \qquad (4)$$

where E is the combined elastic modulus of the specimen and the machine. As the plastic strain-rate is governed by the thermal component of stress σ^*, stress-relaxation occurs at a progressively decreasing rate, and will be practically unobservable when σ_μ† is approached, as shown schematically in fig. 1.

Up to now, most of the investigations using this technique have been concerned primarily with the determination of activation volume V^*; τ^* values were determined separately from flow stress/temperature curves. An exception is the study of dislocation dynamics in irradiated iron by Ohr *et al* [7]. However, they have considered the strain rate to be dependent on the stress relaxed through a power relationship, rather than by equation 1.

†i.e. the tensile stress corresponding to the shear stress τ_μ.

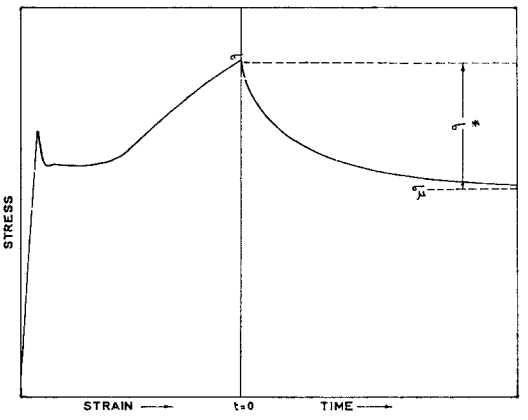

Figure 1 Typical stress/strain and stress/relaxation curves.

The object of the present work was to determine the values of thermal components of flow stress σ^* at various temperatures from yield and flow stress measurements (i.e. by the conventional method) and also from stress-relaxation experiments. This would enable us to compare both these methods, with a view to checking the validity of the stress-relaxation technique. Niobium was chosen for this investigation, since the nature of short-range obstacles for deformation in this metal – namely, the Peierls-Nabarro (P-N) force [8] – is such that τ^* values determined from the conventional method are reasonably accurate, and published data on τ_μ and τ^* values for this metal are available.

Annealed sheet specimens of polycrystalline niobium (grain size ≈ 45 μm, hardness 70 to 74 VPN) were used in the studies. Tensile tests were carried out in a floor model Instron machine at a strain rate of 6.67×10^{-5}/sec at various temperatures ranging from 77 to 448° K, using appropriate temperature baths. The stress/strain curves obtained are shown in fig. 2. Another set of tensile tests were carried out in which the cross-head movement was stopped at various strain levels during the tensile tests, and the stress was allowed to relax. Above 339° K, because of small σ^* values, the relaxation rate became too slow to be detected on the Instron machine in about 2 to 10 min. The stress

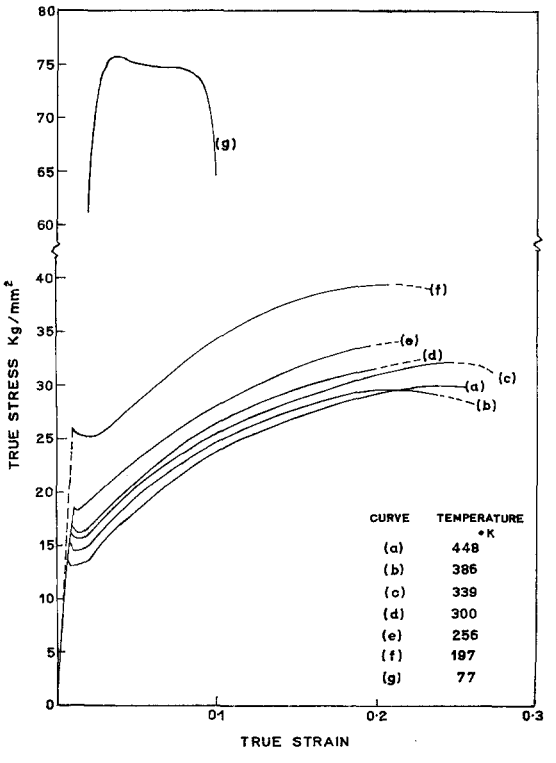

Figure 2 Flow curves of niobium at various temperatures; strain rate = 6.67×10^{-5}/sec.

at which relaxation rate became insignificant was designated σ_μ and the stress relaxed, $\sigma^* = \sigma - \sigma_\mu$, at various strain levels was determined. At low temperatures, relaxation continued even after 30 min, and the following technique was therefore adopted to determine σ_μ and σ^*. The specimen was unloaded in small decrements (by moving the cross head up) and stress-relaxation followed at each stress level, till a stress was reached at which no stress-relaxation could be observed, as shown schematically in fig. 3. This stress was then taken to be equal to σ_μ.

The values of σ^*, the stress relaxed at various strain levels, at different temperatures are compared in table I with the thermal component of the flow stress, $\sigma_T - \sigma_{448}$, showing good agreement between the two sets of values.

The stress-relaxation technique did not give successful results at 77° K (liquid nitrogen temperature). For one thing, when boiling liquid baths are used to maintain the low temperature, relaxation will be interfered with by even a small change in the liquid level due to evaporation, since this will result in a change in dimensions of the pull rods. Even if this difficulty is overcome by maintaining a set-up which allows a constant liquid level in the bath, the

TABLE I Comparison of σ^* values determined by stress-relaxation and the conventional method. (Stresses, other than yield points, have been converted to true stresses).

Temperature (° K)	True strain	$\sigma_T - \sigma_{448}$ (kg/mm²)	Stress relaxed, σ^* (kg/mm²)
448	0.055	0.00	0.66
	0.110	0.00	0.83
	0.180	0.00	0.57
386	Yield point	1.25	
	0.032	1.20	1.22
	0.087	1.10	1.82
	0.150	0.80	2.12
339	Yield point	2.50	
	0.045	1.80	1.59
	0.097	1.80	2.25
	0.136	1.85	1.75
300	Yield point	3.00	
	0.055	2.80	3.23
	0.091	2.70	2.88
	0.128	2.80	3.05
256	Yield point	5.20	
	0.063	4.70	4.95
	0.094	4.60	4.47
	0.125	4.50	5.53
197	Yield point	12.50	
	0.101	10.70	11.74
	0.128	10.70	12.97
	0.160	10.80	10.82

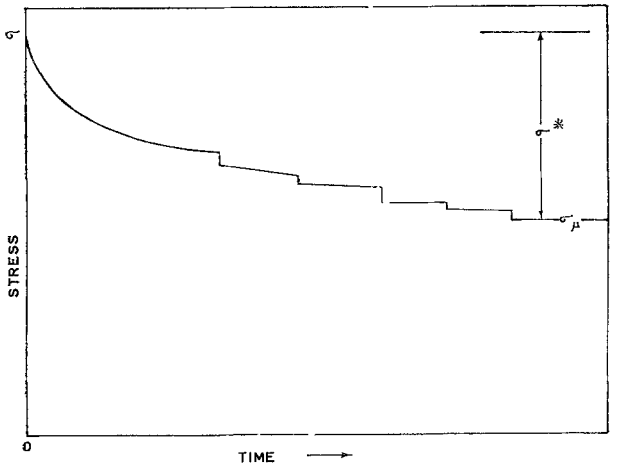

Figure 3 Schematic figure showing technique used for determining σ^*.

relaxation technique will give lower apparent values of σ^* at very low temperatures. This is because what is determined as σ_μ is the stress at which plastic strain rate is so slow that the resultant stress-relaxation is too slight to be detected on the Instron chart. It is estimated that this strain rate will be of the order of 10^{-11}/sec. While at sufficiently high temperatures, the thermal component of stress for this strain rate will be negligible, the error will increase as we go to lower and lower temperatures. In the case of niobium, as results indicate, the method is found to be successful down to as low a temperature as 197° K.

In spite of these limitations at very low temperatures, the method is helpful in evaluating σ^* and σ_μ fairly accurately at not too low temperatures, and can be applied in investigations on metastable structures, where the conventional method fails. σ^* at lower temperatures could then be evaluated by extrapolation to lower temperatures of σ_μ values determined at high temperatures, as in the conventional method. This technique has been successfully employed for studying the deformation behaviour of α-zirconium containing quenched-in hydrogen. The results of this study are being published elsewhere [9].

Acknowledgements

The author is grateful to Dr M. K. Asundi, Head, Physical Metallurgy Section, for his keen interest in this work. He is specially indebted to Dr V. S. Arunachalam for helpful discussions and suggestions.

References

1. H. CONRAD, NPL symposium "The Relation Between the Structure and Mechanical Properties of Metals" (HMSO, 1963) p. 476.
2. *Idem, J. Metals (AIME)* **16** (1964) 582.
3. A. SEEGER in "Dislocations and Mechanical Properties of Crystals", edited by J. Fisher *et al* (Wiley, New York, 1957) p. 243.
4. F. GUIU and P. L. PRATT, *Phys. Stat. Solidi.* **6** (1964) 111.
5. G. A. SARGENT, *Acta Met.* **13** (1965) 663.
6. F. W. NOBLE and D. HULL, *ibid* **12** (1964) 1089.
7. S. M. OHR *et al*, USAEC Report ORNL-3878 (1966) p. 16.
8. G. A. STONE and H. CONRAD, *Acta Met.* **12** (1964) 1125.
9. P. RODRIGUEZ and V. S. ARUNACHALAM, to be submitted to *J. Nucl. Mat.*

7 September 1967

P. RODRIGUEZ
Metallurgy Division
Bhaba Atomic Research Centre
Bombay 1, India

Book Review

Tables of X-ray Mass Attenuation Coefficients

(In German, French, and English)

R. Theisen, D. Vollath

Pp 40 (Verlag Stahleisen, Düsseldorf, 1967) DM 22

This volume gives the determination, by the authors, of the best available data on the values of the mass attenuation coefficients (μ/ρ) for most X-ray lines by most elements. The most reliable values in the literature have been used and interpolated using the formula

$$\mu/\rho = C\,\lambda^\alpha\,Z^B$$

where λ is the wavelength of the radiation and Z the atomic number of the absorbing element. The constants C, α, and B were determined by use of a computer programme.

These data were tabulated with a view to the needs of analysts concerned with X-ray fluorescence, microprobe, and radiographic analyses.

R. D. DOHERTY