Internal friction of alpha titanium

In recent years there have been many investigations into the internal friction behaviour of h c p alpha titanium–oxygen alloys. It is generally believed that oxygen atoms occupy octahedral sites, and as these sites are equivalent, the motion of an oxygen atom to a neighbouring site will not cause a relaxation of the applied stress [1]. Consequently, some workers consider the observed peak to be due to pairs of oxygen atoms [2, 3]. Others, however, maintain that previous investigations have always been conducted on specimens containing small amounts of substitutional metals and that the relaxation peak is due to substitutional–interstitial complexes [4, 5]. Gupta and Wenig [4] could not find a peak at all in pure titanium and suggest that the peak observed by Pratt *et al.* [6] to be due to the presence of iron. Mishra and Asundi [5] attribute the peak measured by Miller and Browne [3] at 420° C (0.8 Hz) in iodide pure titanium to the presence of 0.02 wt% manganese. The peak at 470° C (1.3 Hz) observed by Mishra and Asundi [5] was interpreted by them as being due to the presence of 0.024 wt% iron.

In order to clarify whether interstitial oxygen can cause internal friction on its own, tests have been conducted on titanium containing very low quantities of substitutional elements as impurities. Wire specimens of high purity titanium containing Fe about 30 p.p.m. and Mn less than 50 p.p.m. were prepared. All other substitutional elements (Al, Cu, Zn, Hf) were less than 50 p.p.m. each, with the exception of Zr which was about 100 p.p.m. The specimens were first annealed in a vacuum for 24 h at 850° C and cooled in a furnace down to room temperature, which gave an average grain size of 0.3 mm. They were subsequently oxidized and then reannealed in the internal friction apparatus at 600° C for 12 h. The reannealing procedure was found to reduce the background damping.

Internal friction measurements were carried out between 1 and 2 Hz, using a conventional inverted torsional pendulum in a vacuum better than 10^{-6} torr. The decay was recorded by a Sefram light-spot-follower. The oxygen concentration was analysed after the tests had been performed by a Balzer EOA oxygen analyser.

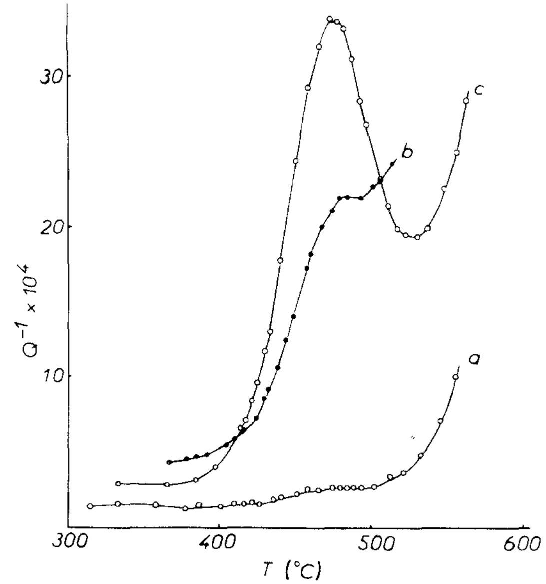

Figure 1 Internal friction of titanium. (a) Unoxidized, (b) 0.9 at.% oxygen, peak temperature 480° C (1.74 Hz), (c) 2.3 at.% oxygen, peak temperature 470° C (1.0 Hz).

In the unoxidized titanium no discernable peak was observed, as shown by curve (a) in Fig. 1. When the oxygen content was raised to 0.9 at.%, a peak emerged at 480° C (1.74 Hz), shown by curve (b) in Fig. 1. Further increase in the oxygen concentration to 2.3 at.% increased the peak height considerably (curve (c) in Fig. 1). The peak heights, after subtracting the background, were 5×10^{-4} and 24.5×10^{-4} respectively. The activation energy associated with the larger peak, calculated from its width at half maximum height, was found to be 49.6 kcal mol^{-1}.

The presence of a peak shows that oxygen on its own can give rise to internal friction. This view is supported by the observation that the peak heights, within experimental error, are proportional to the square of the respective oxygen concentrations. More tests must be conducted at higher oxygen concentrations to confirm this.

Acknowledgements

The author wishes to thank Dr D. J. Bacon for helpful discussions and the Science Research Council (Great Britain) for the provision of a research fellowship. This work was carried out in the Department of Metallurgy and Materials Science, University of Liverpool, England.

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Received 26 July and accepted 19 September 1977.

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Forming limit diagrams and damage in mild steel

Forming limit diagrams (FLD) have been empirically determined [1, 2] to describe the strain states of sheet metal. They are generally plotted in deformation space as ϵ_1 versus ϵ_2, these being the principal strains in the plane of the sheet. The FLDs at necking are of particular interest for determining the moment at which localized deformation appears. They have been studied for several metals and especially for mild steels.

Marciniak has proposed a theoretical analysis of the FLDs at necking [3]. In this analysis, the sheet is characterized by the rheological parameters n and R, the work-hardening exponent and anisotropy ratio respectively, and by an initial "geometrical" defect f_0.

In order to explain the whole FLD from uniaxial tension to biaxial stretching, it has been proposed to choose different parameters n [4] and R, depending on the strain path ρ ($\rho = \epsilon_2/\epsilon_1$); recent studies have confirmed this ρ dependence [5–7].

An alternative approach is to consider that the damage [8] increases continuously with deformation and in a different manner according to the strain path. It is then necessary to introduce an evolutive geometrical defect, f, whose amplitude depends upon the stress or strain tensor. This damage could be due to the existence of slip lines on the surface but this hypothesis is not realistic [9].

TABLE II Rheological parameters of the two SOLLAC steels [12].

Steel	n	R	Steel	n	R
Steel A	0.204	1.70	Steel B	0.208	1.83

The aim of this work is to show that the damage, which controls the necking, can result from the formation of voids at the matrix–particle interface. It has been known for a long time that this damage is responsible for ductile fracture, as proved by the examination of fracture surfaces [8]; but little work has been carried out on the growth of damage during deformation [10, 11] and the effect of different strain paths has never been studied.

The present work has been conducted on two SOLLAC steels of different compositions (Table I) but with similar rheological parameters (Table II). In spite of this rheological similarity their FLDs at necking, in the range where the strain ϵ_2 is positive, are quite different for biaxial stretching (Fig. 1) [12]. This study was carried out along the strain path where $\rho = 1$, which gives the maximum gap between the two FLDs.

TABLE I Chemical composition of the two SOLLAC steels [12].

Steel	Composition (wt % × 10)													
	C	Mn	P	S	N (total)	Al (metal)	Al (oxide)	Al (total)	Si	Cr	Cu	Ni	Mo	Ti
Steel A	51	325	15	13	7.9	42	5	47	58	–	–	–	–	–
Steel B	6	383	8	13	9.4	46	2	48	34	40	67	83	16	213