supports the hypothesis that it is a precursor of the T_g-process. It is interesting to note that we could show the existence of this intermediate process also by measuring the notched impact strength of PC as a function of temperature [22]. The intensity of the impact strength maximum near 370 K diminished rapidly on annealing (the annealing embrittlement effect) which was also consistent with the hypothesis of the T_g-precursor nature of the intermediate process.

On the basis of these preliminary results we conclude that the HPB method is a valuable tool in relaxation studies of polymers. Therefore studies of different construction plastics using this method are in progress in our laboratory.

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Cyclic plastic behaviour of metastable Fe–30 wt % Ni alloy at 293 and 373 K

Until recently the fatigue behaviour of metastable austenite alloys has been reported by only a few investigators [1–3]. The cyclic hardening/softening behaviour is not well established as these alloys transform to martensite during cyclic deformation and it seems necessary to separate the hardening due to the martensite formation from the hardening/softening behaviour due to the multiplication and rearrangement of dislocations. In this note an attempt to separate these phenomena of a metastable austentic steel consisting of 0.005 wt % C, 0.003 wt % Si, 0.002 wt % Mn, 0.003 wt % P, 0.004 wt % S and 30.1 wt % Ni at 293 and 373 K is reported. Fatigue tests were carried out in fully reversed bending using a Schenck type fatigue machine of 450 cpm frequency. A temperature increase in the specimen was not detected under the experimental conditions used when the plastic strain amplitude did not exceed 3.5×10^{-3}.

The variations of plastic strain amplitude were measured dynamically by the mechanical method shown in Fig. 1. The total strain amplitude (Equation 1) can be calculated from the relative displacement a of the specimen measured by a differential transformer. Equation 2 gives the plastic strain amplitude.

$$\epsilon_t = \frac{2e}{l}\tan\theta\left(\theta = \frac{a}{2r}\right) \qquad (1)$$

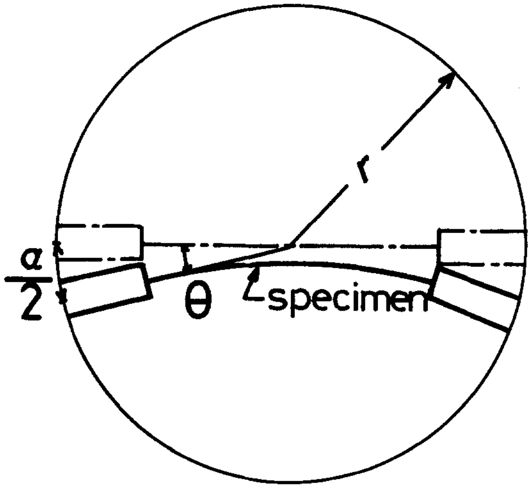

Figure 1 Measurement of strain amplitudes.

$$\epsilon_p = \epsilon_t - \epsilon_e \qquad (2)$$

where, ϵ_e, ϵ_p and ϵ_t are the elastic, plastic and total strain amplitudes respectively, $2e$ is the specimen thickness, l is the specimen length and θ is the deflection angle. X-ray diffraction was used to provide a quantitative estimate of the martensite formed in this alloy during fatigue cycling [4]. This alloy tended to form surface martensite easily, and about 6 to 7% surface martensite was detected prior to cyclic deformation.

Figs. 2 and 3 show the variations of plastic strain amplitude and induced martensite at 293 and 373 K, respectively. It is apparent that this alloy tends to harden by cyclic stress at both temperatures. The interrelationship between the rate of hardening and the rate of α'-martensite formation is presented in Figs. 4 and 5. If the hardening is due to the martensite formation only, these plots would give an approximate straight line. The line in Fig. 4 was obtained by the extrapolation of the plots after 2000 cycles when dislocation rearrangement was presumably stabilized. We call "the hardening meant by this line" HDM (hardening due to martensite formation). The deviation from this line (upward deviation being softening and downward deviation being hardening) presumably suggests that factors other than martensite formation affect the cyclic plastic behaviour. As a straight line could not be obtained at 373 K, the line at 293 K was drawn for convenience and reference in Fig. 5.

When high stress amplitudes (30 or 25 kg mm^{-2}) were applied to this alloy at 293 K, it hardened as a result of other factors in addition to

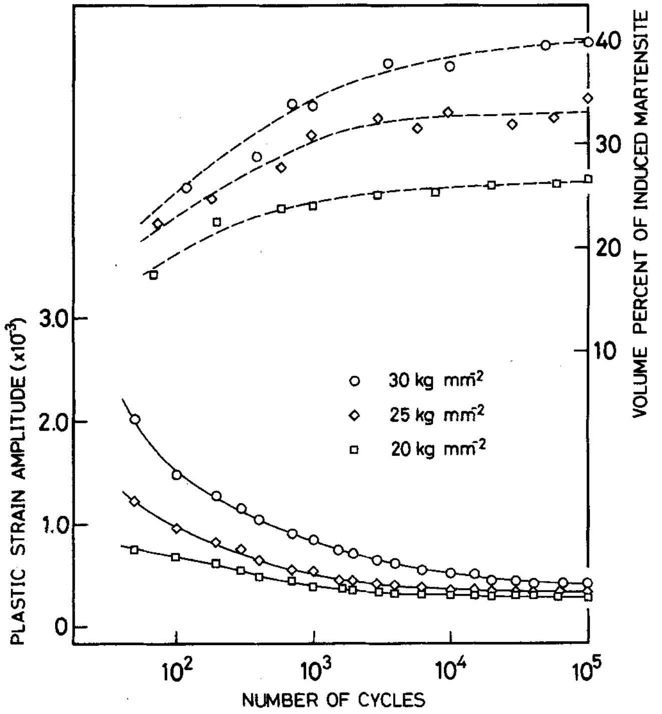

Figure 2 Variations of plastic strain amplitude and the volume fraction of martensite during testing at 293 K.

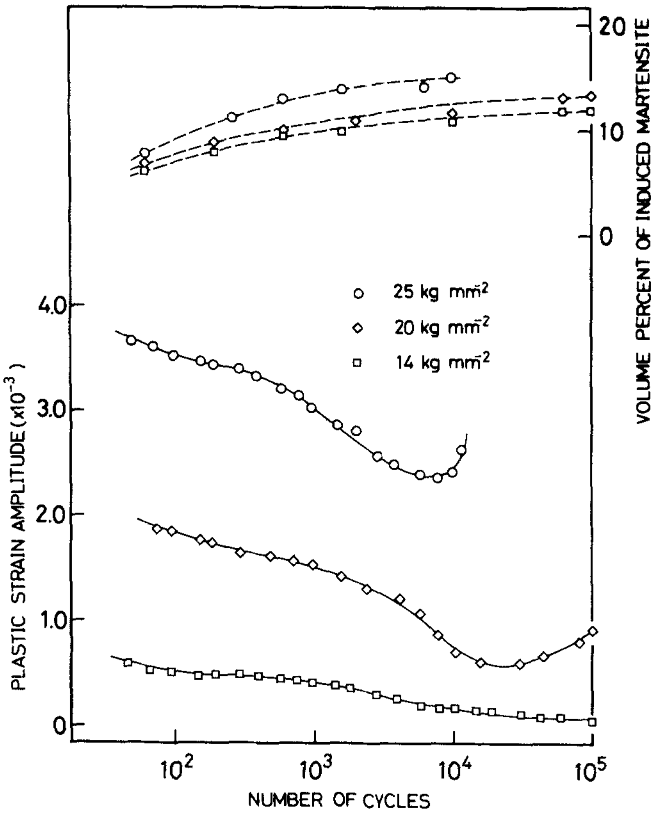

Figure 3 Variations of plastic strain amplitude and the volume fraction of martensite during testing at 373 K.

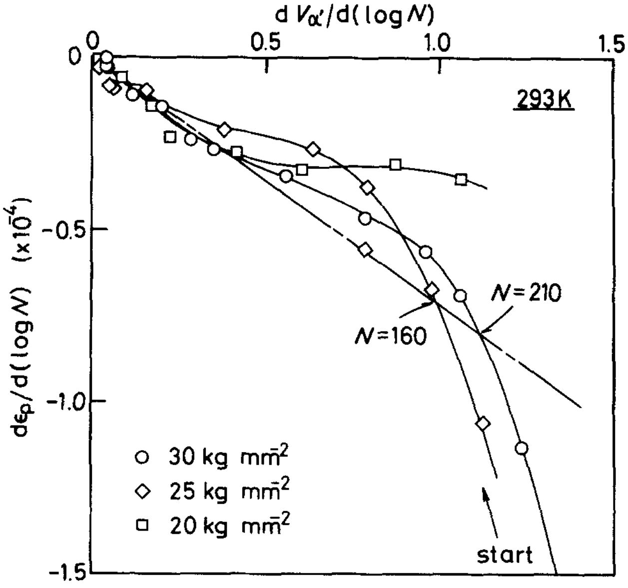

Figure 4 Relationship between the rate of hardening and the rate of martensite formation at 293 K.

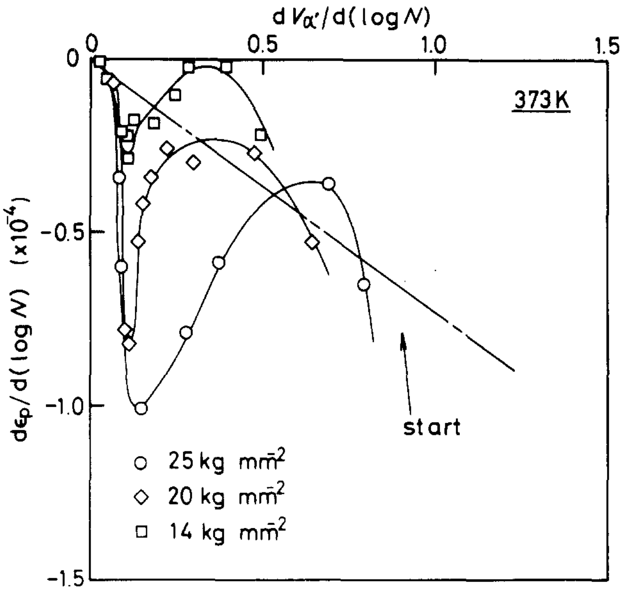

Figure 5 Relationship between the rate of hardening and the rate of martensite formation at 373 K.

HDM in the early stages (up to 210 or 160 cycles), and subsequently it tended to soften up to 10^3–10^4 cycles. Further cyclic deformation led to a tendency of only HDM until failure. When a low stress amplitude ($20\,\mathrm{kg\,mm^{-2}}$) was applied the alloy showed softening not hardening if HDM was excluded, although the specimen itself hardened also during cyclic deformation. The remarkable hardening in the early cycles at high stress amplitudes should be interpreted as the work hardening due to the interaction of dislocations inside or around the induced martensite. Subsequent softening seems to be attributable to the rearrangement of the dislocations.

The specimens behaved at 373 K as they did at 293 K in the early (hardening process) and the middle stages (softening process), but afterwards the conspicuous hardening, except HDM, was observed at each stress amplitude. Although the specimen was highly strained at the elevated temperature by the same applied stress, the austenite phase is so stable that much of the austenite does not transform to martensite but remains as austenite. Therefore, many dislocations might be multiplied and rearranged in this phase and they are considered to form firm and fast cell walls that interrupt the dislocation movement. This may be the reason for the conspicuous hardening in the last stage at the elevated temperature.

From these results and considerations it was concluded that though Fe–30 wt % Ni alloy tended to harden outwardly throughout cyclic deformation, the cyclic plastic behaviour, except the hardening caused by martensite formation, did not always show the tendency of hardening and depended on the test temperature and therefore the stability of austenite.

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