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Dynamic strain-ageing of a copper 1.83 wt % beryllium alloy

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The dependence of dynamic strain-ageing (hardening) on time, temperature, strain and applied stress has been determined for a copper-beryllium precipitation hardened alloy. Strain-ageing increased with increasing time and temperature. The maximum hardening occurred for all times and temperatures at the smallest strains measured (~ 1.5%). Strain-ageing was found to increase with increasing stress drop during strain arrests. No single power law could describe the time dependence, neither could any single activation energy describe the temperature dependence of the hardening.

1. Introduction

Dynamic hardening of metals is often observed following the stress relaxation or unloading of a plastically deformed material. It is manifested as an increase in flow stress when plastic deformation is resumed. Further elongation often produces a yield point effect with the flow stress decreasing to approximately the value of flow stress observed prior to the arrest in elongation.

A comprehensive review of previous work on dynamic hardening has been presented by Hall [1]. In single phase systems, the following observations are generally found.

1. Virtually all materials show dynamic hardening effects, even high purity materials.
2. The magnitude of hardening is small; usually less than a 2% change in flow stress.
3. The hardening is greater if the sample is completely unloaded during the strain arrest.
4. The magnitude of hardening, when plotted against strain, increases to a maximum and then decreases to about zero at large strains.
5. The effect is often more marked at lower temperatures.
6. The time dependence is uncertain; some results show a time dependence while some do not.

In relating the hardening data to models, the presence or absence of a time effect generally determines the model employed. For materials showing a time-dependent hardening, a dislocation-point defect interaction is usually assumed. The Cottrell-Bilby model [2] is the most widely used model with attempts made to find the proper time dependence and an activation energy that can be related to some diffusional process. The time-independent results are thought to arise from dislocation-dislocation interactions [3].

The experimental observations listed above have been made on a large number of pure metals and solid solution alloys. While dynamic hardening has been reported in several precipitation hardened systems [4], no systematic studies have been undertaken to determine if the details of observations in single phase systems can be extended to multi-phase materials. The present work sought to accomplish this objective. Copper-1.83 wt% beryllium was selected as a system to study because dynamic hardening had been previously reported in copper-beryllium alloys [4], and the microstructure has been well characterized. Thus the present study determined the dynamic hardening of a copper-beryllium alloy as a function of relaxation time, temperature, strain, and amount of unloading.

2. Experimental

Copper-1.83 wt%-beryllium (alloy 25) was supplied by Brush Beryllium Corporation. The complete chemical analysis is given in Table I. Samples were prepared from 19 mm extruded rod. After solution annealing 1 h at 1058 K and water quenching, the rod was machined into round, shoulder gripped tensile samples having a 6.4 mm gauge diameter and a 50 mm gauge length. The tensile specimens were aged 3 h at 588 K to achieve maximum tensile strength. The

TABLE I

Be	Co	Ni	Fe	Si	Al	Sn	Zn	Cr	Pb	Cu
1.83	0.30	0.02	0.09	0.10	0.04	0.04	0.02	> 0.005	0.01	Bal

structure generated by this heat-treatment has been shown to be, γ', (body-centred tetragonal) ordered coherent precipitates in a copper rich matrix [5, 6].

The constant total strain (CTS) dynamic hardening tests at room temperature were performed with an MTS hydraulic tensile testing system. The sample was immersed in an alcohol bath to maintain the temperature at 298 ± 0.25K. An Instron G51-11A 25 mm-10% clip gauge extensometer attached to the sample gauge length was used for the system control mode. Most experiments were conducted by loading at a constant strain-rate, arresting the test at some constant strain for a period of time while allowing the applied stress to relax. The sample was then reloaded at the same initial strain-rate. Stability of the system was such that total strain changed by less than 10^{-5} during the arrest. Plastic deformation during relaxation at constant strain is simply the stress drop divided by the sample modulus ($\sim 10^{-4}$). Thus a very negligible amount of work-hardening is expected to occur during relaxation.

The rapid response of this MTS system made it possible to study the time dependence of hardening effects for arrest times as short as 1 sec. The slow response of conventional constant crosshead displacement machines has limited previous studies to times greater than 10 sec. Arrest times of 1, 3, 10, 30, 100, and 300 sec were studied at room temperature and at a strain-rate of $1.6 \times 10^{-5}\,\text{sec}^{-1}$. Samples were also rapidly ($\sim$ 10 sec) unloaded to either 0 or 690 MN m^{-2} stresses and held there for 300 sec to determine if large amounts of unloading would enhance hardening.

Because of the availability of precise constant temperature facilities which have been previously described [7], temperature effect studies were performed on a model TTD Instron machine. In this case, the tests were performed at constant crosshead displacement (CCD). The plastic strain during relaxation was about 10^{-3}; a much larger amount of strain-hardening is expected to occur than during the CTS tests. Strain was again monitored with the G51-11A extensometer. Tests were run at 78, 223, 298, and 373K (± 0.5K) with an arrest time of 300 sec and a strain-rate of $1.6 \times 10^{-5}\,\text{sec}^{-1}$. At 298K, arrest times of 10, 30, and 100 sec were also used to provide comparison with CTS tests.

To relate the results obtained at different strain hardening levels, the data are presented in terms of the fractional hardening, f, given by

$$f = \frac{\sigma_p - (\sigma_0 + \sigma_{wh})}{(\sigma_0 + \sigma_{wh})}\,. \tag{1}$$

Here σ_0 is the flow stress prior to relaxation, σ_p is the flow stress obtained on straining after relaxation and σ_{wh} is the work-hardening which occurs during relaxation. Fig. 1 shows σ_p is taken as the value of flow stress where the slope of the tangent to the curve on restraining is the same as the tangent at the beginning of relaxation. As shown in Fig. 1, the amount of work-hardening occurring during the relaxation, σ_{wh}, is determined by measuring the strain during relaxation and extrapolating the slope of the stress-strain curve just prior to relaxation to that strain.

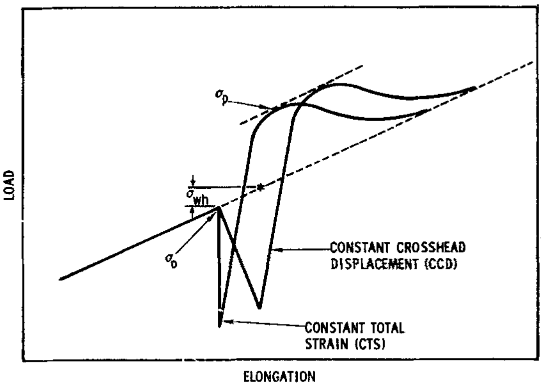

Figure 1 Schematic representation of testing procedure and method of measuring results.

3. Results

All the results are presented as fractional hardening, f, given by Equation 1. The σ_0 values ranged from 1150 to 1420 MN m^{-2}. The work-hardening correction, σ_{wh}, ranged from zero to 7 MN m^{-2} and the actual amount of dynamical hardening ranged from 2 to 24 MN m^{-2}. The time effects studies are presented first with comparison between the constant total strain and constant crosshead displacement results, along with comparison to available theories. Then the effect of unloading on the dynamic hardening is shown. Finally, the temperature results are presented with an attempt to determine an activation energy.

3.1. Time effects

Fig. 2 shows the fractional hardening variation with strain for relaxation times from 1 to 300 sec for the CTS tests. The fractional hardening decreases monotonically with increasing strain. Previously reported results on pure copper [8] and 70-30 alpha brass [9] differ from this finding in that hardening initially increases with strain and then drops off at larger values of strain. The difference between these findings may be related to the fact that the present tests are at a much higher stress at small strains than studies on single phase materials and that dynamic hardening is stress dependent.

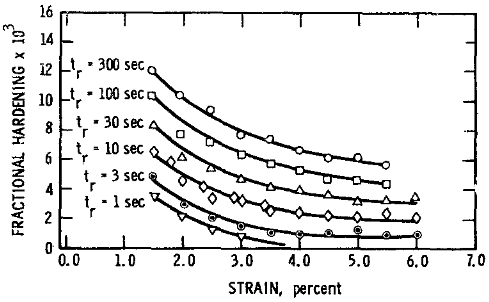

Figure 2 Fractional hardening versus strain for various arrest times. Results are for constant total strain tests (CTS).

To see if any simple time law would fit the results, a log fractional hardening versus log time plot is shown in Fig. 3 for 1.5 and 3.5% strains. The observed slopes vary from 0.8 to 0.2 indicating that no single power law of time satisfactorily describes the results. While Cottrell-Bilby type locking mechanism might be expected at the short times, the results show no consistent slope of 0.67 required by the model.

Also shown on Fig. 3 is a comparison of the constant total strain (CTS) and constant crosshead displacement (CCD) arrest tests. Both results are corrected for strain hardening using Equation 1. At low strains, the fractional hardenings are essentially equal. At higher strains (3.5%) however, there is a significant difference between the two results with the CCD tests consistently showing more hardening. The precise reason for this effect is not clear, but as reported elsewhere [10] there is a considerable difference between the character of stress relaxations conducted in the CTS or CCD mode. The simple extrapolation of the stress strain curve is insufficient to account for work-hardening, σ_{wh}, during relaxation; apparently, more work hardening occurs than is predicted by this extrapolation.

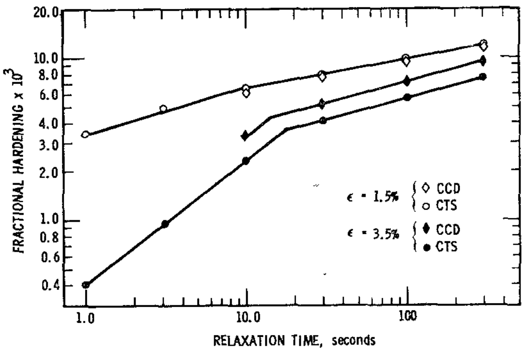

Figure 3 Logarithm of fractional hardening versus logarithm of time for 1.5 and 3.5% strain. Also, comparison of CCD and CTS results.

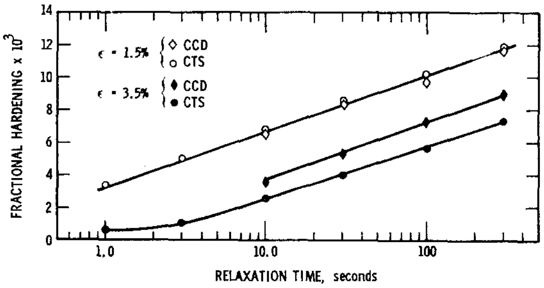

Figure 4 Fractional hardening as a function of logarithm of time.

When the fractional hardening results are plotted against the logarithm of time as seen in Fig. 4, a linear relation is found, except for very short times, $t < 3$ sec. The short time deviation may be due to uncertainty in the time for the 1 sec test (± 0.2 sec). The applied stress versus log time results during stress relaxation behave in the same manner [7]. There is linear variation in stress with log time for all but very short times. There are no models which predict such a time relation for dynamic hardening, but the observed correlation between the hardening and stress relaxation data may be worth further investigation. For example, a model providing a strain-dependent depletion of the mobile dislocation density during a stress relaxation might predict both the stress relaxation and strain-hardening results.

3.2. Unloading effects

The effect of partial or complete unloading during an arrest on the dynamic hardening is shown in Fig. 5. Studies on copper [8] and alpha brass [9] show a similar increase in dynamic hardening when the load is dropped during an arrest. The results also correspond well with the previously reported observation [8] that the fractional hardening varies approximately linearly with the amount of unloading. The increased hardening on unloading is thought to be due to the enhancement of dislocation locking by the backward motion of the dislocations as unloading occurs. The present results appear to agree with that postulate.

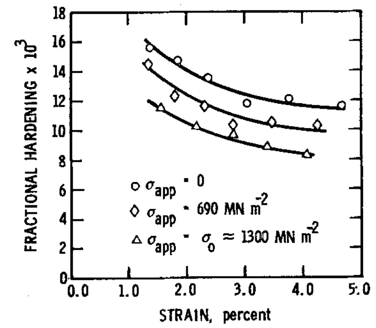

Figure 5 Effect of amount of unloading during arrest on fractional hardening as a function of strain. CTS experiments.

3.3. Temperature effects

Test temperature has a significant effect on the dynamic hardening as shown in Fig. 6. The results at different temperatures follow the same pattern of decreasing fractional hardening with strain. Fig. 7 shows an attempt to obtain an activation energy. The fractional hardening in 300 sec is plotted on a log scale versus the inverse of absolute temperature. The four temperatures tested quite obviously do not yield a single slope for the activated process. Bolling reported similar results for alpha brass [9].

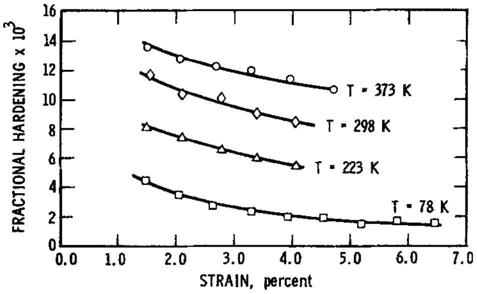

Figure 6 Effect of temperature on fractional hardening as a function of strain. CCD experiments.

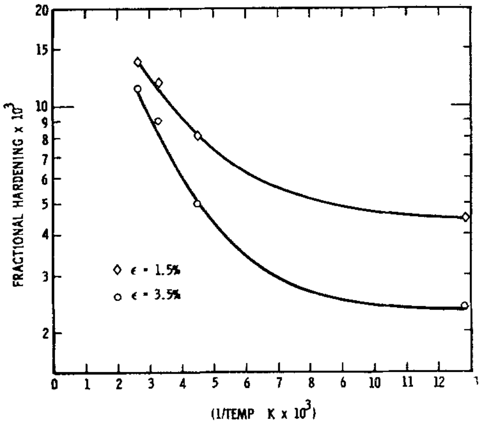

Figure 7 Logarithm of fractional hardening versus one over the absolute temperature. Arrest time equals 300 sec. CCD experiments.

The slope of the straight line through the three higher temperature data points yields an activation energy of only 0.04 eV. An activation energy of this magnitude is well below that for any solute diffusion processes as proposed in models for dynamic hardening. The explanation of the unusually low apparent activation energy is unknown.

4. Conclusions

The conclusions of this research can be summarized in the following set of observations.

1. No presently available models for dynamic hardening can satisfactorily explain either the time or temperature dependence of the results on copper-beryllium.

2. For large values of strain (> 2.0%), simple extrapolation of the stress-strain curve to account for work-hardening during relaxation is not satisfactory to explain difference in hardening in samples undergoing widely varying amounts of plastic strain during the relaxation.
3. The time dependence of the dynamic hardening closely follows a log time plot which is the same time dependence seen in the stress relaxation. This may be a useful correlation in formulating a model which could describe the dynamic hardening process in this material on the basis of a strain-dependent reduction in mobile dislocation density during an arrest.
4. The strain dependence of the dynamic hardening in precipitation hardened copper beryllium is different from that seen in single phase alloys. All single phase systems reach a maximum with strain and then fall off whereas the present study shows the hardening decreasing with increasing strain. This may be due to the precipitation hardening alloy reaching higher stress levels at small values of strain.

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