The Cyclic Stress - Strain Response of Metals and Alloys

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Introduction. When a pure annealed metal is strain cycled, it initially hardens rapidly and subsequently reaches a saturation flow stress particular to the applied strain amplitude. A knowledge of such cyclic stress-strain response is vital to design against fatigue.

In spite of the practical importance of this deformation, studies aimed at understanding the basic mechanisms have been mainly confined to single phase materials. The role of slip character has been elucidated, and the relationship between cyclic and unidirectional deformation established (1). Briefly, long life fatigue deformation is equivalent to the Stage I unidirectional deformation of a single crystal, and short life deformation is equivalent to Stage II. (1)

In two phase materials, a number of ad hoc studies have been made (2-6). One difference from the cyclic response of single phase materials has been the observation that, in certain two phase alloys, softening follows the first stage of hardening; no saturation occurs (3). The present is a brief report of a general investigation into a wide spectrum of two phase microstructural types (7, 8). A new mechanism of softening, along with the circumstances under which it occurs, is offered, and the relationship between the cyclic and unidirectional deformation of these materials is described.

Experimental. To obtain a relatively complete range of two phase and machined into specimens, solution treated, quenched (grain size 0.1 mm), and finally aged by the following treatments: (1) Room temperature aging, producing G.P.I type of clusters. (2) 160° C/5 hours, producing a microstructure containing θ''. (3) 250°C/5 hours producing a microstructure containing θ', designated as the coarse θ'. (4) 2% plastic strain-250°C/3 hours, producing a microstructure containing more closely-spaced θ', designated as the fine θ'. Fatigue tests were carried out on sophisticated electro-hydraulic equipment with regular environmental and testing controls. Plastic strain was the controlling mode.

Results. Phenomenological results for these four microstructures are reflected in the cyclic response curves shown in Figures 1-3. It is seen that the microstructures containing G.P. I and θ'' quickly hardened to a peak stress, then gradually softened. The microstructures containing θ' hardened within a few reversals to a saturation stress for the duration of life.

Structural observations on the microstructures containing G.P. I and θ'' revealed dense intragranular bands of dislocations (Figure 4), with less intense dislocation débris between bands (Figure 5), for all strain ranges.

The microstructures containing the coarse and fine θ' cycled to fracture revealed a single array of dislocations on all the interfaces and single dislocations extending between plates, (Figure 6), at low strains. At high strains, the fine θ' structure showed intersecting arrays on the interfaces with single dislocations between plates. In

the coarse θ' structure at high strains dislocation cells were delineated by the plate interfaces and extended into the matrix as well (7, 8).

Discussion. The results clearly fall into two classes; when the precipitates are small and closely packed, considerable hardening first occurs and then softening sets in: for larger more widely spaced plate-like precipitates, the hardening develops in a few reversals and saturation occurs.

Since the bowing stress is very high in the θ'' microstructure, the precipitates have to be cut by the dislocations. Hardening occurs by intense dislocation multiplication both throughout the grains and, even more concentrated, within deformation bands. The precipitates serve to impose a friction stress on the dislocations by most of the regular dislocation-precipitate interactions including a contribution due to creating APB's in the ordered θ''. However, by the time the peak stress occurs, most of the deformation occurs in the bands and softening is strongly associated with them. No evidence supports the mechanisms heretofore advanced for softening - precipitate reversion, aging inhomogeneities, overaging, (see 9, 10 for review), or particle fracturing (3). Instead, we believe that the precipitates become disordered by irregular to-and-fro dislocation cutting and that softening is mainly due to the loss of the APB contribution to the friction stress (disorder hypothesis). Both the magnitude of the softening supports this view (7, 8), and also the fact that softening occurs in ordered single phase materials such as

Ni_3Mn (11). Moreover, this view explains the general behavior of two-phase alloys containing small precipitates. For example, both Fe-Cu (5) and Cu-Co (2) alloys do not show softening because their precipitates are essentially pure solute, not subject to disordering. On the other hand, in superalloys (3) and pure Al alloys, the precipitates are ordered and, as expected, softening occurs.

The hardening which occurs in microstructures containing larger precipitates such as θ' can be explained (7, 8) in terms of "geometrically-necessary" dislocations (12), previously advanced to explain unidirectional hardening when the particles are not readily penetrable by dislocations (12). Most of the geometrically-necessary dislocations can be generated in the first cycle, and the regularity of the arrays easily permits them or associated single dislocations to shuttle to and fro with successive cycles. This explains both the rapidity of hardening, the detailed differences in flow stress between microstructures of different spacing, and the stability of the saturated state. Of course, at very large precipitate spacings or at the highest strains, "statistically necessary" dislocations can develop by mutual trapping and cell generation, causing the material to behave as if the particles were not present. This was the situation in the case of Leverant and Sullivan's TD nickel (4).

In summary, we note that the cyclic response of two-phase alloys can be treated similarly to unidirectional deformation, both in the application of detailed hardening contributions, and in the sense that there is a precipitate cutting regime, and for coarser microstruc-

tures, an Orowan regime. These conclusions should be useful in predicting the cyclic response of multi-phase alloys in general.

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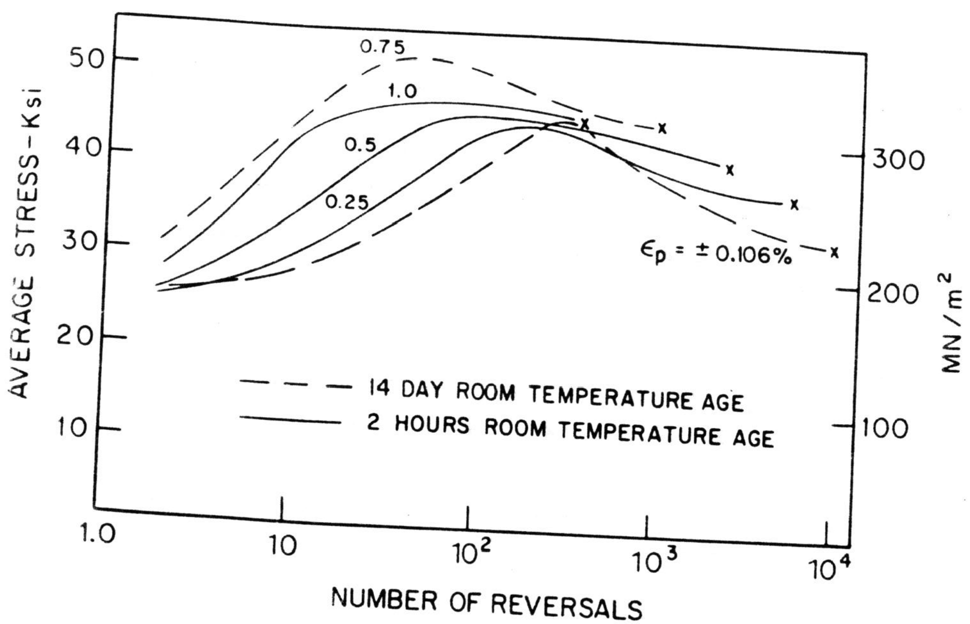
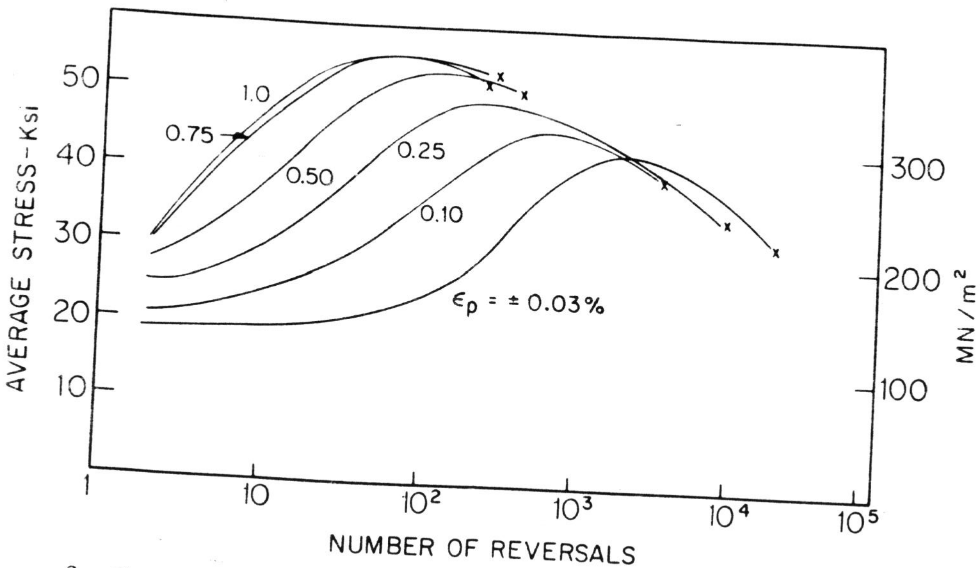
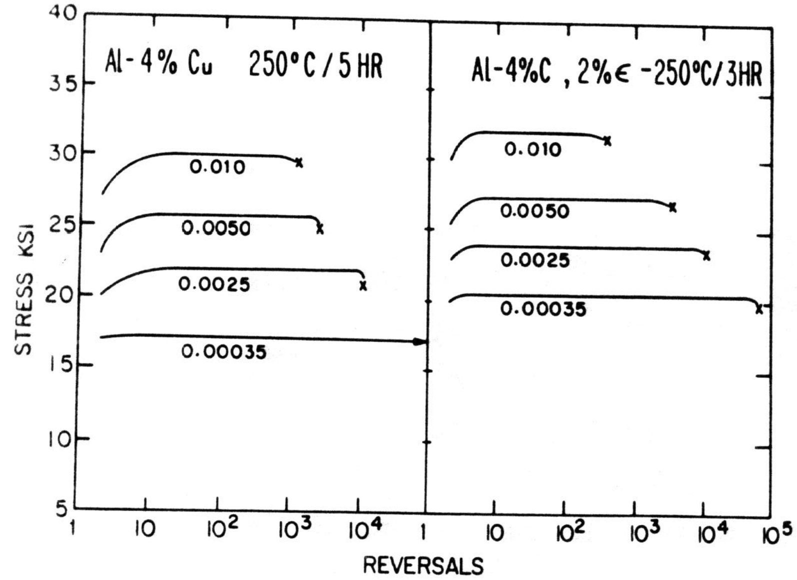


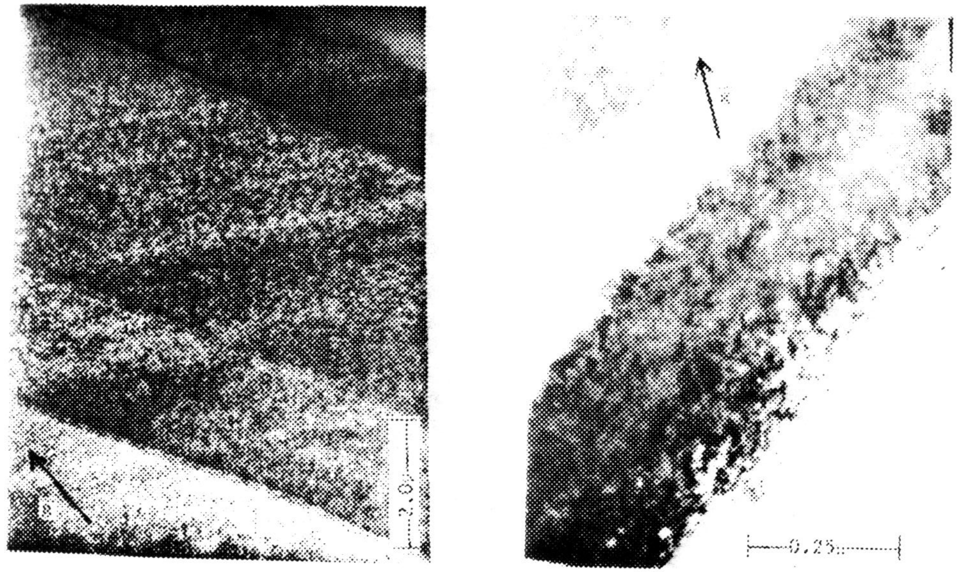
1. Cyclic response curves for a range of plastic strains (ϵ_p) in AL-4% Cu alloy solution treated and aged at room temperature for the times indicated.



2. Cyclic response curves for a range of plastic strain in Al-4% Cu alloy aged at 160° C/5 hrs (θ'' microstructure).



3. Cyclic response curves for a range of ϵ_p in Al-4% Cu alloy - coarse and fine θ' structures.



a) ϵp = $\pm 0.03\%$ $\bar{g}$ = <111> $\overline{ZA}$ = <112>

b) ϵp = $\pm 1.0\%$ $\bar{g}$ = <220> $\overline{ZA}$ = <110>

Figure 4

Microstructure containing θ'' after cycling to fracture showing the typical intense, intragranular deformation bands.

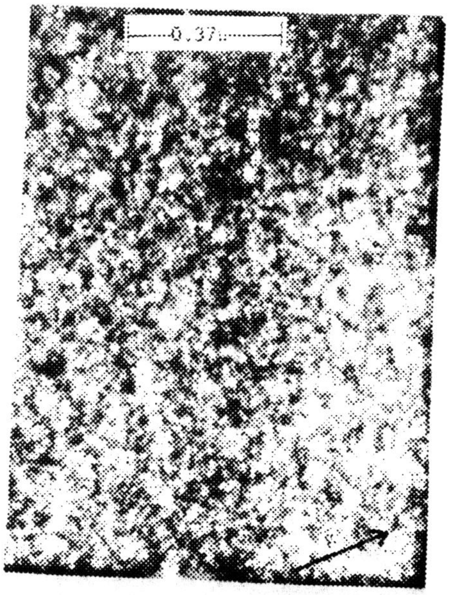


Figure 5

Microstructure containing θ'' cycled at $\varepsilon\rho$ = $\underline{+}$1.0% until fracture showing interband dislocation debris.

$\overline{g}$ = <111>   ZA = <110>

Figure 6

Fine θ' structure cycled at $\varepsilon_p = \pm 0.035\%$ to fracture:

$\overline{g}$ = [002],   ZA = [110].