Analytically Speaking

Dr. Rolf Weil
Professor Emeritus
Stevens Institute of Technology
Hoboken, NJ 07030

Mechanical Properties— Plastic Properties & Fracture

Previous discussions dealt with the means of measuring the mechanical properties of deposit coatings and electroforms. The elastic behavior and the transition to plastic properties (*i.e.*, yielding) were also discussed. To complete the discussion of mechanical properties, plastic behavior and fracture are described.

Plastic Behavior & Strain Hardening

The stress-strain curves of many relatively large-grained and ductile deposits exhibit two slopes, as shown in Fig. 1. The three graphs belong to different tensile specimens prepared from the same copper electrodeposit.

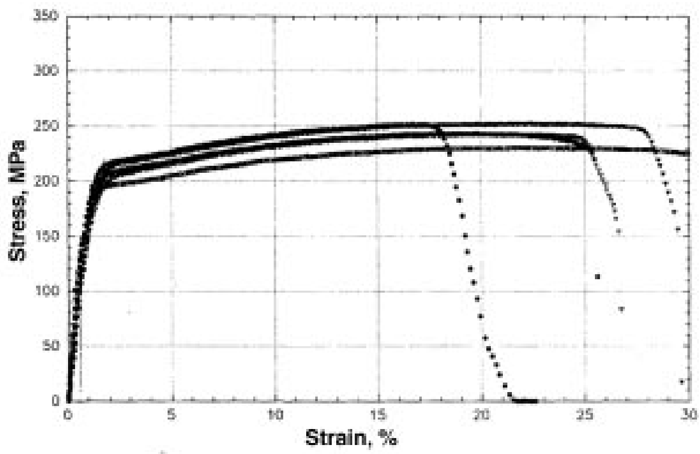

Fig. 1—Nominal stress-strain curves of electrodeposited copper.

In the first part of the curves after the small straight-line, initial elastic portion, the slope is steep. The slope is indicative of the rate of work hardening, which is the strengthening of the material by plastic deformation. Previously, it was discussed how plastic deformation occurs primarily by the movement of dislocations. Obstacles to their movement are the cause of work hardening. Initially, as the dislocations move as a result of plastic deformation, they become tangled. After further plastic deformation, these tangles develop into the walls of areas of relatively low dislocation density, called cells. The cells, with their walls of tangled dislocation, are shown in Fig. 2. Arrows show the cell walls. Cells form only if the grain size is larger than 2 μm. The tangles and cell walls are the main obstacles to the movement of other dislocations and, consequently, to further plastic deformation. The steep slope in the plastic portion of Figure 1 is therefore a result of the formation of tangles and cell walls.

The effect of cell size on strength is shown in Fig. 3. The graphs obtained from fatigue tests show the relationship between the cell size and the maximum stress applied during each cycle that did not cause fracture. The cell size was found[1] to decrease with increasing applied stress. It is again seen that the cell walls are a primary cause of strength. A reason for the difference between the electrodeposited copper from an additive-free plating solution (labeled AF) and wrought foils (labeled W) was probably the surface finish. The boundaries between the columnar grains of the plated copper were small crevices tending to cause fracture at a smaller stress.

The almost horizontal part of the stress-strain curves of Fig. 1 indicates that the deposits experienced much less work hardening. A small amount of strengthening was probably offset by the decrease in the cross sectional area of the tensile specimen. The decrease in cell size was very small, so there were fewer obstacles to further dislocation movement.

Deposits with grain sizes smaller than 2 μm exhibited only the steep-

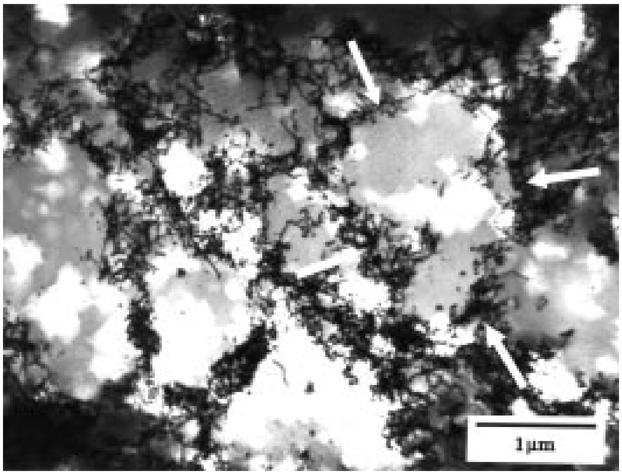

Fig. 2—Transmission electron micrograph showing a dislocation cell structure in electrodeposited copper.

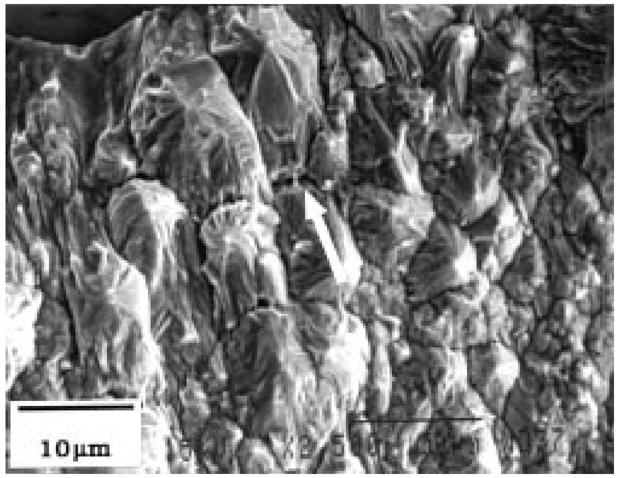

Fig. 3—Graphs of fatigue stress vs. dislocation cell size in electrodeposited copper.

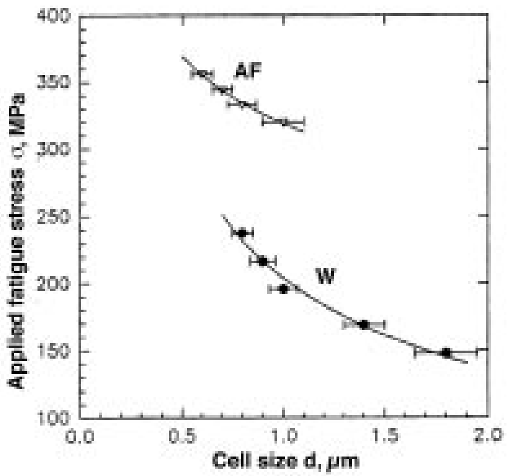

Fig. 4—Scanning electron micrograph of the surface of a copper electrodeposit showing a fatigue crack.

slope portion of the stress-strain curves. Cells did not form in these deposits. The grains were probably too small for cells to form. Accordingly, the grain boundaries became the primary obstacles to dislocation movement. The grain boundaries could not sustain the low work-hardening rates.

The formation of dislocation cell walls has been observed[2] to occur first in the near-surface layer. By plastically deforming during plating, the near-surface layer containing the cells became an interior one; a more uniform distibution of cell walls throughout the thickness was obtained, resulting in a stronger deposit. The dislocation cells in the interior exhibited slip lines, indicating that they had formed while the layers were near the surface. The cells then experienced plastic deformation after they became part of the interior. It is then seen again that cell walls are a primary cause of the strength of some electrodeposits.

Ductility & Fracture

Electrodeposits are ductile if they are relatively free of the defects where fracture can start. Fracture usually begins at defects such as crevices in the surface. The ductility has been found[3] to increase with decreasing ratios of surface roughness to thickness. Figure 4 shows the beginning of a crack where fracture started in a fatigue-tested copper deposit. The crack is indicated by an arrow. It appears that the boundaries between columnar grains are the crevices where fracture starts. Groups of fine-grained deposits often form nodules that appear as hillocks on the surface. The hillocks are often surrounded by voids where fracture can start.[4] Deposits exhibiting the hillocks are less ductile than those having smooth surfaces.

Defects in the interior of deposits can also affect the ductility. A high density of growth twins lowers the ductility.[5] Micropores formed by the incomplete coalescence of crystallites are often filled with hydrogen and result in lower ductility. The codeposition of addition agents that decompose, resulting in development of porosity, reduce the ductility. When the pores are filled with hydrogen, they are under tensile stress. As a result, the pores become sites of stress concentrations that lead to premature fracture and, consequently, low ductility. The lower ductility of one of the copper specimens in Fig. 1 was caused by voids. As also seen in Fig. 1, the voids apparently did not affect the rate of work hardening. Deposits containing large concentrations of hydrogen generally exhibit low ductility. The fracture path in electroless copper deposits, which usually contain high hydrogen concentrations, has been observed to follow the voids.[4] The brittleness of hard gold has also been attributed to grain-boundary voids.[6]

The ductility of composites consisting of alternate layers of nickel and copper was found[7] to increase with decreasing layer thickness to about 3 µm.

Ductility is also affected by the way specimens are tested. The ductility is different if the specimens are tested uniaxially in a tensile test or biaxially in a bulge test. The strain of copper deposits at fracture has been found[7] to increase when the thickness increased from 3 to 25 µm, then decreased. Deposits having a non-cubic crystal structure, such as zinc, have a different ductility depending on whether they have a preferred orientation. If the orientation is such that the slip planes are mostly perpendicular to the tensile direction, the deposits appear brittle.

The sharp decrease in the stress seen in the three graphs of Fig. 1 was a result of necking. Necking is a large local reduction in the cross sectional area within a small part of the gauge length. The sharp decrease appeared because in Fig. 1 the engineering stress (which is the load divided by the original cross sectional area), was plotted. If it had been possible to measure the instantaneous area to calculate and plot the true stress (the load divided by the instantaneous area), it would have increased. Because of the sharp decrease in the cross sectional area when necking occurred, an increase in the true stress would still involve a decrease in the load. Accordingly, the engineering stress, which is proportional only to the load, decreased.

In the necked portion, the reduction in the cross sectional area at fracture may be very large, indicating high ductility. The ductility, however, as measured from the longitudinal length change, is then usually much smaller than the reduction in area. The reason is that while there can be a large length change in the necked portion, if it is averaged over the whole gauge length, the elongation is small. Consequently, deposits that may be quite ductile appear relatively brittle. If the deposit adheres well to a more massive substrate, it cannot exhibit necking. The strain in the deposit is then governed by the properties of the substrate. If the deposit is more brittle than the substrate, however, it will eventually crack. **P&SF**

References

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