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CREEP PROPERTIES OF TWO DIFFERENT MO BASE ALLOYS (TZM)

H. Calderon-Benavides, G. Kosterz and G. Ullrich*,
 Institut für Angewandte Physik, ETH Zürich, CH-8093 Zürich,
 * Eidgen. Institut für Reaktorforschung, CH-5303 Würenlingen

The creep properties of the TZM alloy (Mo-1Ti-0.08 Zr-0.16 C at.%) have been studied. Two commercial alloys, one produced by vacuum melting, the other by powder metallurgy, have been compared. Special attention has been given to the microstructure in the as-received 50 mm diameter bars and after creep deformation at 1150°C (~half the melting temperature).

Differences are found in the distribution, size and chemical composition of the precipitates as well as in the dislocation densities and grain and subgrain sizes in the two alloys under consideration. In all cases the grains are elongated along the extrusion direction. A higher aspect ratio R ($R = L/w$, where L is the mean length and w the mean width) is found in the vacuum melted alloy. This alloy shows a grain shape that depends on the location in the bar.

The alloy produced by a powder metallurgy process contains large particles (5-10 μm) and a distribution of semicoherent (TiC) particles (10-20 nm in diameter). The large particles were identified as ZrO_2 from EDX measurements and diffraction analysis. The small particles are homogeneously distributed throughout the subgrains and the subgrain boundaries. According to the alloy composition and analysis based on the observed contrast in electron microscopy, they can be tentatively identified as TiC. The dislocation density as measured by transmission electron microscopy depends on the location of the specimens in the bar. At the edge of the bar, the dislocation density is found to be $5 \times 10^{11} \text{ cm}^{-2}$. At the bar center it is $8 \times 10^9 \text{ cm}^{-2}$. In addition to the subgrain boundaries, some other dislocation arrangements, e.g. networks, walls, etc., are seen, but they are apparently in an intermediate stage of formation.

The vacuum melted alloy contains only incoherent particles (25-50 nm). Electron diffraction analysis made on extracted particles indicates that they correspond to $\text{Zr}_x\text{Ti}_{1-x}\text{C}$ precipitates where x varies between 0.5 and 0.6. The dislocation density is $0.6 - 1.4 \times 10^{10} \text{ cm}^{-2}$ at the center of the bar. At the bar edge, a very low density of dislocations is observed (approximately 10^7 cm^{-2}). Most dislocation arrangements, e.g. networks, are almost perfectly developed.

The particle-dislocation interactions during deformation at high temperature are characteristic of each alloy and are related to the different particle chemistry. This affects directly the creep behavior, e.g. the relation between steady-state creep rate and applied constant stress (Fig.1). All tests were performed in compression in a screw-driven testing machine. The creep response of the sintered alloy is approximately homogeneous throughout the section of the bar. For the vacuum melted alloy, the creep resistance near the center of the bar is higher than that for the sintered alloy. However, the creep rate $\dot{\epsilon}$ increases considerably near the edge of the vacuum melted bar.

The creep behavior of these alloys at different temperatures and stresses has been the subject of previous investigations (1,2). Nevertheless, basic

Fig.1 Steady-state creep rate $\dot{\epsilon}$ as function of applied stress σ for two TZM alloys at 1423 K. p = sintered alloy, c = vacuum melted alloy. The position of the specimen in the original bar is indicated by a number between 1 (edge) and 5 (center).

microstructural information regarding the deformation mechanisms is lacking at present. In this investigation, transmission electron microscopy has been used to obtain this type of information. Some of the different dislocation arrangements (subgrain boundaries, networks, etc.) have been analyzed after creep deformation. Most of the observed arrangements are in equilibrium, i.e., they do not exhibit any long-range stress distributions, but their role during creep requires more analysis. On the other hand, small precipitates (3-5 nm in diameter) are observed to appear during creep deformation with a density and distribution that depend on the alloy under analysis and probably also on the applied stress and induced deformation. These fine precipitates, most of them located at subgrain boundaries, interact very strongly with dislocations, giving rise to an effective hardening with a very low total volume fraction of precipitate (< 1%). Precipitation is also observed during ageing at 1150°C, but in this case, the particles are homogeneously distributed.

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