The $\omega$-Phase as an Example of an Unusual Shear Transformation

J. C. WILLIAMS, D. de FONTAINE, AND N. E. PATON

The conditions leading to formation of the metastable $\omega$-phase in Ti, Zr, and Hf alloys are described and earlier experimental observations directly related to $\omega$-phase formation are summarized. New experimental results are presented which show that increased oxygen content suppresses athermal $\omega$-phase formation in Ti-V alloys. A mechanism of $\omega$-phase formation based on the formation of a linear displacement defect is described and it is shown that both the earlier and new experimental results are consistent with this mechanism. Additional experimental results on variations in electrical and physical properties in Ti alloys are also shown to be consistent with the linear displacement defect model.

**BACKGROUND**

It is well-known that the $\omega$-phase can be formed in metastable bcc alloys of Ti, Zr, and Hf during quenching (athermal $\omega$) or during aging (isothermal $\omega$); further, these two forms of $\omega$-phase have been separated in many earlier treatments of the subject. Such separation has largely been based on observed variations in intensity and breadth of the $\omega$-phase X-ray diffraction lines leading to the term "diffuse $\omega$" for athermal $\omega$. As will be demonstrated below, no physical basis exists to justify such a separation. Thus, in the following portions of this paper the terms athermal and iso-thermal will only be used to describe the thermal conditions whereby the $\omega$ under discussion has been formed.

The composition range over which athermal $\omega$-phase forms has been shown to depend on the type of solute and, according to some investigators, on the electron-atom ratio of the alloy which is controlled by the solute valence and concentration. The justification for the latter point is not clear since several schemes have been employed to assign valence to transition element solutes and all seem to produce a correlation. Furthermore, the solute-rich composition limit for athermal $\omega$-phase formation has been subject to disagreement; the bulk of these disagreements are largely traceable to the relative detectability of athermal $\omega$ by various experimental techniques. For example, early workers usually employed variations in hardness to indicate the presence of athermal $\omega$-phase. Recent studies have shown that little or no hardening accompanies $\omega$ formation until the vol fraction exceeds 0.2 to 0.25; thus the use of hardness measurements introduces a significant underestimate of the composition limit for athermal $\omega$ formation. X-ray diffraction has probably been the most successful technique for $\omega$ detection although many recent studies have employed selected area electron diffraction for this purpose. Comparison of the electron and X-ray diffraction results have been complicated by two factors. Firstly, the detection of small vol fractions of $\omega$ can only be reliably accomplished using single crystal X-ray techniques, such techniques have not gained wide acceptance due to the relative scarcity of single crystals. Secondly, the electron diffraction patterns contain complex networks of diffuse intensity, the nature of which will be described below. The diffuse intensity overlaps the $\omega$ diffraction maxima positions, causing the detection of $\omega$ in such patterns to become somewhat subjective. Thus, while the $\beta \rightarrow \omega$ transformation in principle occurs at a well-defined temperature ($\omega$) for a given concentration, in practice the experimental difficulties enumerated above have caused the solute rich composition limit to be ill-defined.

In addition to the above experimental difficulties, it has recently been shown by Paton and Williams that minor alloying element concentration can have a large effect on $\omega$. They found that the amount of oxygen in a series of Ti-V alloys could change $\omega$ by more than 100°C over a range of oxygen commonly encountered
in laboratory heats. These results are shown in Fig. 1. Much of the disagreement over the composition dependence of $\omega_s$ may thus be attributed to variations in concentration of minor alloying elements such as interstitial O. This effect of interstitials on $\omega_s$ may also have important implications in terms of the transformation itself, as discussed later.

The diffuse streaking in electron diffraction patterns has been reported in a wide variety of bcc Ti, Zr, and Hf base alloys. This streaking, which roughly can be described as sheets of intensity lying on (111) rel-planes, is illustrated in Fig. 2. As either the temperature or the solute concentration increases, the streaks become more curved (Fig. 3). The origin of this streaking has been analyzed in a previous article and this analysis will be described and elaborated in a later section of this paper.

In alloys which exhibit well-defined athermal $\omega$ intensity maxima, the phase has been shown to occur as 20 to 40Å diam particles of roughly equiaxed morphology. Sass$^{14}$ has reported that the particles often form in rows, however the generality of this effect is not clear at present.

Recent cooling stage experiments performed in the electron microscope have also demonstrated that $\omega$ can form athermally at temperatures as low as 90 K.$^{13}$ These experiments have also demonstrated the total reversibility of the athermal $\beta \rightarrow \omega$ transformation. Earlier Russian work$^{13}$ showed that extremely fast quench rates as high as 11,000$^\circ$C per s were insufficient to suppress the athermal formation of $\omega$. The combined observation of very rapid kinetics and low temperature transformation lead to an earlier suggestion that $\omega$ formed martensitically,$^{14}$ however the product phase does not conform to the requirements of a classical martensitic product. It has been suggested recently that $\beta \rightarrow \omega$ transformation occurs by correlated displacements,$^{15}$ the details of this latter treatment will be described in a subsequent section since they comprise the central theme of this paper.

The structure of athermal $\omega$-phase has been the subject of an earlier dispute between Silcock,$^{16}$ who claimed it belonged to space group D$_{3d}$ (P6/mmm) and Bagariatskii$^{17}$ who claimed it belonged to D$_{3d}$ (P$^\beta$m1). Recently Sass and Boric$^{18}$ have shown that increasing solute concentration causes the sixfold symmetry charac-

---

**Fig. 1**—Omega start temperature ($\omega_s$) plotted vs oxygen composition for a series of Ti-V alloys.

**Fig. 2**—Selected area diffraction patterns of a Ti-12 wt pct Fe alloy illustrating (111) rel-plane streaking. (a) [110]$_\beta$ zone normal, (b) [102]$_\beta$ zone normal, (c) [111]$_\beta$ zone normal.