next publication. On the other hand, the important parameter mismatch between the oxygen sublattices of both structures (about 5% linear) induces important mechanical strains, the relaxation of which was observed to occur (for reaction extents higher than in the present case) as spontaneous cleavages predominantly on (0 0 1) type planes of the crystals habit, that is on (0 0 1) type planes of both structures.

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Observation of transgranular slip in an austenitic stainless steel

In recent years, a considerable amount of effort has been directed at producing theoretical models for transgranular slip; i.e. the passage of crystal deformation from grain to grain across the boundary which separates them [1–6]. In this short note, we present some examples of transgranular slip observed in thin foils of a 20% Cr–25% Ni austenitic stainless steel. During the course of the present investigation transgranular slip has only been observed when the two active slip planes (one in either grain) and the grain boundary have a common zone axis. This condition allows the passage of crystal slip from one grain to the other without any rotation of the dislocation line vector as it crosses from Grain A to Grain B. Similarly, unambiguous evidence of transgranular slip was restricted to the case where dislocation motion occurred during or after foil preparation and not in the bulk specimen. In the former case, stress relations at the foil surfaces allow dislocations to move more easily, whilst the position and orientation of the slip bands formed are defined by the slip traces at the foil surfaces.

Fig. 1 shows an example of transgranular slip (from Grain A to Grain B). In this instance an array of grain-boundary dislocations is observed on one side of the slip plane–boundary intersection. This array of dislocations arises by the following dislocation reaction:

$$\mathbf{b}_A \longrightarrow \mathbf{b}_B + \mathbf{b}_{gb}$$

where $\mathbf{b}_A$ and $\mathbf{b}_B$ are the matrix Burgers vectors in Grains A and B, and $\mathbf{b}_{gb}$ is the Burgers vector of a residual boundary dislocation. A further example of this type of reaction is given in Fig. 6 of Howell *et al.* [7]. It should be noted that the residual boundary dislocations will not, in general, be glissile. Progressive transgranular slip generated by a single active slip plane on one side of the boundary will only occur when diffusive fluxes allow associated climb and glide of the residual boundary dislocations away from the slip plane–boundary intersection. A further example of transgranular slip is shown in Figs. 2a to 2c. For this case, however, the passage of crystal slip from Grain A to Grain B has resulted in the emission of dissociated dislocations rather than total dislocations. This observation was common to many of

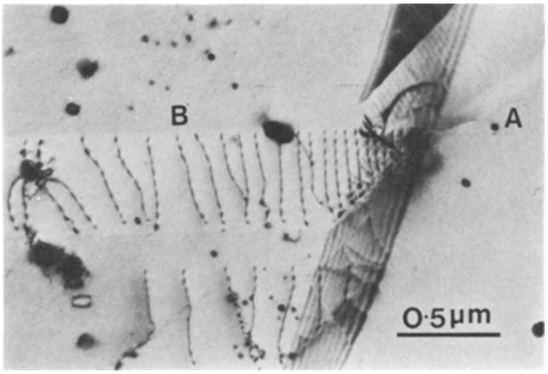

Figure 1 Transgranular slip across a high-angle grain boundary. An array of grain-boundary dislocations is observed on one side of the slip plane–boundary intersection.

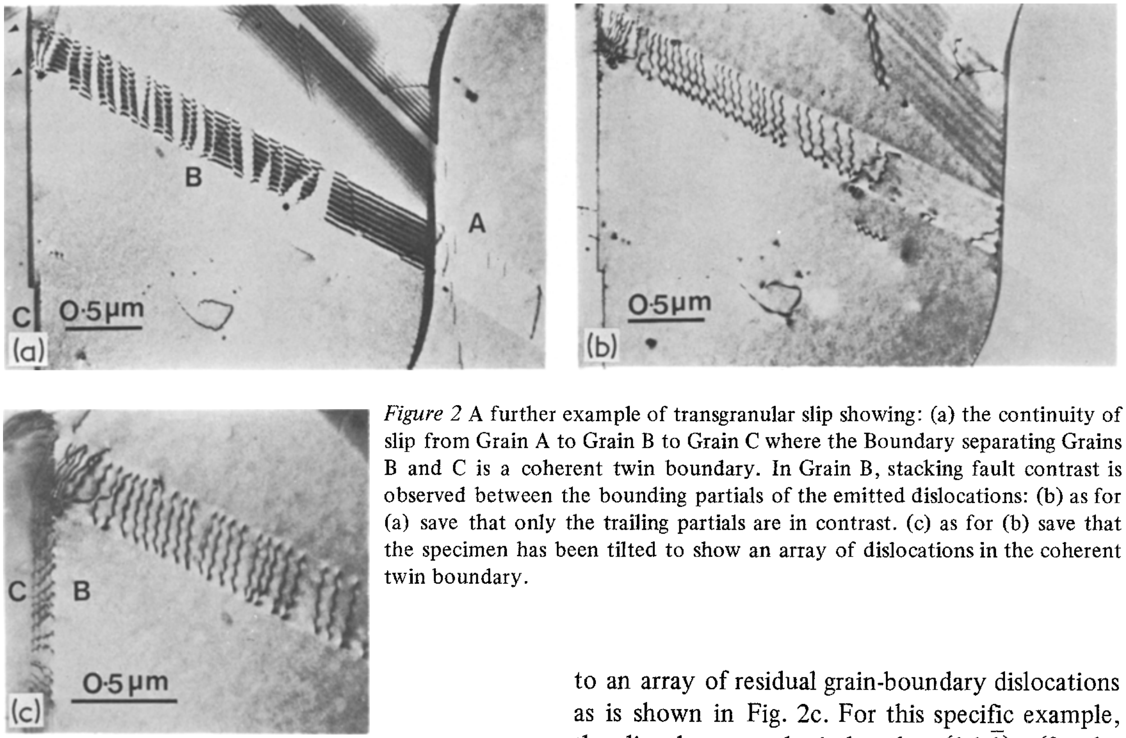

Figure 2 A further example of transgranular slip showing: (a) the continuity of slip from Grain A to Grain B to Grain C where the Boundary separating Grains B and C is a coherent twin boundary. In Grain B, stacking fault contrast is observed between the bounding partials of the emitted dislocations: (b) as for (a) save that only the trailing partials are in contrast. (c) as for (b) save that the specimen has been tilted to show an array of dislocations in the coherent twin boundary.

the transgranular slip events observed in material which had been deformed after foil preparation. However, for bulk deformed material, dislocations were not observed to be extended. Thus it would appear that the observation of partial dislocation emission is a phenomenon associated with the deformation of thin foils rather than an effect characteristic of bulk deformation. This conclusion is supported by the work of Kuhlmann-Wilsdorf and Wilsdorf [8] where it was intimated that dislocations moving in the bulk are constricted due to the effect of absorbed point defects whilst in thin foils, dislocations can lose these absorbed point defects to the free surfaces by pipe diffusion. Thus, it would appear that the observation of stacking fault emission, whether as a singular process, or as a part of a transgranular slip event, cannot be unambiguously related to bulk deformation processes.

A further point of interest with regard to the images shown in Fig. 2 is the continuation of the crystal slip process from Grain B to Grain C across the coherent twin boundary; the slip trace in Grain C being arrowed in Fig. 2a. This transgranular slip event, in common with that shown in Fig. 1, leads to an array of residual grain-boundary dislocations as is shown in Fig. 2c. For this specific example, the slip planes can be indexed as $(1\,1\,\bar{1})_B$ (for the slip plane in Grain B) and $(\bar{1}\,\bar{1}\,1)_C$ (for the slip plane in Grain C) for a $(1\,1\,1)_{B,C}$ coherent twin plane. These three planes meet in a common zone axis ($[\bar{1}\,1\,0]_B$) such that transgranular slip can be maintained without any dislocation line rotation. Similarly, the observed dislocation reaction is consistent with:

$$a/2\,[1\,0\,1]_B \rightarrow a/2\,[\bar{1}\,0\,\bar{1}]_C + a/6\,[\bar{1}\,\bar{1}\,2]_B$$

(and see Fig. 5 of [5]).

The retained twin boundary dislocation ($a/6\,[\bar{1}\,\bar{1}\,2]_B$) is glissile in the boundary plane and may glide along the twin away from the slip band–twin boundary intersection (namely Fig. 2c). Hence, for this particular example, the process of transgranular slip does not appear to lead to the formation of a large step in the boundary as has been suggested (for example [6]).

The examples shown above indicate that transgranular slip may occur under favourable circumstances. However, it should be appreciated that this phenomenon only occurs infrequently, which implies that most transgranular deformation events will be far more complex than the simple situations presented in Fig. 1 and 2.

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Sintering and grain growth in doped ZnO

Recent new uses of ZnO-based ceramics in electronics demand close control of grain size. Although several authors reported on sintering and grain growth of ZnO, the influence of various additives is not yet satisfactorily explained.

On heating ZnO becomes oxygen deficient. Material transport during sintering may be controlled either by the diffusion of Zn^{2+} or O^{2-} ions and the published data support both possibilities [1, 2]. This note reports data which show how the onset of sintering and the grain size of sintered specimens can be influenced by adding a small amount of another oxide.

Experiments were carried out with ZnO p.a.* with a grain size below 1 μm (Fig. 2). Up to 0.5 wt% of TiO_2 and Sb_2O_3 were added as powders, and Li_2O and K_2O were added as aqueous solutions of Li_2CO_3 and K_2CO_3 respectively. The shrinkage curves of cylindrical pellets of 57 ± 1% of theoretical density were determined by the Netzch dilatometer with a heating rate of $300° C h^{-1}$. After reaching 1180° C the samples were anealed for 2 h at this temperature and were examined for density and microstructure.

During the sintering of 0.5% doped ZnO the most remarkable effects are the rapid sintering of Li_2O-doped samples and the shift of the onset of sintering of Sb_2O_3-doped ZnO towards higher temperatures (Fig. 1).

The beneficial effect of Li on the sintering of ZnO was observed previously by Marshal *et al.* [1], who concluded that sintering is controlled by the diffusion of oxygen vacancies through the lattice. On the other hand later studies of the sintering of ZnO [2] showed strong indications that the lattice diffusion of zinc is the rate-controlling step.

The concentration of interstitial zinc ions in ordinary zinc oxide is very small [3]. The incorporation of Li into the ZnO lattice produces Zn-

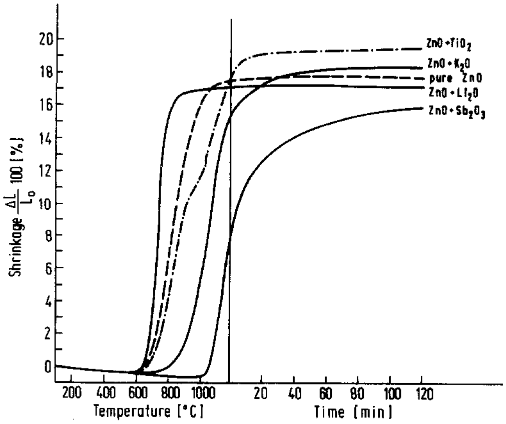

Figure 1 Sintering curves for pure and 0.5% doped ZnO samples.

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*p.a. = pro analysis (equivalent to reagent grade).