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Quantitative atomic force microscopy analysis of slip traces in Ni_3Al yield stress anomaly

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Abstract

It is well known that Ni_3Al intermetallic compounds of the $L1_2$ ordered structure exhibit positive temperature dependence of yield stress over a finite temperature range. In this work, we examined the slip markings produced by plastic deformation at the free surface of $Ni_3(Al,Hf)$ specimens deformed *in situ* under an atomic force microscope at room temperature. The results demonstrate that a high exhaustion in the mobile dislocation density does take place in the yield-stress anomaly of this alloy.
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Keywords: Ni_3Al; Cross-slip; Slip traces; AFM

1. Introduction

The positive temperature dependence (PTD) of flow stress of Ni_3Al intermetallic compounds, currently referred to as the yield-stress anomaly (YSA), has not yet received a satisfying explanation. Several theoretical works succeed in modelling the flow stress increase with temperature, but they generally fail in predicting other deformation characteristic, such as the high work-hardening rate (which also exhibits a PTD) and the strain-rate sensitivity of the flow stress (for a review see [1]). A majority of the proposed models considers that a thermally activated cross-slip process plays a key role in the YSA. However, depending on the considered dislocation dynamics, the cross-slip process leads with increasing temperature either to a decrease in the dislocation velocity (v) [2,3] or to a decrease in the mobile dislocation density (ρ_m), referred to as velocity-type or exhaustion-type anomaly, respectively [4,5].

The purpose of this paper is to report on Ni_3Al slip traces at the atomic force microscopy (AFM) resolution level. The goal of this ongoing study [6] is two-fold. First, we want to unambiguously discriminate whether the YSA of Ni_3Al corresponds to a velocity-type or an exhaustion-type anomaly. Second, we expect to fully characterise the mechanism responsible of the YSA and to develop a model that will predict all the deformation characteristics of the Ni_3Al phase (for a concise survey see [7]). For this, we examined the overall features of slip traces produced during *in situ* deformation of $Ni_{74.8}Al_{21.9}Hf_{3.3}$ (at.%)specimens under AFM. Particular attention was paid to follow the evolution of slip line length together with the number of emerging dislocations per slip line. The results are discussed in regard of velocity versus exhaustion type models that have been proposed for explaining the YSA.

2. Experimental

$Ni_{74.8}Al_{21.9}Hf_{3.3}$ specimens were spark-eroded from a single crystalline rod kindly provided by D.P. Pope at the University of Pennsylvania. Prior to deformation, the specimens were mechanically polished for AFM observations using both alumina powder and silica suspension. The compression axis was for all specimens along the $[\bar{1}\,2\,3]$ crystallographic orientation with $\pm[5\,4\,\bar{1}]$ and $\pm[1\,\bar{1}\,1]$ oriented faces. The characteristics of such orientations have already been presented in detail in [8]. This orientation has a high Schmid factor on the $(0\,1\,0)$ cube cross-slip plane, for promoting dislocation glide on this plane, while the Schmid factor on the $(1\,\bar{1}\,1)$ octahedral cross-slip plane is zero. Slip markings corresponding to the $[\bar{1}\,0\,1](1\,1\,1)$ primary octahedral glide system are visible on two opposite $\pm[5\,4\,\bar{1}]$ specimen faces only and are lying at 62° from the compression axis. The unit step height h, corresponding to the emergence of one superdislocation, is equal to two times the Burgers

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vector component of a superpartial dislocation projected onto the surface normal, i.e. $h = 0.234$ nm.

The specimens were deformed at room temperature (RT) under constant strain-rate condition. The AFM observations were performed *in situ* during the deformation tests by stopping the moving grips after given plastic strain increments. The AFM observations are thus performed under relaxation deformation conditions, for which plastic strain can be considered negligible [9]. A detailed description of the deformation set up combined with AFM facility has been reported in [10]. Note that the YSA domain of $Ni_{74.8}Al_{21.9}Hf_{3.3}$ starts at a temperature far below RT. Therefore for this compound, RT corresponds to a temperature which is located inside the YSA domain.

3. Results and discussion

The slip markings observed for $Ni_{74.8}Al_{21.9}Hf_{3.3}$ are similar to those of $Ni_{75}Al_{24}Ta_1$ [6]. In particular, the traces belong to the primary (1 1 1) slip plane without appreciable deviations at the level of AFM resolution. This indicates that, in the YSA of Ni_3Al, the cross-slip distance is small, even null, and cross-slip events are rather scarce. As an example, the fine structure of slip lines is presented in Fig. 1, for 0.48% plastic strain. A single step resulting from the emergence process of dislocations appears as an individual dark line in such signal error contact mode AFM observations. These AFM images are not calibrated in the direction perpendicular to the observation face, but have visual advantage of fine detail enhancement in the plane surface. The scratch feature at the upper left part on the image was used during the *in situ* experiment as a guide to follow the slip trace evolution. The main characteristics are the numerous unusual slip lines ending within the scan size. It results from the decrease of slip lengths with increasing temperature [6], so that the distance travelled by mobile dislocations becomes shorter with the increasing temperature in the YSA.The scan size defined for the quantitative analysis results from a compromise between good lateral resolution, for discriminating between parallel slip traces, and a scan area large enough to allow a statistical analysis of slip trace nucleation and propagation. The number of emerging dislocations per slip line has been estimated from topographical contact mode AFM observations, together with the number of slip lines, with increasing plastic strain over representative scan areas. It must be emphasised that the high strain-hardening rate of $Ni_{74.8}Al_{21.9}Hf_{3.3}$ leads to strong increase of the applied stress per strain increment. The results of the statistical slip line analysis are shown in Fig. 2a. The number of emerging dislocations per slip line does not evolve with increasing plastic strain and is rapidly constant about a mean value of 3–4, while new slip traces are continuously appearing. This clearly demonstrates that sources do not emit more than 3 or 4 dislocation loops before to be blocked.

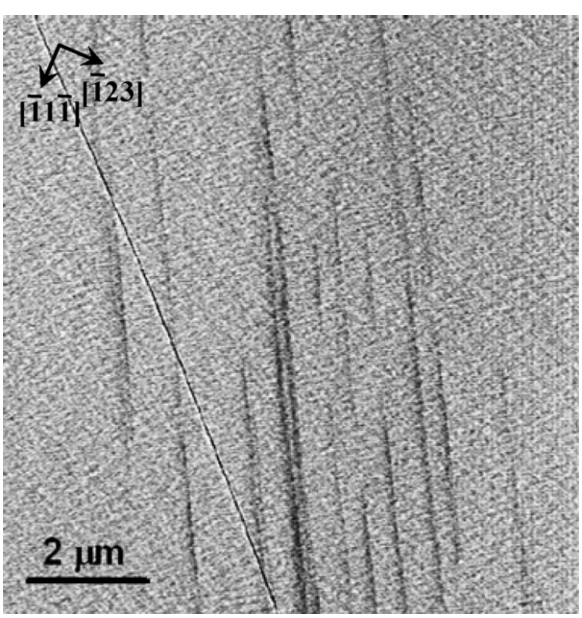

Fig. 1. Signal error AFM image of a $Ni_3Al(Hf)$ single crystal deformed at room temperature in the yield-stress anomaly for 0.48% plastic strain. Nanometer scale steps resulting from the emergence process of superdislocations appear as dark lines. The scan size is 10.8 μm × 10.8 μm. The scratch feature at the upper left part was used for the statistical analysis as a guide to follow the slip trace evolution. The fine slip line structure is characterized by numerous traces ending within the selected scan size.

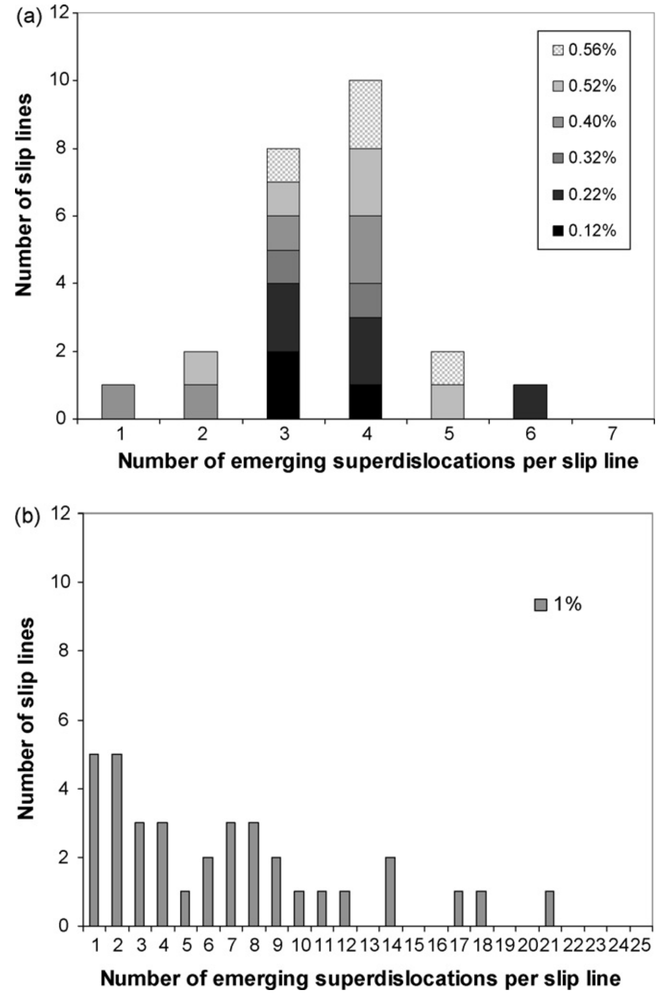

Fig. 2. Number of emerging superdislocations per slip line for various plastic deformation values, in the case of (a) $Ni_3Al(Hf)$ and (b) $Ni_3Al(Ta)$.

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