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## Dislocation Exhaustion and Stress Anomaly in $L1_2$ Alloys: The ‘E.L.U.’ Model

François Louchet

Groupe Physique du Métal, LTPCM(\*), ENSEEG Institut National Polytechnique de Grenoble  
BP 75, 38402 St Martin d’Hères cedex, France

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**Abstract.** — A statistical model is proposed to account for strength anomalies in  $L1_2$  alloys. It is based on a competition between locking of superdislocations into Kear Wilsdorf locks and by-passing these locks by macrokinks. From the statistical distribution of macrokink heights, it is shown that the population of mobile dislocations is exhausted with an increasing efficiency as temperature is raised. This yields either an exhaustion creep at constant applied load, or a strength anomaly in constant strain rate conditions. Several consequences on strain rate sensitivity, work-hardening, or flow stress reversibility are envisaged.

### 1. Introduction

Though it is generally agreed that strength anomalies in intermetallic alloys originate from a variety of dislocation mechanisms, some other common features, as a small strain rate sensitivity, a large work hardening rate or an almost reversible flow stress are usually reported. It has been shown in a previous work [1,2] that, in the case of  $L1_0$  alloys, a statistical description of the evolution of the dislocation population (LPU model) yields an exhaustion of mobile dislocations, accounting for all these features. The aim of the present paper is to show that a rather similar statistical treatment, the ‘Extended-Locking / Unzipping’ or ‘ELU’ model, can be applied to the dislocation mechanisms specific of  $L1_2$  alloys, leading directly to the strength anomaly as well as to the other associated characteristic properties.

There is now an almost general agreement based on both post-mortem and *in-situ* TEM observations on the nature of dislocation mechanisms that control the plasticity of  $L1_2$  alloys in the domain of strength anomaly. Superdislocations consist in nearly screw segments connected by macrokinks (MK) lying in the octahedral plane. They are supposed to lock at places into Kear Wilsdorf (KW) barriers through thermally activated cross-slip toward cube planes. There is still some controversy on whether KW locks are overcome directly by an unlocking process [3] or merely by-passed by MKs [4]. There is actually some evidence of unlocking events [3] in *in-situ* experiments, but in foils generally thinner than the average separation between MKs [5].

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(\*) CNRS-URA29

There is also some evidence that KW locks can sometimes bow out in the cube plane [6], which makes the unlocking process quite unlikely at least in this case. This is why the present model is based on the by-passing (or ‘unzipping’) mechanism rather than on the unlocking one. This assumption will be further justified by some of its consequences.

## 2. Exhaustion and Multiplication Rates

The height  $\lambda$  of a MK is directly related to the free flight distance of a loop expanding in the octahedral plane. The locking probability being the same on each atomic row swept by the dislocation, we can write the MK height distribution (in terms of a number of MK of size  $\lambda$  per unit dislocation length) in the form of an exponential decay [5]:

$$n(\lambda) = n_0 \exp \left[ -\nu_0 \frac{\lambda}{\nu(\sigma^*)} \exp \left( \frac{-\Delta G}{kT} \right) \right] = n_0 \exp \left( \frac{-\lambda}{\lambda_0} \right) \quad (1)$$

where  $\nu(\sigma^*)$  is the velocity of the dislocation in the octahedral plane,  $\sigma^*$  the effective stress,  $\nu_0$  an attempt frequency,  $\Delta G$  the cross-slip activation energy. If MK coalescence is neglected (see Ref. [5]),  $n(\lambda)$  represents the MK height distribution, characterized by a ‘standard’ MK height  $\lambda_0$ , given by:

$$\lambda_0 \propto \frac{\nu(\sigma^*)}{\nu_0} e^{(\Delta G/kT)} \quad (2)$$

The standard MK height  $\lambda_0$  is therefore a rapidly decreasing function of temperature.

The lateral motion of a MK is controlled by the motion in the cube plane, along the screw direction, of the bowed out KW [7]. It requires that the Peach-Köhler force on the MK exceeds the force  $F_0$  necessary to drag this bowed out segment.  $F_0$  actually depends on the extension of the KW in the cube plane, but will be taken constant for the sake of simplicity. The MK height  $\lambda$  has therefore to fulfill the condition:

$$\lambda > \lambda_P = \frac{F_0}{\sigma^* b} \quad (3)$$

During its lateral motion, a MK collides and combines with other MKs of both signs. The dislocation thus explores a number of configurations, and can occasionally find one of them in which no MK is larger than  $\lambda_P$ . Such a configuration is sessile. This mechanism can be considered as responsible for a mobile dislocation exhaustion.

In order to allow a constant strain rate deformation, this exhaustion must be compensated by some multiplication, replacing the lost dislocations by new ones. It can be reasonably assumed that a MK of height  $\lambda$  such that:

$$\lambda > \lambda_M \approx \frac{\mu b}{\sigma^*} \quad (4)$$

will bounce out in the octahedral plane. During its expansion, it can either be trapped by preexisting immobile KW and form KW dipoles (or occasionally annihilate with these KW), or overcome their attractive interaction and develop as new dislocation loops by a “Frank-Read like” multiplication mechanism. The new dislocation loops thus produced can lock at places into KW segments during their expansion in the octahedral plane, thus restoring continuously the MK height distribution of equation (1).

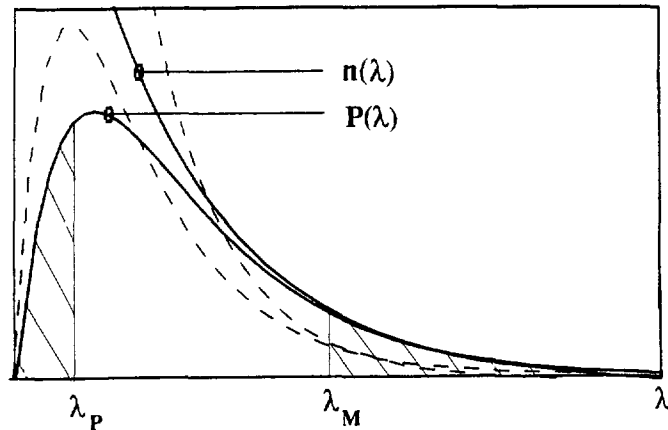


Fig. 1. — Height distribution of a MK  $n(\lambda)$  and of the highest MK on a dislocation  $P(\lambda)$ . A constant mobile dislocation density requires that the two hatched areas are equal. The dashed curves correspond to a higher temperature than the solid ones.

The related multiplication rate  $\Pi_M$  per unit dislocation length is proportional to the number of MKs larger than  $\lambda_M$ :

$$\Pi_M \approx \kappa \int_{\lambda_M}^{\infty} n(\lambda) d\lambda = \kappa n_0 \lambda_0 e^{\left(\frac{-\lambda_M}{\lambda_0}\right)} \quad (5)$$

as seen from equation 1, where  $\kappa$  is a “multiplication efficiency” factor, taking into account both the number of loops emitted at a multiplication site, and the possibility of killing a source by KW annihilation or mutual trapping.

The exhaustion rate is slightly more complicated to obtain. Indeed, the condition for moving a MK is given by equation (3), but the condition for unzipping a superdislocation is that *at least one* MK is larger than  $\lambda_P$ . Conversely, if a dislocation gets in such a configuration that *none* of its MKs is larger than  $\lambda_P$ , it becomes definitively sessile. Each mobile dislocation will sooner or later get in such a sessile configuration, which results in a continuous exhaustion of the mobile dislocation population, as mentioned above. The exhaustion rate can be written:

$$\Pi_E \propto \int_0^{\lambda_P} P(\lambda) d\lambda \quad (6)$$

where  $P(\lambda)$  is the height distribution of the *largest* MK on dislocations.  $P(\lambda)$  is close to  $n(\lambda)$  for large  $\lambda$  values, but departs significantly from it for small  $\lambda$  values, since the probability that a small MK is nevertheless the larger on the dislocation falls down to zero with the MK size  $\lambda$  (see Fig. 1).

### 3. Strength Anomaly and Other Characteristic Features

Though the detailed calculations of  $P(\lambda)$  and  $\Pi_E$  and the analytical derivation of the strength anomaly are given in reference [9], qualitative conclusions can be simply derived here as follows. Constant strain rate deformation requires that the mobile dislocation density is maintained constant, i.e. that the exhaustion rate  $\Pi_E$  is balanced by the multiplication rate  $\Pi_M$  [1].

This is shown graphically in Figure 1, where the ratio of the two hatched areas corresponding respectively to exhaustion and multiplication is constant. An increase of temperature yields a fall off of the standard flight distance  $\lambda_0$ , which in turn flattens the  $n(\lambda)$  and  $P(\lambda)$  distributions, mainly in the range of large  $\lambda$  values. The only possible way to restore the balance of the hatched areas is to shift  $\lambda_P$  and  $\lambda_M$  toward the left, i.e. to increase the stress (see Eqs. (3) and (4)).

At constant applied stress (creep tests) on the opposite, the previous balance between the hatched areas has no reason to be maintained. At low applied stresses (creep conditions), the  $\lambda_P$  and  $\lambda_M$  values are shifted toward the right as compared to the constant strain rate steady state of Figure 1. Exhaustion thus prevails upon multiplication. The creep rate slows down continuously, as experimentally observed [8] (exhaustion creep).

As in the LPU case [1], the multiplication and exhaustion processes that are individual dislocation mechanisms follow first order kinetics, and the resulting steady state is independent of the mobile dislocation density. The consequences on the mechanical properties are therefore the same here as they are in the LPU case. In particular, the steady state being independent of the mobile dislocation density, and characterized by a single value of stress for a given temperature, the strain rate sensitivity is expected to be zero. In a same way, the continuous storage of exhausted dislocations leads to a strong work hardening rate, as observed experimentally. The observed flow stress reversibility also results from the fact that the steady state effective stress only depends on temperature: a specimen slightly<sup>(1)</sup> deformed at a temperature  $T_1$ , unloaded, and subsequently loaded at  $T_2$ , will deform at the same stress than a specimen directly deformed at  $T_2$  (Cottrell-Stokes experiment).

The present model is based on mechanisms fairly similar to the ingredients of Mills and Chrzan simulations [4]. Though these simulations were performed at imposed stress, and necessarily limited in strain, both approaches lead to the same conclusion that, at a given temperature, exhaustion is dominant at low stress levels and multiplication occurs at high stress levels. In addition, and despite the roughness of the present approach, its statistical and "multi-dislocation" character allows it to account for both the stress anomaly and most of the associated specific features observed during constant imposed strain rate experiments.

#### 4. Conclusion

The present paper is a brief outline of the 'E.L.U.' model thoroughly worked out in reference [9]. In this simplified statistical approach of the behaviour of the mobile dislocation population in L1<sub>2</sub> alloys in the anomalous domain, superdislocations evolve through a variety of macrokink configurations, some of which being sessile. The resulting exhaustion is balanced by multiplication in order to cope with the constant applied strain rate. Though the dislocation mechanisms are significantly different from those responsible for the strength anomaly in L1<sub>0</sub> alloys [1], the results are formally equivalent. As temperature is raised, exhaustion is enhanced and multiplication becomes more difficult, unless stress is increased. Other consequences, as a low strain rate sensitivity, a large work hardening rate and an almost reversible flow stress, are derived both from the thermally activated exhaustion and from the independence of the steady state on the microstructure.

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<sup>(1)</sup> the lower the strain, the closer the flow stress to the effective stress

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