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Abstract

The stress decrease above the stress peak temperature (750 K) is studied in <123> single crystals of Ni₃(Al,3at%Hf). Two thermally activated deformation mechanisms are evidenced on the basis of stress relaxation and strain rate change experiments. From 500 to 1070K, the continuity of the activation volume/temperature curves reveals a single mechanism of activation enthalpy 3.8eV/atom and volume 90b³ at 810 K with an athermal stress of 330MPa. Over the very same temperature interval, impurity or solute diffusion towards dislocation cores is evidenced through serrated yielding, peculiar shapes of stress strain curves while changing the rate of straining and stress relaxation experiments. This complicates the identification of the deformation mechanism which is likely connected with cube glide. From 1070K to 1270K, a high temperature mechanism has an activation enthalpy and volume of 4.8eV/atom and 20 b³ respectively, at 1250 K.

Key words:

intermetallics –dislocation core – flow stress – work-hardening – thermal activation of deformation

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1. Introduction

The strength of Ni_3Al compounds has been extensively studied and modelled in the temperature domain that corresponds to the anomalous flow stress behaviour with respect to temperature (see reviews : [1-3]). However, above the temperature T_p at which the stress peaks, information is much less abundant. Early papers report a marked decrease of stress as temperature increases [4], an increase of stress for increasing strain-rates [5], no stress asymmetry in tension and compression tests [6]. Such observations are the signature of thermally activated mechanisms in which the dislocation interaction with stress is less complex than in the stress anomaly domain. Slip trace analysis and microstructural features in single slip oriented crystals are also available, that will be recalled in section 4.3. Therefore, the point of this study is to contribute to the characterisation of the plasticity mechanisms which take over in Ni_3Al above T_p .

The material is presented as follows: first, mechanical test data are exposed with emphasis on the flow stress as a function of temperature at various strains. Then, the strain-rate effect on stress is investigated through stress relaxation and strain-rate change experiments. The activation volume is examined as a function of temperature, stress and strain, so as to determine over which domains specific mechanisms operate. Activation energies are determined as well. All the results are discussed in terms of the possible rate controlling processes that underlie the loss of strength under the present conditions, taking into account published micro structural observations.

2. Experimental procedures

2.1 Materials

$\langle 123 \rangle$ oriented single crystals were used of composition $\text{Ni}_3(\text{Al}, 3\text{at}\% \text{Hf})$. Previous studies were performed on the same crystals such as the determination of fault energies in dislocation cores [7], the connection between flow stress and dislocation densities [8] or the evaluation of the stress component related to the long range dislocation interactions through dip tests [9].

2.2 Mechanical tests

Specimens 3.5 x 3.5 x 7 mm³ were cut with a diamond saw and subsequently mechanically polished. Compression tests were performed on a Schenck RMC 100 machine at a shear strain-rate $\dot{\gamma}$ close to 10⁻⁴ s⁻¹ (unless otherwise specified) between 670 and 1270K, under a helium atmosphere. τ and γ are respectively shear stresses and strains, resolved along the primary octahedral system. $\dot{\gamma}_p$ is the plastic shear strain rate.

Strain-rate change tests as well as single and repeated stress relaxation experiments [10] were performed. They provide the apparent activation volume V_a which expresses the stress dependence of the strain rate, as well as the microscopic activation volume, which characterises the stress dependence of the dislocation velocity [11]. Several such transient tests were performed along monotonic curves, at strains distant enough not to disturb the general shape of the curve.

3. Results

3.1. Deformation curves.

Typical stress strain curves exhibit a high work-hardening rate in the preplastic stage which decreases gradually to a moderate value after the yield stress. Then, it remains approximately constant up to a 18% strain. The general aspect of the curves stays the same for increasing temperatures.

The stress $\tau_{0.2}$ which corresponds to a 0.2% plastic shear strain is first considered and represented on Fig.1a as a function of temperature.

[Insert Figure 1 about here]

Two sets of data points are shown on Fig.1a : those that correspond to the present study (670 to 1270K), together with earlier data from [12] between 120 and 800K. The latter were obtained on crystals cut from the same rod, with the same orientation and deformed at the same strain rate with the present set up. Fig.1a shows that the two sets fit rather well together. Other results below confirm this fair agreement. As the temperature increases, the stress reaches a maximum for $T_{p0.2}$, with $673 \text{ K} < T_{p0.2} < 773 \text{ K}$. It then decreases till T=1100K. This corresponds to a thermally activated *deformation mechanism at intermediate temperatures*. At higher temperatures, $\tau_{0.2}$ decreases more steeply with increasing temperature thus revealing a transition towards a *high temperature mechanism*. The work hardening rate measured at the same plastic strain is plotted on Fig.1b as a function of temperature. It shows a moderate decrease of this parameter with temperature up to 1150K, then a more abrupt decrease.

This is thought to correspond to the transition between the intermediate and high temperature mechanisms.

3.2. Strain-rate dependence on stress.

This dependence can be characterised by the apparent activation volume V_a which is defined by:

$$V_a = kT (\partial \ln \dot{\gamma} / \partial \tau)_T \quad (1)$$

where k is the Boltzmann's constant, T the absolute temperature. The importance of this parameter will be illustrated below. Experimentally, V_a can be determined, provided a change in strain-rate $\Delta \ln \dot{\gamma}$ and the corresponding change in stress $\Delta \tau$, are known from the same experiment. The latter can be a stress relaxation test, remembering that the stress-rate is proportional to the plastic strain-rate. Therefore, measuring the stress-rate at 2 points along the relaxation curve allows one to obtain V_a , using relation (1). It can also be determined through a strain-rate jump experiment, in which the strain-rate is suddenly increased and the corresponding change in stress is measured. Both methods have been used extensively in the present study.

3.2.1. Stress relaxation tests.

[Insert Figure 2 about here]

The results are presented on Fig.2 where V_a is measured at a resolved plastic strain $\gamma_p = 0.2 \times 10^{-2}$. It is first plotted as a function of stress (Fig. 2a) then of temperature (Fig.2b), for stresses below 420 MPa and temperatures above 700 K. On Fig.2b data by Spätig et al [12] are added for comparison. They refer to lower temperatures as compared to the present study, i.e. to the anomalous behaviour of stress with respect to temperature. Fig.2b shows that the two data sets fit satisfactorily together (see V_a values around 680 K and 770 K). In addition, they seem to follow the same $V_a(T)$ curve.

Fig.2 exhibits the classical features of the $V(\tau)$ or $V(T)$ curves as different thermally activated mechanisms take over following changes in stress or temperature [13]. More precisely, Fig.2 shows that for stresses decreasing from 420MPa, or temperatures above 750K (see data points 2 and 3), the volume is approximately constant and close to $90b^3$ (where b is the Burgers vector of the superpartial dislocation). This corresponds to the first thermally activated deformation mechanism above T_p , i.e. the *intermediate temperature mechanism*. Data points 4 and 5 on Fig.2 show that, as the stress gets close to 330MPa and the temperature close to 1070K, V_a

suddenly increases to very high values. This is the signature of a change in dislocation motion mechanism, the *intermediate temperature mechanism* becoming athermal, with a long range elastic interaction stress of about 330 MPa and an athermal temperature of 1070K. This sudden increase in V_a is justified in the appendix. This change was also illustrated in section 3.1 on the curve of Fig.1 a : it shows that above 700 K, the stress decreases and reaches some kind of plateau in the vicinity of $\tau=330$ MPa and $T=1070$ K. This example illustrates once more that the athermal stress and temperature of a given mechanism are more accurately determined on respectively the $V_a(\tau)$ and $V_a(T)$ curves as compared to the $\tau(T)$ curves [13].

After this discontinuity of the activation volume, data points 6 and 7 on Fig.2 indicate that V_a exhibits a low value of the order of $20 b^3$ and is approximately constant from 300 to 170 MPa or 1170 to 1270K. It corresponds to the *high temperature* thermally activated *mechanism* in agreement with the decreasing $\tau(T)$ curve of Fig.1, at and above 1150 K (points 6 and 7).

3.2.2 Strain-rate change experiments.

To compare the activation volumes measured by stress relaxation and strain-rate change techniques, the latter experiments have been performed at plastic strains between 10×10^{-2} and 16×10^{-2} for which the stress jump is more easily measured. The sample response to an increase of strain-rate (by a factor of about 5) is observed to depend on temperature: at 1073K, for instance, the $\tau(\gamma)$ curve exhibits, after the jump, a rounded yield point of moderate amplitude. It can be attributed to the activation of dislocation sources as the strain-rate is suddenly increased. Then some kind of new steady state is achieved, that corresponds to a slightly lower stress. Conversely, at 1173K, a normal type of response is observed, i.e. the stress increases monotonically following the increase of strain-rate. These two types of transients will be discussed in section 3.3.

The good agreement between V_a values measured by stress relaxation and strain-rate change techniques, at a plastic strain close to 10×10^{-2} , is illustrated on Fig.3 as a function of temperature. V_a values are alike at 873K and 1073K. In addition, both sets of data seem to follow the same $V_a(T)$ curve.

In order to better determine the temperature domains associated to given mechanisms, the stress measured under the same conditions is also represented on the same figure. At temperatures corresponding to strain-rate jump experiments, the stress measured at the higher strain-rate is also plotted on Fig.3. This figure shows that the stress corresponding to the lower strain-rate, as well as V_a , exhibit trends as a function of temperature similar to those at lower strains (see Fig.1 and Fig 2(b) for comparison). At 673 K, $\tau_{10\%}$ is increasing with temperature and strain-rate insensitive, while V_a

decreases. These features are well known for the temperature domain that corresponds to the strength anomaly, see e.g. [12]. The temperature of maximum stress is T_{p10} , with $780 \text{ K} < T_{p10} < 900 \text{ K}$, under such conditions and the stress is slightly strain-rate sensitive at 773 and 873 K. The uncertainty on T_{p10} is bound to the shallow maximum of the curve. In particular, it is not possible to say whether T_{p10} is higher or lower than $T_{p0.2}$. Between 873 and 1073 K, the stress decreases and becomes strain-rate dependent. This corresponds to the intermediate temperature mechanism, with an activation volume smaller than or of the order of $200 b^3$.

[Insert Figure 3 about here]

Around 1073 K a marked change in slope of the $\tau(T)$ curve is observed while V_a increases towards large values. These features are the signature of a change in mechanism with an athermal temperature of about 1070 K for the intermediate temperature mechanism. The corresponding athermal stress (on the $\tau(T)$ curve) is 490MPa. Above 1073 K, the high temperature thermally activated mechanism is also observed on Fig.3. The stress decreases steeply for increasing temperatures and is strain-rate dependent. V_a is very low (close to $20 b^3$) and constant with temperature.

3.3 Evidence of diffusive effects

Under adequate temperatures corresponding to slow diffusion, dislocations moving slowly or arrested at obstacles may attract solute atoms or impurities towards their cores. This results in slowing down or pinning the slow dislocations with no effect on the fast ones. These mechanisms are sometimes reported at specific temperatures, in only a few of the numerous studies available on the plasticity of Ni₃Al compounds. For example, some of the curves presented by Demura and Hirano [14] show evidence of serrated yielding in single crystals of stoichiometric Ni₃Al between 290 and 530 K for a strain-rate of $8.3 \times 10^{-6} \text{ s}^{-1}$. Similarly, in the early work of Thornton et al. [4], a zero or negative strain rate sensitivity of the stress is measured in similar compounds. Though not much attention is generally paid to these effects they are likely to influence somehow the mechanical behaviour over some temperature range. The conditions of observation of these effects are first listed and then the temperature range over which they take place is determined.

They are thought to account for the two different types of observations during the strain rate jump experiments of section 3.2.2. At high enough temperatures (e.g. 1173K), diffusion is fast enough for impurity or solute clouds to move along with dislocations. Imposing a higher straining rate increases the stress regularly so as to increase the dislocation velocity accordingly. At lower temperatures (such as 1073K), the smooth yield point that follows the jump suggests the following interpretation: at

low strain rate, few dislocations are proceeding slowly (impurity or solute pinning) while during the jump, an excess of stress is temporarily required for unpinning to activate dislocation sources and increase their velocity.

Serrated yielding along the stress strain curves is also observed at intermediate temperatures for plastic strains larger than about $5 \cdot 10^{-2}$. This effect is present with or without performing stress relaxation experiments during straining. An example of serrations is shown on Fig.4, at 673K. It is thought that such a behaviour is due to dynamic strain ageing as in metallic alloys.

[Insert Figure 4 about here]

A peculiar aspect of some repeated stress relaxation curves illustrated on Fig.5 has also been observed. Usual stress relaxation series exhibit a stress drop for each relaxation which decreases along the series, at least in the case of hardening [11].

[Insert Figure 5 about here]

This is also observed here, on Fig.5. However, the strain-rate at the end of a usual relaxation is commonly found to be smaller than the strain-rate at the onset of the following one. This is so because the sample is reloaded after each relaxation, a higher stress inducing a higher deformation rate. In the present case, the table of Fig. 5 displays these various rates, with the onset ones in the left column. It shows a different trend: the plastic resolved strain-rate $\dot{\gamma}_p$ in (1') is slightly larger than in (2), larger in (2') than in (3) and larger in (3') than in (4). This is opposite to the usual behaviour and reflects a negative strain-rate sensitivity. This can be explained by considering that during reloading (e.g. between (1') and (2)) diffusive effects pin practically immobile dislocations, so that the following relaxation starts with a lower mobile dislocation density. In particular, this effect prevents the determination of the microscopic activation volume though this is straightforward in the case of conventional relaxation series [15].

At last, Fig. 5 shows a yield point during constant strain-rate reloading just after the relaxation series. This feature is again not commonly observed. It is thought to correspond here to an excess stress necessary for dislocation multiplication or unpinning, subsequently to dislocation exhaustion during the transient test. The temperature domain where these various effects are observed is also shown on Fig. 3. They operate essentially from 673 to 1073K.

The diffusive processes in question can consist of i) solute atom or impurity diffusion to the superdislocation core (Cottrell effect) ii) the relaxation of the atomic

structure along the antiphase boundaries (APB) [16] and iii) the change of APB plane by climb. These three processes affect similarly the dislocation velocity. The present study also shows that they are more marked at larger strains as more dislocations are involved. They operate at temperatures high enough to have diffusion over short distances but low enough to prevent fast diffusion.

4. Discussion

4.1 One or two mechanisms above T_p ?

In sections 3.1 and 3.2, the temperature (and stress) domains that correspond to two thermally activated mechanisms have been evidenced above T_p considering both the changes in slope of the $\tau(\gamma)$ curves and the discontinuities of V_a . These two mechanisms are now critically examined in the light of the present data as well as earlier results.

Among the few available studies of Ni₃Al compounds above T_p , Heredia and Pope [17, 18] report only one monotonic decrease of stress for temperatures rising above T_p . In these two papers, single crystals of binary Ni₃Al and ternary compounds, containing from 0.2 to 1 at% B or 1 to 3.3 at% Hf or 1 and 2.5 at% Ta, were deformed in air at a rate of 10^{-3} s^{-1} up to 1350K. Conversely, in a deformation study of Ni₃(Al, 1at%Ta) single crystals in the $\langle 123 \rangle$ orientation, Baluc et al [19, 20] report $\tau_{0.2}(T)$ curves quite similar to those of Fig.1: above T_p they exhibit two temperature domains with different rates of stress decrease with temperature. Moreover, as the strain rate is changed from $7 \times 10^{-5} \text{ s}^{-1}$ to $7 \times 10^{-4} \text{ s}^{-1}$, the stress is observed to increase in both temperature domains [19], similarly to Fig.3. The difference between the two types of data sets can be explained if one compares the experimental procedures in the two studies. The former authors deform successively the same sample up to a 0.2% strain at each investigated temperature. Conversely, in the present study and in [19], a new sample is tested at each temperature. It has been shown by Spätig et al. [12] that only the latter procedure is reliable. Therefore the existence of two successive thermally activated processes operating above T_p seems to be established for Hf and Ta containing compounds.

4.2 Activation energies

The identification of both mechanisms is extended considering activation energy values and available substructure data.

Activation enthalpy values can be obtained from the present results using the relation:

$$\Delta H = -TV_a \left(\frac{d\tau}{dT} \right)_\gamma \quad (2)$$

Determination of ΔH has been attempted at temperatures corresponding to minima in the $V_a(T)$ curves, i.e. as far as possible from the athermal conditions and at a plastic strain of 0.2% where data are more accurate (Fig.1 and 2b). ΔH is found to be close to 3.8 eV/atom at 810K (intermediate temperature mechanism) and 4.8 eV/atom at 1250K (high temperature mechanism). It is worth noting that these results differ from those of a preliminary study [21] due to more accurately determined V_a values in the present work.

4.3 Rate controlling mechanisms

Table 1 summarises the characteristics of the deformation mechanisms identified in the present study.

[Insert Table1 about here]

The continuity of the $V(T)$ curve illustrated on Fig.2 and 3 suggests unambiguously that a single intermediate temperature mechanism operates from 500K to 1050K (for $\gamma_p = 0.2 \cdot 10^{-2}$), i.e. below and above T_p respectively. The corresponding activation energy is 3.8eV/atom and at temperatures T_{ath} close to 1050K, it becomes athermal and the stress value τ_{ath} of about 340MPa represents the long range elastic interaction of the mobile dislocations with the substructure.

For the intermediate temperature conditions around T_p , transmission electron microscope observations and slip line studies reveal dislocations gliding on the cube planes (see e.g. [1, 22] for the present samples and the in situ experiments on Ni₃(Al, 0.25at%Hf) in [23]). These observations reveal friction forces acting on both screws and edges, due to their core sessile with respect to cube glide. It is also remarkable to note that the temperature range where diffusive effects operate coincides with that of the intermediate temperature mechanism (Fig.3). Several hints about the operation of such processes have appeared throughout this study in the various experiments performed. They consist successively in a peculiar aspect of the $\tau(\gamma)$ curve after a strain-rate change (section 3.3.2), the occurrence of serrated yielding on the monotonic curves with or without transient tests (Fig.4), of a yield point of small amplitude at the onset of reloading after a repeated stress relaxation experiment (Fig.5), or an unusual behaviour of stress relaxation curves (Fig.5) depending on the conditions. The present observation of the intermediate temperature interval that includes T_p is quite remarkable, It implies that a same mechanism which is thermally activated operates over the stress peak and

below it, at least under relaxation conditions. Assuming the classical decomposition of the applied stress ($\tau = \tau^* + \tau_i$ –see equation A3 in the–appendix), the present results indicates that the stress anomaly would be related to the behaviour of the athermal part of the stress τ_i . Fig.6 illustrates the corresponding variation of τ , τ^* and τ_i for the intermediate and high temperature mechanisms respectively. This constitutes a milestone in the study of the role of thermal activation in the deformation of Ni₃Al which started more than 15 years ago (see [24] and e.g. [3]).

At higher temperatures, as the diffusive processes above do not impede anymore dislocation motion, from about 1070K the mechanism which takes over has an activation enthalpy of 4.8eV/atom, an activation volume smaller than or equal to $20b^3$ and does not become athermal before 1300K, the maximum temperature explored. Some TEM observations report “exotic” dislocation configurations [25] consisting of i) two superpartials, each being extended outside the cube plane ii) a split configuration along a dihedron. $\langle 100 \rangle$ dislocations have also been seen in Ni₃Al(1at%Ta) at 1272K [26, 27].

It is worth noting that two mechanisms have also been mentioned on the basis of creep experiments [28] in a Ni₃(Al,Hf,B) compound. The intermediate one is observed between 903 and 923K with no activation energy reported. At 1273K, a second mechanism takes over with an activation energy of 4.96 ± 0.56 eV/atom that the authors identify as dislocation climb on the basis of diffusion data. It is clear that TEM observations are requested on the samples deformed in the high temperature domain to identify more precisely the operative dislocation mechanism.

5. Conclusions

In Ni₃(Al,3at%Hf) single crystals, the loss of strength above the stress peak temperature is due to the thermal activation of two different processes. This is clearly established in the present compound, as well as in Ni₃(Al, 1at%Ta) both in the $\langle 123 \rangle$ orientation, on the basis of the variation with temperature of the stress and the activation volume.

Between 550 and 1070 K for $\gamma_p = 0.2 \times 10^{-2}$ ($T_p \sim 750$ K), a single intermediate temperature mechanism is evidenced with an activation volume and enthalpy of respectively $90b^3$ and 3.8eV/at at 810 K. It seems to be connected with cube glide. It operates exactly over the domain where diffusive processes are evidenced through various types of experiments. This complicates the further identification of the rate controlling process. The latter becomes athermal at 1070K with an internal stress of 330MPa. Between 750K and T_p , under relaxation conditions dislocation segments

proceed under the same intermediate temperature mechanism than slightly above T_p . This observation sheds some new light on the upper end of the stress anomaly domain.

From 1070K to 1270K, a high temperature mechanism takes over with an activation volume smaller than or equal to $20b^3$ and an activation enthalpy of 4.8eV/at at 1250 K. Its complete identification requires further studies particularly TEM observations of dislocation structures.

It is hoped that the present results on these intermediate and high temperature mechanisms will stimulate further research on the diffusive processes (activation energies, diffusing species) and on high temperature cube glide mechanisms.

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Appendix

Since the early days of mechanical testing, the large increase of the activation volume as a deformation mechanism becomes athermal has been experimentally evidenced. This is observed in a number of crystals deforming under conventional thermally activated processes and for various deformation conditions (constant strain rate tests, creep). Examples can be found in [29] and [30] for Ti, in [31] for Zr, in [32] for Mg. However, the interpretation of this behaviour has never been proposed, to the best of our knowledge. In the following a brief justification is developed.

The plastic strain rate is expressed via the Orowan relation:

$$\dot{\gamma}_p = \rho_m b v \quad (A1)$$

where ρ_m is the mobile dislocation density, b their Burgers vector and v their average velocity.

The velocity v obeys thermal activation and depends on stress and temperature via :

$$v = v_0 \exp(-Q/kT) \operatorname{sh}(\tau^*V/kT) \quad (A2)$$

where Q and V are respectively the activation energy and volume of the velocity and τ^* is the effective stress defined by:

$$\tau^* = \tau - \tau_i \quad (A3)$$

where τ_i is the long range elastic interaction stress. In relation (A2), the hyperbolic sine function, contrary to the more commonly used exponential function, takes into account the dislocation jumps in the direction opposite to the applied stress. These are especially important as the mechanism becomes athermal (weak τ^* values). Under the latter conditions, $\operatorname{sh}(\tau^*V/kT)$ becomes equivalent to τ^*V/kT so that according to (A1) and (A2), $\dot{\gamma}_p$ is proportional to τ^* . Referring to (1), V_a reaches very high values. The value of τ_i that corresponds to a vanishing τ^* is called τ_{ath} in the above.

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Figure captions

Figure 1: a) Stress $\tau_{0.2}$ as a function of temperature. $\dot{\gamma} = 10^{-4} \text{ s}^{-1}$. The data point numbers refer to Fig.2. b) Work hardening rate WHR measured at a plastic strain of 0.2 % as a function of temperature

Figure 2: Apparent activation volumes measured by stress relaxation experiments at $\gamma_p = 0.2 \times 10^{-2}$. a) as a function of stress b) as a function of temperature. Open circles are data by Spätig et al [12]. The data point numbers are the same as on Fig. 1. (b is the Burgers vector of the superpartial dislocation)

Figure 3: Temperature dependence of the stress and the activation volume measured at a 10% plastic strain. V_a values are determined by both stress relaxation and strain-rate jump experiments. Stresses correspond both to the nominal strain-rate and to that strain-rate multiplied by 5. The temperature range of diffusive effects is marked by the horizontal bar. (see text).

Figure 4: Example of serrated yielding along a stress strain curve at 673K. An enlargement of the curve is shown in the insert.

Figure 5: Stress/time curve for a series of 4 successive relaxations performed along a monotonic curve at 873K. $\gamma_p = 10^{-2}$. Points (1) to (4) correspond to the onset of each relaxation curve and (1') to (4') to the end. Plastic strain - rates measured at these points are displayed in the table. Constant strain-rate is applied before point (1) and after point (4'). Note the yield point at reloading ($t \approx 975 \text{ s}$).

Figure 6: Schematic representation of the temperature dependence of τ , τ^* and τ_i for a) the upper end of the intermediate T domain and b) the high temperature mechanism.

Tables

γ_p		T_{int}	HT
0.2 %	τ_{ath}	330 MPa	< 170 MPa
	T_{ath}	1070 K	< 1270 K
	V_{amin}	90 b ³	≤ 20 b ³
	ΔH	3.8 eV/at.	4.8 eV/at.
$\cong 10\%$	τ_{ath}	490 MPa	< 180 MPa
	T_{ath}	1050 K	> 1270 K
	V_{amin}	≤ 200 b ³	$\cong 10$ b ³

Table 1: Characteristic parameters for the mechanisms at intermediate (T_{int}) and high (HT) temperatures respectively

Figure 1: a) Stress $\tau_{0.2}$ as a function of temperature. $\dot{\gamma} = 10^{-4} \text{ s}^{-1}$ The data point numbers refer to Fig.2. b) Work hardening rate WHR measured at a plastic strain of 0.2 % as a function of temperature.
197x147mm (300 x 300 DPI)

Figure 1: a) Stress $\tau_{0.2}$ as a function of temperature. $\dot{\gamma} = 10^{-4} \text{ s}^{-1}$ The data point numbers refer to Fig.2. b) Work hardening rate WHR measured at a plastic strain of 0.2 % as a function of temperature.
209x150mm (72 x 72 DPI)

Figure 2: Apparent activation volumes measured by stress relaxation experiments at $\Gamma = 0.2 \times 10^{-2}$. a) as a function of stress b) as a function of temperature. Open circles are data by Spätig et al [12]. The data point numbers are the same as on Fig. 1. (b is the Burgers vector of the superpartial dislocation)

186x142mm (300 x 300 DPI)

Figure 2: Apparent activation volumes measured by stress relaxation experiments at $\Gamma = 0.2 \times 10^{-2}$. a) as a function of stress b) as a function of temperature. Open circles are data by Spätig et al [12]. The data point numbers are the same as on Fig. 1. (b is the Burgers vector of the superpartial dislocation)

193x134mm (300 x 300 DPI)

Figure 3: Temperature dependence of the stress and the activation volume measured at a 10% plastic strain. V_a values are determined by both stress relaxation and strain-rate jump experiments. Stresses correspond both to the nominal strain-rate and to that strain-rate multiplied by 5. The temperature range of diffusive effects is marked by the horizontal bar. (see text).

187x132mm (300 x 300 DPI)

Figure 4: Example of serrated yielding along a stress strain curve at 673K. An enlargement of the curve is shown in the insert.
201x154mm (300 x 300 DPI)

Figure 5: Stress/time curve for a series of 4 successive relaxations performed along a monotonic curve at 873K. Gamma p= 10-2. Points (1) to (4) correspond to the onset of each relaxation curve and (1') to (4') to the end. Plastic strain - rates measured at these points are displayed in the table. Constant strain-rate is applied before point (1) and after point (4'). Note the yield point at reloading (t = 975 s).

144x134mm (300 x 300 DPI)

Figure 6: Schematic representation of the temperature dependence of τ , τ^* and τ_i for a) the upper end of the intermediate T domain and b) the high temperature mechanism.
89x96mm (300 x 300 DPI)

Figure 6: Schematic representation of the temperature dependence of Tau, Tau* and Taui for a) the upper end of the intermediate T domain and b) the high temperature mechanism.
88x96mm (300 x 300 DPI)