High temperature dislocation climb in the γ' rafts of Single Crystal Superalloys:**The hypothesis of a control by dislocation entry into the rafts**

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Abstract

The good mechanical resistance to high temperature creep of [001] oriented single crystal superalloys is due to the properties of the rafts, i.e. platelets of the $L1_2$ γ' phase embedded in a γ matrix. At temperatures higher than 900°C, the plastic strain of the rafts results from the climb of dislocation pairs with a total $a \cdot \langle 100 \rangle$ Burgers vector and/or from the climb at the γ/γ' interfaces of $a/2 \cdot \langle 110 \rangle$ dislocation segments. This climb motion involves the exchange of vacancies between these dislocations and vacancy sinks such as pores and the specimens' surfaces. In this paper we suggest that the entry of $a \cdot \langle 100 \rangle$ into the rafts requires the overcoming of a threshold stress and show that this hypothesis gives a natural explanation to some of the most salient aspects of their mechanical behavior.

25

26 Introduction

27 During most part of their high temperature creep life (i.e. during stage II), single crystal
 28 superalloys can be described as composites where hard L1₂ platelets (γ' rafts) perpendicular to
 29 the [001] tensile axis are embedded within a soft fcc γ matrix [1,2]. Despite this simplicity and
 30 the high symmetry of the microstructure, the task of building dislocation-based constitutive
 31 laws has been impeded by two major facts: the γ (resp. γ') phase in channels (rafts) does not
 32 behave as the bulk γ (resp. γ') phase, and the stress and plastic strain state of each phase are
 33 generally unknown.

34 Under simplifying hypothesis, it is possible to overcome the second difficulty by measuring the
 35 so-called ‘perpendicular mismatch’, i.e. the relative difference between the lattice parameters
 36 in the [100] or [010] directions $\delta_{\perp} = 2(a_{\gamma'_{[100]}} - a_{\gamma_{[100]}})/(a_{\gamma'_{[100]}} + a_{\gamma_{[100]}})$ [3]. This can
 37 be done in a non-destructive way by hard X-Rays high resolution diffractometry [4]. Thanks to
 38 the very intense beams provided by synchrotron sources the acquisition time of a (200) peak
 39 can be reduced to less than ten seconds [5]: we can determine not only the behavior of both
 40 rafts and channels under steady state conditions, but also their transient plasticity following a
 41 change in load and/or temperature. Within the present paper we shall focus on the high
 42 temperature behavior of the γ' rafts and try to establish basic equations with parameters which
 43 can be fitted to the results of in situ real time tests.

44 At temperatures higher than 900 °C, plastic strain within the γ channels takes place by glide of
 45 $a/2 \cdot \langle 110 \rangle$ dislocations, as soon as the resolved shear stress becomes larger than the
 46 Orowan stress σ_{Or} . These dislocations leave segments trapped at the intersection of their slip
 47 planes and both γ/γ' interfaces on each side of the channel. The increased dislocation density
 48 decreases the algebraic value of $\delta_{\perp}$, increases the $\sigma_{xx} = \sigma_{yy}$ internal stress components, and
 49 decreases the Von Mises stress. If the strain rate of the rafts is nil, the Von Mises stress within
 50 the channels will return to the Orowan stress, and as no new dislocation can enter the channels,
 51 plastic strain will stop. If the strain rate of the rafts is non-zero, the Von Mises stress within the
 52 channels will settle at an equilibrium value slightly larger than the Orowan stress, so that the
 53 strain rates of both phases are equal, and $\delta_{\perp}$ and σ_{xx} are constant. The internal stress
 54 components $\sigma_{xx} = \sigma_{yy}$ will thus be slightly lower than or equal to $\sigma_{zz} - \sigma_{Or}$.

55 The new dislocation segments left at the interfaces later react with those already present to form
 56 a low energy dislocation wall [6-7], with edge dislocation segments having a $a/2 \cdot \langle 110 \rangle$ or
 57 a $a \cdot \langle 100 \rangle$ Burgers vector and perpendicular $\langle 110 \rangle$ or $\cdot \langle 100 \rangle$ orientations. These

reactions require dislocation climb in the interface, which is expected to contribute to the plastic strain, at least in the [001] direction [8-10]. Molecular Dynamics simulations suggest that these interfacial reactions can be further complicated when the interface is somewhat wavy [11]. This is indeed the case, as interface dislocations seem to be associated to ledges of the interface [12].

Plasticity within the rafts is known to take place mainly by climb of pairs of dislocations [13-14] with a total $a' \cdot \langle 001 \rangle$ Burgers vector. Such segments have been observed by TEM bulging from the interface into the rafts. Their climb motion requires the emission of vacancies by $a' \cdot [001]$ dislocations (population 1 sensitive to the tensile stress $\sigma'_{zz} = \sigma_a$) and their absorption by $a' \cdot [100]$ and $a' \cdot [010]$ dislocations [10,15] merged here into a population 2 (sensitive to the compressive $\sigma'_{xx} = \sigma'_{yy}$ stress) for the sake of simplicity. The climb of population 2 dislocations has been referred as S-climb in [10] while the climb of matrix dislocations at the interface was called G-climb. An excess density of vacancies will slow down dislocations of population 1 and interface dislocations and accelerate those of population 2. The diffusion of these vacancies between sources and sinks thus plays an important part in plastic strain. However, as shown in [16] the flux of vacancies between dislocations climbing within the rafts and interface dislocations is too large by several orders of magnitude to be the factor limiting the climb rate.

The climb velocities of dislocations as a function of the stress and temperature, as well as the actual concentration of vacancies, are unknown. However, some features of the creep behavior can be used to test hypothesis on the underlying mechanisms:

- As shown by Epishin et al. in a seminal experiment [10], the low stress high temperature creep results in an increase of the relative volume of pores equal to the plastic strain, i.e. the strain in the transverse directions is zero.
- A size effect, sometimes called the “thickness debit effect” is observed [17]: the stage II creep rate of thin specimens (from 0.5 mm to a few mm) is significantly higher than this of bulk specimens and can be decreased by coating the specimens.
- The dislocation densities measured within the rafts are in the 10^{11} - 10^{12} range [3,18], with a majority of dislocations having a Burgers vector perpendicular to the tensile axis (i.e. belonging to population 2).
- Some hysteresis is observed in the creep behavior: the strain rate can increase for a while after a load increase before it levels, and a limited amount of plastic strain is measured after a drop of the load. [3]

In this paper, we show that these features can be explained if we admit that:

- 1) The plastic strain of the rafts is not directly controlled by the mobility of $a' \cdot \langle 100 \rangle$ dislocation pairs, but by their entry rate into the rafts.
- 2) For both populations, the entry mechanism of dislocations into the rafts requires the overcoming of the same, weakly temperature dependent threshold stress, but is more efficient for population 2, probably because there are more candidates.

In the following, we shall first derive a few basic equations on the evolution of the density of vacancies, their diffusion to pores, and the osmotic force they exert on dislocations. We shall then estimate the contribution of interface climb to the plastic strain of the material. We shall deduce the first consequences of the above hypothesis on the material's behavior under low and high stresses first without taking account of the climb of dislocations at the interface, then allowing for it. We shall then compare these predictions to available experimental data and use this comparison to put constraints on the possible forms of entry laws.

I Vacancies

The equilibrium density of vacancies within the γ' rafts under load at temperature T can be approximately written as:

$$c_0 \approx \frac{1}{\Omega} \cdot e^{-\left(G_0 - \frac{\Omega}{3} \cdot (\sigma'_{xx} + \sigma'_{yy} + \sigma'_{zz})\right) / k \cdot T} \quad (1)$$

where Ω is the atomic volume, G_0 the free energy for vacancy creation, and k the Boltzmann constant. In a material with dislocation-free rafts, a non-equilibrium vacancy concentration c will result in a vacancy diffusion flux φ from or to the pores. Taking φ as constant around a pore, an equilibrium concentration c_0 at the pore surface (i.e. neglecting the Gibbs-Thomson effect), a pore radius r_0 , a concentration c at a distance R from the pore, and a vacancy diffusion coefficient D , the vacancy concentration around the pores will vary as $1/r$, and we can write:

$$\varphi = 4\pi \cdot D \cdot (c - c_0) \cdot \left(\frac{1}{R} - \frac{1}{r_0}\right)^{-1} \approx -4\pi \cdot D \cdot r_0 \cdot (c - c_0) \quad (2)$$

The same flux will change the vacancy concentration within a volume $\approx 8R^3$ of the material:

$$\varphi \approx 8R^3 \cdot \frac{d}{dt} (c - c_0) \quad (3)$$

Thus, in a volume containing a density $\rho_p \approx (2R)^{-3}$ of pores, a supersaturation will decrease with time following approximately an exponential law with a time constant t_0 :

$$t_0 = (4\pi \cdot D \cdot \rho_p \cdot r_0)^{-1} \quad (4)$$

Depending on temperature and on the density and size of pores, this relaxation time may vary from a few seconds to an hour or more.

If now we have a net creation of vacancies within the rafts (volume fraction f) because the plastic strain rates $\dot{\epsilon}_1^p$, $\dot{\epsilon}_2^p$, and $\dot{\epsilon}_{int}^p$ due to the climb of dislocations from populations 1 and 2 and at the interface do not strictly compensate and the rafts' volume is not conserved, these vacancies diffuse through both phases and we get:

$$\frac{d(c-c_0)}{dt} = \frac{f}{\Omega} \cdot (\dot{\epsilon}_1^p + \dot{\epsilon}_{int}^p + \dot{\epsilon}_2^p) - \frac{c-c_0}{t_0} \quad (5)$$

The vacancy supersaturation $c - c_0$ also results in an osmotic force on dislocations [19,10]:

$$\sigma_c = \frac{k.T}{\Omega} \cdot \ln\left(\frac{c}{c_0}\right) \approx \frac{k.T}{\Omega \cdot c_0} (c - c_0) \quad (6)$$

(The $k.T/\Omega$ factor is quite large, and osmotic stresses are of the same order of magnitude as the applied load. Thus c and c_0 can differ only by a few per cent and the logarithmic term can be linearized.) By combining equations (5) and (6) we can also derive a variation law for the osmotic force:

$$\frac{d\sigma_c}{dt} = \frac{f \cdot k.T}{\Omega^2 \cdot c_0} \cdot (\dot{\epsilon}_1^p + \dot{\epsilon}_{int}^p + \dot{\epsilon}_2^p) - \frac{\sigma_c}{t_0} \quad (7)$$

The effective stress on dislocations of population 1 and interface dislocations (2) will thus be $\sigma_1 = \sigma_a - \sigma_c$ ($\sigma_2 = -\sigma'_{xx} + \sigma_c$). Last, under steady state conditions, we expect the Von Mises stress within the channels to be equal to or slightly larger than the Orowan stress σ_{Or} . As there is no net force in the x and y directions on the specimen we also have [3]:

$$\sigma'_{xx} \approx -\frac{1-f}{f} \sigma_{xx} \approx -\frac{1-f}{f} (\sigma_a - \sigma_{Or}) \quad (8)$$

II Climb in the γ/γ' interfaces

As $a/2 \cdot \langle 110 \rangle$ dislocations glide in the γ channels, they leave dislocations segments at the $[110]$ and $[1\bar{1}0]$ intersections between their slip planes and the γ/γ' interfaces. The total rate of arrival of new segments per unit length of interface is $\dot{\rho}_{int} = 2\sqrt{2} \dot{\epsilon}_\gamma^p / a \text{ m}^{-1} \cdot \text{s}^{-1}$ and should be evenly spread between the eight active slip systems. (Figure 1). Half of the segments with a $[1\bar{1}0]$ line (red in figure 1b) move in the interface by a combination of glide and climb with a velocity $+v_{int}$ (bypassing the interface dislocation network), and the other half (blue) move in

the opposite direction (velocity $-v_{int}$). This one-dimensional view of the motion of dislocation segments is analogous to the motion of kinks along a dislocation segment [20]. The dislocation segments (linear density: ρ_+ (m^{-1})) can move until they reach the extremity of a raft in a $l_r/2v_{int}$ average time or meet another segment (density: ρ_-) with the same line direction climbing in the opposite sense and react to form a new dislocation of the interface network (black). In the second case, the average travel distance $1/2\rho_-$ is travelled by both dislocations moving in opposite directions in a time $\tau = 1/4v_{int} \cdot \rho_-$. The density of ρ_+ segments for one climb system varies as:

$$\dot{\rho}_+ = \frac{1}{4} \frac{2\sqrt{2}\dot{\epsilon}_\gamma^p}{a} - 4 \cdot v_{int} \cdot \rho_+ \cdot \rho_- \quad (9)$$

Assuming steady state conditions and $\rho_+ = \rho_-$:

$$\rho_+ = \frac{1}{2} \left(\frac{\dot{\epsilon}_\gamma^p}{a \cdot v_{int} \sqrt{2}} \right)^{1/2} \quad (10)$$

As we have two populations along the $[110]$ and $[1\bar{1}0]$ intersections, the total density of moving interface dislocations $\rho_{int} = 2(\rho_+ + \rho_-) = 4\rho_+$ is:

$$\rho_{int} = 2 \left(\frac{\dot{\epsilon}_\gamma^p}{a \cdot v_{int} \sqrt{2}} \right)^{1/2} \quad (11)$$

There are two γ/γ' interfaces for each period of the microstructure, their contribution to the non-conservative plastic strain rate of the material in direction z is [10]:

$$\dot{\epsilon}_{int}^p = 2 \cdot \dot{\rho}_{int} \cdot \frac{l_r}{2} \cdot \frac{b_z}{\lambda} = 2 \cdot \frac{2\sqrt{2}\dot{\epsilon}_\gamma^p}{a} \cdot \frac{l_r}{2} \cdot \frac{a/2}{\lambda} = \dot{\epsilon}_\gamma^p \cdot \frac{l_r \sqrt{2}}{\lambda} \quad (12)$$

In the first (non-collision) case, and:

$$\dot{\epsilon}_{int}^p = 2 \cdot \rho_{int} \cdot v_{int} \cdot \frac{b_z}{\lambda} = 2 \frac{a}{\lambda} \cdot \left(\frac{\dot{\epsilon}_\gamma^p \cdot v_{int}}{a \sqrt{2}} \right)^{1/2} \quad (13)$$

in the second.

Figure 1: a; Climb in the γ/γ' interface of two (red) dislocation segments left by a perfect dislocation gliding in the γ matrix. b; Climb in opposite directions of segments with different Burgers vectors. c; Strain rate of the γ matrix and at the interface vs. the strain rate of the matrix (left hand scale) and travel length (TL) of the dislocation segments (right hand scale) for a $1.25 \cdot 10^{-9} \text{ m.s}^{-1}$ interface dislocation velocity.

If the strain rate of the γ channels is low, or v_{int} is high, a new interface dislocation segment has enough time to reach the end of a raft before another dislocation gliding in the channel meets the interface. The interface strain rate is proportional to the strain rate of the channels and as the ratio $l_r/\sqrt{2} \cdot \lambda$ is larger than one the interface contribution to the overall plastic strain is much larger than the $(1 - f) \cdot \dot{\epsilon}_\gamma^p$ contribution of the channels. Last, all segments pile up at both ends of the rafts and cannot create an interface network. On the contrary, if the strain rate of the γ channels is high, or v_{int} is low, the travel distance of the new segments becomes much shorter than half the raft's length, and an interface dislocations network can form. The interface strain rate is then proportional to the square root of the strain rate of the channels.

Under usual strain conditions, an interfacial dislocation network forms during rafting, for strain rates $\dot{\epsilon}_\gamma^p$ in the 10^{-5} s^{-1} range. After rafting, even under strain rates in the 10^{-7} range (100 MPa, 1090°C), our experiments show variations of the perpendicular mismatch $\delta_\perp$ i.e. an increase of the density of the interface dislocation network: we are still in the segment collision regime. The travel distance $1/2\rho_- = 2/\rho_{int}$ of the climbing segments must then be shorter than half the raft's length. Using equation 11 this gives an upper limit on the segments' climb velocity v_{int} :

$$v_{int} \ll \frac{\dot{\epsilon}_\gamma^p \cdot l_r^2}{4\sqrt{2}a} \approx 1.25 \cdot 10^{-9} \text{ m.s}^{-1} \quad (14)$$

This maximum value is smaller than the estimated climb rate of $a \cdot [001]$ dislocation pairs within the rafts under the same conditions. If the above analysis is correct, the difference might be explained by the different core structure and by the necessity for moving interface dislocations to bypass the existing dislocation network, and the associated fluctuations of the stress field and of the interface position. Besides the high ratio between the strain rates of the interface and the channels another surprising feature is the $t_r = \rho_{int}/\dot{\rho}_{int}$ relaxation time necessary to reach an equilibrium concentration of dislocations within the interface. Assuming a constant $1.25 \cdot 10^{-9} \text{ m.s}^{-1}$ velocity, this time would be about 1000 seconds at $\dot{\epsilon}_\gamma^p = 10^{-7} \text{ s}^{-1}$, and 300 seconds at $\dot{\epsilon}_\gamma^p = 10^{-6} \text{ s}^{-1}$: the interface strain rate should increase linearly with time then settle at a constant value.

III Entry and propagation of dislocations within the rafts

204 Basic model of plasticity (without climb at the interface)

205 In this first attempt, we do not at first seek physical realism, but rather to show clearly what
 206 may be expected if the entry of dislocations into the rafts requires the overcoming of a threshold
 207 stress σ_t . We shall first neglect the contribution of dislocations climbing at the interface and
 208 assume the same linear dislocation velocity law for both populations:

$$209 \quad v_d = b_i \cdot \sigma_i \quad (15)$$

$$210 \quad \text{with:} \quad b_i = 5.10^{-16} \text{ m.s}^{-1} \cdot \text{Pa}^{-1}$$

211 We shall also assume the same linear entry law for dislocations for both populations:

$$212 \quad \dot{\rho}_{mi} = a_i \cdot (\sigma_i - \sigma_i^t) \text{ if } \sigma_i > \sigma_i^t \text{ and zero otherwise.} \quad (16)$$

$$213 \quad \text{With:} \quad a_1 = a_2 = 200 \text{ dislocations.m}^{-2} \cdot \text{Pa}^{-1} \cdot \text{s}^{-1}$$

$$214 \quad \sigma_i^t = 70 \text{ MPa}$$

215 These a_i and b_i parameters were chosen so that under realistic experimental conditions the
 216 travel time of a dislocations along a raft's length is in the 100 seconds to 1000 seconds range,
 217 and the density of climbing dislocations is 10^{11} m^{-2} to 10^{12} m^{-2} . The dislocations are assumed
 218 to enter the rafts (length l_r) at one end, climb to the other end and annihilate after a l_r/v_d travel
 219 time. In order to show the behavior of the material during transients, we choose to simulate
 220 experiments with a stepwise loading with 25 MPa increments every 500 seconds, and a load
 221 drop to 100 MPa at $t = 6500$ seconds. Each load increment is supposed to be followed by a
 222 burst of plastic strain within the γ channels. The Von Mises stress then returns to the Orowan
 223 stress following an exponential with a fixed 250 seconds relaxation time. σ_{xx} will increase after
 224 a loading step and the internal stress within the rafts $-\sigma'_{xx}$ will then increase to a new
 225 equilibrium value (equation 8). As the variations of the osmotic stress following equation (7)
 226 can be quite sharp we use a variable time increment so that the difference in the vacancy
 227 concentration between two steps is less than 0.2% of the equilibrium concentration. The other
 228 parameters of the simulation are:

$$229 \quad \text{Rafts' lengths} \quad l_r = 5.10^{-6} \text{ m}$$

$$230 \quad \text{Period of the microstructure} \quad \lambda = 5.10^{-7} \text{ m}$$

$$231 \quad \text{Burgers vector} \quad b = a = 3.65.10^{-10} \text{ m}$$

$$232 \quad \text{Temperature} \quad 1300 \text{ K}$$

$$233 \quad \text{Equilibrium vacancy concentration} \quad c_0 = 5.10^{23} \text{ m}^{-3} (\approx 6.10^{-6}/\text{atomic volume})$$

Relaxation time of vacancies 10 s (arbitrary)

Figure 2: Evolution of the stresses (a), strain rates (b), and dislocation densities (c) during a simulated stepwise loading followed by unloading to 100 MPa. (No climb at the interface.) The thicker lines in domain 2 (a) result from quick variations of the supersaturation even during very short time steps.

262 As it can be checked in Figure 2, for increasing stresses, three domains may be distinguished.
 263 For applied loads σ_a (black) lower than the threshold stress σ_t (orange) (domain 0, from $t=0$ to
 264 $t=1500$ s) no dislocations enter the rafts, the vacancy concentration is at equilibrium and the
 265 osmotic stress (green) is zero. The strain rate for both populations is naught.

266 Between 1500 seconds and 3000 seconds (domain 1) σ_a is larger than σ_t and dislocations of
 267 population 1 can enter the rafts. Each increase of σ_a results in a burst of dislocations entering
 268 the rafts, and a sharp increase of $\dot{\epsilon}_{zz}^p = \dot{\epsilon}_1^p$. As a result, the osmotic force increases abruptly,
 269 and the effective stress on the dislocations of population 1 σ_1 falls under σ_t . New dislocations
 270 can no longer enter the rafts, but those already present go on climbing (at about 3.10^{-8} m.s⁻¹),
 271 and the plastic strain goes on at a lower rate. When the dislocations reach the ends of the rafts
 272 the $\dot{\epsilon}_1^p$ strain rate decreases, and the osmotic force decreases until σ_1 stabilizes at a level slightly
 273 larger than σ_t . As σ_2 remains lower than σ_t , the dislocations of population 2 are inactive.

274 Between 3000 seconds and 6500 seconds (domain 2), both populations 1 and 2 are active. A
 275 step increase of σ_a also results in a short burst of plastic strain for population 1, but as
 276 dislocations of population 2 are present and can absorb some of the newly created vacancies,
 277 the increase of σ_1 is smaller. As $-\sigma'_{xx}$ (light blue) as well as σ_2 (dark blue) gradually increase,
 278 the entry rate of dislocations from population 2 also increases, until σ_2 becomes nearly equal to
 279 σ_1 and $\dot{\epsilon}_2^p$ to $\dot{\epsilon}_1^p$. Between two σ_a steps, σ_c continuously decreases, but its values just after or
 280 just before a step increase linearly with σ_a . Last, as it can be checked from Figure 2b, the
 281 transition from domain 1 to domain 2 results in an increase of the strain rate $\dot{\epsilon}_1^p$ by several orders
 282 of magnitude.

283 The unloading at 6500 seconds results in a sharp drop of σ_1 (red arrow) and of the velocities of
 284 dislocations of population 1. As vacancies are still absorbed by population 2 dislocations, a
 285 drop of the osmotic stress follows, so that σ_1 and σ_2 become equal again at a level smaller than
 286 before $t=6500$ seconds, but larger than σ_t . Dislocations of both populations can still enter the
 287 rafts, but at a lower and decreasing rate. ($-\sigma'_{xx}$ decreases.) The densities of dislocations of both
 288 populations decrease, as dislocations which entered the rafts before 6500 seconds gradually
 289 reach the ends of the rafts and annihilate. When both σ_1 and σ_2 reach the σ_t threshold, σ_1
 290 remains stuck there while σ_2 goes on decreasing with $-\sigma'_{xx}$. The dislocations from population
 291 2 stop entering the rafts, and only those already present will climb with a decreasing velocity
 292 (as $-\sigma'_{xx}$ decreases) until they reach the ends of the rafts and the state of the material returns
 293 to domain 1.

As seen above, even under simple assumptions, the introduction of a threshold stress for dislocations' entry into the rafts makes the material's behavior quite complicated.

Basic model of plasticity (with climb at the interface)

We shall now consider the effect of interface climb of matrix dislocations through the pre-existing interface network. From equation 9 the total linear density of dislocations $\rho_{int} = 4 \cdot (\rho_+ + \rho_-)$ varies as:

$$\dot{\rho}_{int} = \frac{2\sqrt{2}\dot{\epsilon}_y^p}{a} - \frac{1}{4} \cdot v_{int} \cdot \rho_{int}^2 \quad (17)$$

where the climb velocity $v_{int} = v_1/100$ follows the same law as population 1 dislocations with an arbitrarily chosen reduced coefficient. The variations of the plastic strain of the matrix are supposed to bring back the Von Mises stress to the Orowan stress level following load increases, or when dislocations from population 2 enter the rafts and need to be replaced by new matrix dislocations to keep $\delta_\perp$ constant. In the latter case, the strain rates in direction x of both phases must be equal. As dislocation glide in the matrix keeps the volume constant we have $\dot{\epsilon}_{y-xx}^p = \dot{\epsilon}_{y-yy}^p = -\frac{1}{2}\dot{\epsilon}_{y-zz}^p$, the strain rate of the matrix in direction z $\dot{\epsilon}_y^p$ is thus taken as:

$$\dot{\epsilon}_y^p \cong -A \cdot \dot{\sigma}'_{xx} - \dot{\epsilon}_2^p \quad (18)$$

It is assumed to be zero following the load drop at $t = 6500$ s. Coefficient A (taken here as $2 \cdot 10^{11} \text{ Pa}^{-1}$) depends on the elastic constants of both phases and their volume fractions. The contribution of interface climb to the plastic strain rate of the specimen in direction z is:

$$\dot{\epsilon}_{int}^p \cong \frac{2}{\lambda} \cdot \rho_{int} \cdot v_{int} \cdot \frac{a}{2} \quad (19)$$

As can be seen in Figure 3, the presence from the beginning of a high density of climbing interface dislocations changes the behavior of the material in a drastic way: even in domain 0b (up to 2500s) plastic strain in direction z takes place. The result is a supersaturation in vacancies and an osmotic stress only slightly lower than the applied stress. As the osmotic stress increases (decreases) the effective stress on population 2 (1) dislocations, the first dislocations which enter the rafts belong to population 2 (domain 1b) and not to population 1 as in domain 1. This results in a slightly decreasing osmotic stress, and population 1 dislocations can enter the rafts only later, when their effective stress exceeds the threshold stress at $t = 3500$ seconds. In domain 2b, the behavior of the rafts is akin to that of figure 2 in domain 2: the magnitude of the strain rate in direction x is slightly lower than the total strain rate in direction z. However, as part of

latter results from interface climb, the population 1 contribution is lower. We shall now describe the steady state behavior within domains 1 (and 0b) and 2 (1b and 2b).

Figure3: Evolution of the stresses (a), strain rates (b), and dislocation densities (c) during a simulated stepwise loading followed by unloading to 100 MPa. (With climb at the interface.)

Steady state behavior under low applied stress (no climb at the interface)

In the following, we shall assume a more general form of the dislocation entry rate laws:

$$\dot{\rho}_{m1} = a_i \cdot f(\sigma_i - \sigma_t) \text{ if } \sigma_i > \sigma_t^t \quad (20)$$

and zero otherwise. Function f can be a power law, a hyperbolic sine, or any more complicated but strictly increasing function of σ_i . Taking equation (7) under steady state conditions, without climb at the interface and using the Orowan equation $\dot{\varepsilon}_1^p = (\dot{\rho}_{mi} \cdot l_r / v_i) \cdot b \cdot v_i = \dot{\rho}_{mi} \cdot l_r \cdot b$ we get:

$$\frac{f \cdot k \cdot T}{\Omega^2 \cdot c_0} \cdot \dot{\varepsilon}_1^p = \frac{f \cdot k \cdot T}{\Omega^2 \cdot c_0} \cdot \dot{\rho}_{m1} \cdot b \cdot l_r = \frac{\sigma_c}{t_0} \quad (21)$$

or:

$$\dot{\rho}_{m1} = \frac{\Omega^2 \cdot c_0}{f \cdot k \cdot T \cdot t_0 \cdot b \cdot l_r} \cdot \sigma_c = K \cdot \sigma_c = K \cdot (\sigma_a - \sigma_1) \quad (22)$$

Reasonable values of coefficient K are less than unity (≈ 0.05). Equations (20) and (22) can be used to determine the entry rate of dislocations: a graphic solution is given in Figure 4. The working point is found at the intersection of the blue line and the red curve (black circle). As K is much smaller than a_1 (the slope of the entry rate for population 1), the effective stress σ_1 is found near the threshold stress. An increase of the applied load from σ_a to σ'_a (dotted line) results in a small shift of σ_1 and a slight increase of $\dot{\rho}_{m1}$. Assuming that the effective stress σ_1 remains nearly equal to the threshold stress, we get:

$$\dot{\varepsilon}_1^p \approx K \cdot l_r \cdot b \cdot (\sigma_a - \sigma_t) = \frac{\Omega^2 \cdot c_0}{f \cdot k \cdot T \cdot t_0} \cdot (\sigma_a - \sigma_t) = \frac{4\pi \cdot D \cdot \Omega^2 \cdot c_0}{f \cdot k \cdot T} \cdot \rho_p \cdot r_0 \cdot (\sigma_a - \sigma_t) \quad (23)$$

$$\dot{\varepsilon}_2^p \approx 0 \quad (23b)$$

In domain 1 (0b), the strain rate depends on the applied load, on the density and diffusion coefficient of vacancies, on the volume fraction of the γ' phase, on the density and size of pores, and on the difference between the applied stress and the threshold stress, but neither on dislocation densities and mobilities nor on their entry rate. As the contribution of population 2 to the plastic strain is naught, the change in volume due to $\dot{\varepsilon}_1^p = \dot{\varepsilon}_{zz}^p$ will result in the creation of vacancies, which will diffuse to the specimen surface and to the pores.

The same conclusion holds for interface dislocations in domain 0b, as equation 22 remains valid if we take $\sigma_t = 0$. It is indeed possible to write an equation with the same form as equation 20 by taking an equivalent entry rate of interface dislocations per unit surface of the rafts:

$$\dot{\rho}'_{int} = \frac{2v_{int}}{\lambda \cdot f \cdot l_r} \cdot \rho_{int} = \frac{4}{\lambda \cdot f \cdot l_r} \left(\frac{\dot{\varepsilon}_y^p \cdot v_{int}}{a\sqrt{2}} \right)^{1/2} \quad (25)$$

373

374 *Figure 4: Graphic solution of equations (20) (entry rate of dislocations as a function of σ_1 ,*
 375 *red line) and (22) (linear relation between the entry rate and the osmotic stress, blue line).*
 376 *The working point (black circle) lies at the intersection of both lines, and σ_1 is only barely*
 377 *larger than σ_1^t . An increase of the load to σ'_a moves the working point slightly to the right.*

378

379 Steady state behavior at higher stresses: the effect of climb at the interface

380 If only the two populations of dislocations are active and climb does not take place at the
 381 interface, equation (7) becomes:

$$382 \quad \dot{\rho}_{m1} - \dot{\rho}_{m2} = \frac{\Omega^2 \cdot c_0}{f \cdot k \cdot T \cdot l_r \cdot b \cdot t_0} \cdot \sigma_c = K \cdot \sigma_c \quad (26)$$

383 A graphical solution of (26) and (20) (without climb) can be found in Figure 5a. In a first step,
 384 we neglect the second hand of the equation, and use:

$$385 \quad \sigma_1 + \sigma_2 = \sigma_a - \sigma_c - \sigma'_{xx} + \sigma_c = \sigma_a - \sigma'_{xx} \quad (27)$$

386 (The $\dot{\rho}_{m2}$ curve is dilated, i.e. a coefficient $a_2 > a_1$ was chosen for reasons explained below.)
 387 An approximate working point σ_1 is found at the intersection of the red $\dot{\rho}_{m1}$ and blue $\dot{\rho}_{m2}$
 388 curves (black circle), as well as an approximate value of $\sigma_c = \sigma_a - \sigma_1$. However, to keep the
 389 vacancy supersaturation constant, we must have $\dot{\rho}_{m1}$ slightly larger than $\dot{\rho}_{m2}$ (and ε_1^p slightly
 390 larger than ε_2^p), i.e. the working point is slightly shifted towards higher σ_1 and lower σ_2 . We
 391 find a refined solution (red and blue circles) at:

392

393

394 Figure 5: Graphic solution of equations (26-28), without climb at the interface (a), and
 395 climb at the interface (b) (29-30). In (a) the approximate solution (black circle) is found at the
 396 intersection of the entry rates of populations 1 (red) and 2 (blue), and to keep the vacancy
 397 supersaturation constant, the working points must be shifted to the right (red and blue circles).
 398 In (b) the approximate solution is found at the intersection of the black (i.e. green + red) and
 399 blue lines. Under low stress (dotted blue line and circle) only interface dislocations and
 400 population 2 dislocations must be considered.

$$401 \quad \sigma'_1 = \sigma_1 + \Delta\sigma \approx \sigma_1 + \frac{K \cdot \sigma_c}{\left(\frac{d\dot{\rho}_{m2}}{d\sigma_2} + \frac{d\dot{\rho}_{m1}}{d\sigma_1}\right)} \quad (28)$$

If we now allow dislocations to climb at the interface, equation 26 becomes:

$$\dot{\rho}_{m1} + \dot{\rho}'_{int} - \dot{\rho}_{m2} = \frac{\Omega^2 \cdot c_0}{f \cdot k \cdot T \cdot l_r \cdot b \cdot t_0} \cdot \sigma_c = K \cdot \sigma_c \quad (29)$$

We now find the working point (figure 5 b) at the intersection of the black $\dot{\rho}_{m1} + \dot{\rho}'_{int}$ and blue $\dot{\rho}_{m2}$ curves (black circle). As $\dot{\rho}'_{int}$ is larger than zero, the working point has shifted to the left, and the osmotic stress is larger. The entry rate of population 1 (2) dislocations is thus smaller (larger) than without climb. The same correction as in equation 28 (not shown) can be done:

$$\sigma'_1 = \sigma_1 + \Delta\sigma \approx \sigma_1 + \frac{K \cdot \sigma_c}{\left(\frac{d\dot{\rho}_{m2}}{d\sigma_2} + \frac{d\dot{\rho}_{m1}}{d\sigma_1} + \frac{d\dot{\rho}'_{int}}{d\sigma_1}\right)} \quad (30)$$

Last, if $\sigma_a - \sigma'_{xx}$ is only slightly larger than σ_t , as in domain 1b, only interface dislocations and population 2 dislocations are active (dotted line and circle on the bottom left.)

The high load strain rate does not depend in the applied load only, but in the difference $\sigma_a - \sigma'_{xx}$, i.e. on the microstructure and internal stresses. It also depends strongly in the choice of the entry law for dislocations, while the contribution of the diffusion of vacancies to the pores is limited to the $\Delta\sigma$ shift. As nearly all vacancies created by population 1 dislocations and interface dislocations are absorbed by population 2 dislocations, very few diffuse to the pores: the specimen's volume remains nearly constant. From these solutions, we get a threshold value of the $\sigma_a - \sigma'_{xx}$ stress for the entry of population 2 dislocations into the rafts:

$$\sigma_a - \sigma'_{xx} \approx \sigma_a + \frac{1-f}{f} (\sigma_a - \sigma_{0r}) \geq \alpha \cdot \sigma_t \quad (31)$$

where α is one (figure 5b) if interface climb takes place, and two otherwise (figure 5a).

IV Consequences of the threshold stress hypothesis

Plastic strain and growth of pores

As seen above, the existence of a threshold stress for dislocation entry into the rafts directly results in a low stress domain (domain 1 or domain 0b) where all vacancies created by the climb of population 1 dislocations or interface dislocations diffuse to the surface and to pores, and a high stress domain (domain 2 or domain 1b) where the major part of these vacancies is absorbed by the climb of dislocations of population 2.

Epishin et al. measured the density of CMSX-4 and CMSX-10 specimens after interrupted creep tests at 1100°C. They showed that the specimens' densities decreased during creep. More

precisely for short creep times (up to 200 h) under moderate applied loads (120 MPa and below) the decrease in density is equal to the plastic strain in the direction of the tensile axis. For higher loads (135 MPa or more) or longer creep times the decrease in density is lower than the plastic strain.

The low stress creep tests at high temperature should correspond to domain 1 (0b), and the results under a 135 MPa load (0.3% decrease of the density for a 0.9% ε_{zz}^p strain) can be expected at the lower limit of domain 2 (1b). Assuming a 0.6 volume fraction of γ' phase, and a 700 nm period of the microstructure, the average width of the γ channels is about 270 nm. Taking a 70 GPa average Young modulus and a 0.42 Poisson coefficient, the Orowan stress [3] should be about 33 MPa. For a 120 MPa applied load, we obtain from equation 31 a threshold stress: $\sigma_t \approx 89$ MPa (no interface climb) or $\sigma_t \approx 178$ MPa (with interface climb).

For longer tests, the ripening of the microstructure increases its period from an initial 700 nm to 950 nm after 400 h. Keeping the other material parameters constant, the Orowan stress should decrease to 25 MPa, and the $\sigma_a - \sigma'_{xx}$ stress (using equation 31) should increase to 183 MPa, i.e. higher than $2\sigma_t$. (σ_t) This would result in a transition from domain 1 to domain 2.

Heterogeneity of the plastic strain rate and size effect

Equation (5), which describes the time behavior of the vacancy supersaturation, results from a 'mean field' approach. It does not hold in the vicinity of vacancy sinks such as pores and the specimen surface, where the vacancy concentration remains at or very near to the equilibrium concentration. The average osmotic stress at a surface layer is small or zero, the effective stress on dislocations of population 1 and interface dislocations (resp. 2) is σ_{zz} (resp. $-\sigma'_{xx}$). For the same applied load, the local effective stress for population 1 (resp. 2) and the strain rate in the [001] (resp. [100] and [010]) direction is larger (resp. lower) than in the bulk material. The threshold stress for entry into domain 2 (1b) becomes $\sigma_2 = -\sigma'_{xx} = \alpha \cdot \sigma_t$ and is reached for higher values of the local tensile stress σ'_{zz} . Even without taking into account the stress concentration effect of the pores, this will result in soft inclusions within a harder matrix and in a softer surface layer.

Figure 6: a; Schematic drawing of the vacancy depleted layer at a specimen's surface and around pores. b; Effect of a coating (light grey) on the density of vacancies within the specimen (darker grey). Continuous (dashed) line: without (with) coating.

The thickness of this surface layer (Figure 6) should be of the same order of magnitude as the distance between pores, i.e. much larger than the γ' -free zone observed in surface oxidized specimens. Its presence explains the higher creep rate measured for uncoated thin specimens.

A continuous coating, either deposited before the test [22] or formed by oxidation in a hydrogenated argon atmosphere [23], impedes the diffusion of vacancies to the surface of the specimens. It shifts the concentration of vacancies at the interface between the coating and the superalloy from c_0 to c'_0 , and decreases the difference between the surface layer and the bulk material. Further leaks through the porosities of the coating only reduce c'_0 .

Mobile dislocation densities in domain 2: TEM observations

In domain 2 (no interface climb), as the entry rates of dislocations of both populations are nearly identical, the density of mobile dislocations of population i is:

$$\rho_{mi} \approx \dot{\rho}_{mi} \cdot \left(\frac{l_r}{v_i} \right) \quad (32)$$

and the ratio of dislocations densities is:

$$\frac{\rho_{m2}}{\rho_{m1}} \approx \frac{v_1}{v_2} \quad (33)$$

In domain 2b (with interface climb) the dislocation densities remain of the same order of magnitude, especially under high loads, as interface climb becomes relatively less important. However, TEM observations [3,18] consistently show a one to five ratio between [100]

dislocations on one side, and $[100]$ et $[010]$ dislocations on the other side. As dislocations of both populations have the same core structure, they should have the same mobility, and a lower velocity can be explained only by a lower effective stress. This may result in part from the interface climb but also if, as in Figure 5, the coefficients of the entry law are different, i.e. $a_2 > a_1$. Assuming that during unit time a dislocation segment from each population has the same probability to enter a raft under the same conditions, the entry rate should depend on the number of dislocation segments, i.e. on the surface through which they can enter.

Figure 7 shows a parallelepipedal γ' raft embedded within a homogeneously strained γ matrix. We assume that the perfect $a/2 \cdot \langle 110 \rangle$ dislocation segments reaching the interfaces have paired to form edge $a \cdot \langle 100 \rangle$ dislocations with their additional plane on the γ' side which can bulge into the rafts. However, $[001]$ dislocations (red) can form only on the (100) and (010) surfaces (total area: $4l_r w$, where w is the raft thickness), while $[100]$ ($[010]$) blue (green) dislocations can form on the (010) and (001) ((100) and (001)) surfaces (total area: $2l_r w + 2l_r^2$ for each). Even accounting for the double density of $[001]$ segments, the ratio between the numbers of segments should be $\cong l_r/2w$, i.e. between 5 and 10. We shall thus take:

$$a_2 = 10 a_1 \quad (34)$$

Figure 7: Parallelepipedal γ' raft embedded within a homogeneously strained γ matrix. The interfacial network of $a \cdot \langle 100 \rangle$ edge dislocations is formed by reaction of matrix dislocations. As the signs of the Burgers vectors of matrix dislocations reaching the faces on different sides are different, $a \cdot [001]$ (resp. $a \cdot [100]$, $a \cdot [010]$) dislocation segments (red, resp. blue, green) can form only on (100) and (010) (resp. (010) and (001) , (100) and (001)) interfaces.

Mobile dislocation densities in domain 2: load drops

An alternative way to evaluate the dislocation densities within the rafts is to follow the transient mechanical behavior following a jump from a high stress or temperature (i.e. from domain 2). As shown in [3], while the plastic strain of the γ channels stops immediately, it carries on for a

while at a reduced rate within the γ' rafts in direction [001]: the dislocations already present within the rafts or at the interfaces are still moving. Besides (figure 8), the lattice mismatch $\delta_{\perp}$ increases in algebraic value and settles after about 15 minutes: the plastic strain also goes on in the [100] and [010] directions. The same behavior is observed in figures 2 and 3 in the post unloading domain. As $\delta_{\perp}$ results from dislocations at the γ/γ' interfaces and the plastic strain rate of the channels is supposed to be zero, it can change only if the rafts are plastically strained and we have:

$$\Delta\delta_{\perp} = -\Delta\varepsilon'_{xx} = -\Delta\varepsilon'_{yy} \approx \frac{1}{2}\Delta\varepsilon'_2 \quad (35)$$

Figure 8: Experimental variation of $\delta_{\perp}$ following a drop of the applied stress from 165 MPa to 92 MPa during an in-situ test. (1090 °C). [24]

As pointed in [25], the ρ_{m2} density of dislocations climbing within a raft before the jump move afterwards on an average distance $l_r/2$ before reaching its end and annihilating. The residual plastic strain for population 2 is then also:

$$\Delta\varepsilon'_2 \approx -\rho_{m2} \cdot b \cdot \frac{l_r}{2} \quad (36)$$

Thus, we have a relation between the density of population 2 dislocations before the jump and the variation of $\Delta\delta_{\perp}$ afterwards:

$$\rho_{m2} \approx \frac{4}{l_r \cdot b} \Delta\delta_{\perp} \quad (37)$$

The data recorded during in situ tests in [3, 24, 25] have been plotted together with post mortem TEM results as a function of the value of the perpendicular mismatch $-\delta_{\perp}$ before the load or temperature decrease in Figure 6. There is some dispersion of the data points: this is not unexpected, as they result from tests under quite different stress and temperature conditions: the volume fractions of γ' phases are different, and we expect a wide range of climb velocities. For small magnitudes of the initial mismatch ($0 < -\delta_{\perp} < 0.0035$) the dislocation densities are at most a few 10^{11} m^{-2} . (The relaxation of $\delta_{\perp}$ is barely visible.) The dislocation density then seems to increase linearly with the magnitude of $-\delta_{\perp}$, and reaches values of a few 10^{12} m^{-2} consistent with the TEM results.

Equation (32) states that the dislocation density should vary as $\dot{\rho}_{mi}/v_i$. For whatever reason, the existence of an empirical relation $\rho_m = f(x)$ between the dislocation density within the rafts and any parameter x , which is valid for all temperatures and all climb velocities, is possible only if the entry rate of dislocations is proportional to their velocity: $\dot{\rho}_{mi} \propto v_i \cdot f(x)$.

Figure 9: Dislocation densities within the rafts prior to a decrease of the temperature [25] (black squares, dots and diamonds corresponding to different specimens) and to a decrease in applied stress (green squares [24] and red dots [3]) plotted vs. the perpendicular lattice mismatch. The total dislocations densities (both populations with a strong majority of population 2 dislocations) measured post mortem on TEM thin foils (red circles) are scattered but are of the same order of magnitude.

V Modified constitutive law for the γ' rafts

Taking into account the remarks above, we shall write the entry rate of $a' \cdot \langle 001 \rangle$ dislocations into the rafts as:

$$\dot{\rho}_{mi} = a_i \cdot v_i \cdot (\sigma_i - \sigma_t) \text{ if } \sigma_i > \sigma^t \quad (38)$$

with $a_2 = 10 a_1$, and $\sigma^t = 178$ MPa.

We shall use again equation 17 for the population of interface dislocations. The remaining difficulty is to define the stress dependence of the dislocation climb velocity within the rafts and at the γ/γ' interfaces. The climb motion is supposed to depend on one or probably several thermally activated mechanisms (generation and motion of jogs along a dislocation line, diffusion of vacancies, drag by solute diffusion...), each one having its own activation energy and activation volume, and cannot be discussed here. For simplicity, and following [25] we choose to use an exponential law:

$$v_i = b(T) \cdot \exp(\sigma_i/\sigma_0) \quad (39)$$

where $b(T)$ and σ_0 are adjustable parameters. As population 1 dislocations and interface dislocations feel the same effective stress σ_1 , we take $v_{int} = c \cdot v_1$ with c a constant smaller than one. The evolution of the stresses, strain rates, and dislocation densities is plotted in figure 10 for a continuously increasing load for comparison with experimental results from [24]. The fit parameters are given in Table 1: the only ones which differ in the two sets are the temperature, the volume fraction (determined during the experiment), and $b(T)$. Figure 10 requires a few comments:

- The transition between domains 0b and 1b takes place within less than 10 MPa: as the applied stress varies from 138 MPa to 150 MPa, the ratio of the strain rates $-\dot{\epsilon}_{xx}/\dot{\epsilon}_{zz}$ varies from zero to 0.58, i.e. 100% or 42% of the strain along direction z results in voids. The transition is marked as above by a sharp increase of the strain rate.
- The stress dependence of the strain rate in domain 1b is quite different for both experiments. In the simulation, this results from a different slope of the $-\sigma'_{xx}$ vs. $\sigma_{zz} = \sigma_a$ curve, i.e. a change in the volume fraction of γ' from 0.62 (970 °C) to 0.42 (1090 °C).
- The osmotic force (and the concentration of vacancies) is maximum at the transition: the nucleation of new pores and the growth rate of existing pores should be maximum too.
- As it may be expected from equation (23), the low stress behavior is nearly independent from the parameters of the model (entry rate of dislocations, climb velocities) except the density and size of the pores which can be measured. Contrarily, the actual entry

law (equation 28) and climb velocity law for (equation 29) for $a'.\langle 001 \rangle$ dislocations need to be determined from experimental creep data.

Figure 10: Evolution of the stresses (a), strain rates (b), and dislocation densities (c) during a continuous loading with modified entry laws and dislocation velocities. d; Comparison with experimental data (dots [24]). (Assuming climb at the interface.)

628

629 **VI Concluding remarks**

630 As seen above, a simple model taking into account the diffusion of vacancies created during
 631 plastic strain can reproduce qualitatively the high temperature mechanical behavior of a single
 632 crystal superalloy with a rafted microstructure, if and only if it includes a threshold stress for
 633 the entry of dislocations. By contrast, a model without threshold would result (as in domains 2
 634 and 2b) in nearly equal strain rates in both x and z directions at any applied load and could not
 635 reproduce the most salient features of experiments.

636 As pointed above, this threshold can have several origins: a stress necessary to bypass the
 637 dislocation network at the interface [26], the line energy of the dislocation segment bulging into
 638 a raft, the necessity to escape from an atmosphere of alloying elements segregated at the
 639 dislocations [27], an interface reaction [11], or the work needed to form a groove in the interface
 640 when an isolated matrix dislocation [28, 29] or perhaps a pair of dislocations approaches.

641 The necessity to include the velocity of dislocations in the entry law is an interesting feature of
 642 the model. The simplest explanation is that a dislocation segment climbing within a raft exerts
 643 a back stress on other candidates, which blocks them until the dislocation has travelled some
 644 distance or has been replaced at the interface by two new dislocations come from the channels.
 645 A moving a . [001] dislocation reduces the thickness of a raft by about 0.1% and reduces the
 646 σ_{zz} stress by several tens of MPa, while a a . [100] dislocation reduces the length of the same
 647 raft by 0.01% and decreases $-\sigma'_{xx}$ by a few MPa only. This should result in a slowdown of the
 648 former dislocations vs. the latter: this is an alternative explanation for the lower entry rate
 649 coefficient for population 1 dislocations. However, under large stresses, and large strain rates
 650 this entry law would result in a depletion of the interface dislocation network: it is probably no
 651 longer correct.

652 Another feature of the model is the limited impact of dislocation climb at the interface at high
 653 stresses: as the interface contribution to the overall strain rate in direction z should vary as the
 654 square root of the strain rate of the matrix it cannot sustain large strain rates. The entry of
 655 population 1 dislocations is necessary, as shown by TEM observations.

656 Some of the details of the model remain hypothetical, as we have very few data on the climb
 657 velocity of dislocations, especially at the γ/γ' interfaces. This may be alleviated by reanalyzing

existing experimental data, such as the dependence of the strain rate on the applied load both at low and intermediate stresses. More specific experiments are also possible:

- There should be a correlation (with a strong hysteresis) between the plastic strain at the interface and the plastic strain of the channels. This might be investigated by following the behavior of the material during and after short load or temperature excursions (with a finite strain increment in the γ phase) under loads for which population 1 dislocations cannot enter the rafts. Another possibility would be to relate the density of dislocations climbing in the interface and the experimental conditions. The transition between the low stress and higher stress domains (domain 1 and 2 or domain 0 and 1b), i.e. decrease of the density of the specimens equal to or lower than the plastic strain should shift with temperature as the volume fraction of the γ' phase changes.
- In the low stress domain, for a given temperature, the strain rate should depend only on the density and size distribution of the pores, but not on the overall volume of pores. As shown in [30], a reduction of this density by Hot Isostatic Pressing indeed decreases the minimum strain rate by more than an order of magnitude. The evolution of the pores distribution during high temperature creep has been investigated by microscopy [31-32] and in bulk by X-Ray Tomography [33-34]. In the latter case they were limited to the largest pores, while the maximum of the pores' sizes distribution seems to be around three or four micrometers [31]. However, new high-resolution X-Ray Tomography techniques able to detect pores with sizes in the micrometer range are now becoming available.

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- 736

737 **Table 1**

738 Fit parameters for Figure 10d. (The parameters which differ in the two experiments are in bold case.)

		parameter	symbol	Exp. S1	Exp. S2
Vacancies		Temperature (°C)	T	1090	970
		Equilibrium concentration (m ⁻³)	c_0	3 10²³	8 10²²
		Diff. coeff. (m ² .s ⁻¹)	D	2.1 10⁻⁹	9.2 10⁻¹⁰
Pores		Density (m ⁻³)	ρ_p	10 ¹²	10 ¹²
		Radius (μm)	r_0	5	5
		Relax. time (s)	t_0	7.6	17
Dislocation entry		Coeff. 1 (m ⁻³ .Pa ⁻¹)	a'_1	5. 10 ⁸	5. 10 ⁸
	$\dot{\rho}_{mi} = a'_i \cdot v_i \cdot (\sigma_i - \sigma_t)$	Coeff. 2 (m ⁻³ .Pa ⁻¹)	a'_2	5. 10 ⁹	5. 10 ⁹
		Thresh. stress (MPa)	σ_t	178	178
Climb velocity		Coefficient (m.s ⁻¹)	$v_0(T)$	2 10⁻¹⁰	3 10⁻¹²
	$v_i = v_0(T) \cdot \exp\left(\frac{\sigma_i}{\sigma_0}\right)$	Base stress	σ_0	30 MPa	30 MPa
		Orowan stress (channels) (MPa)	σ_{Or}	50	76
Others		γ' phase fraction	f	0.49	0.62
		Raft's length (m)	l_r	5 10 ⁻⁶	5 10 ⁻⁶

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741

742 Figure captions

743 Figure1: a; Climb in the γ/γ' interface of two (red) dislocation segments left by a perfect
 744 dislocation gliding in the γ matrix. b; Climb in opposite directions of segments with different
 745 Burgers vectors. c; Strain rate of the γ matrix and at the interface vs. the strain rate of the matrix
 746 (left hand scale) and travel length (TL) of the dislocation segments (right hand scale) for a 1.25
 747 10^{-9} m.s⁻¹ interface dislocation velocity.

748 Figure 2: Evolution of the stresses (a), strain rates (b), and dislocation densities (c) during a
 749 simulated stepwise loading followed by unloading to 100 MPa. (No climb at the interface.) The
 750 thicker lines in domain 2 (a) result from quick variations of the supersaturation even during
 751 very short time steps.

752 Figure3: Evolution of the stresses (a), strain rates (b), and dislocation densities (c) during a
 753 simulated stepwise loading followed by unloading to 100 MPa. (With climb at the interface.)

754 Figure 4: Graphic solution of equations (20) (entry rate of dislocations as a function of σ_1 , red
 755 line) and (22) (linear relation between the entry rate and the osmotic stress, blue line). The
 756 working point (black circle) lies at the intersection of both lines, and σ_1 is only barely larger
 757 than σ_1^t . An increase of the load to σ'_a moves the working point slightly to the right.

758 Figure 5: Graphic solution of equations (26-28), without climb at the interface (a), and with
 759 climb at the interface (b) (29-30). In (a) the approximate solution (black circle) is found at the
 760 intersection of the entry rates of populations 1(red) and 2 (blue), and to keep the vacancy
 761 supersaturation constant, the working points must be shifted to the right (red and blue circles).
 762 In (b) the approximate solution is found at the intersection of the black (i.e. green + red) and
 763 blue lines. Under low stress (dotted blue line and circle) only interface dislocations and
 764 population 2 dislocations must be considered.

765 Figure 6: a; Schematic drawing of the vacancy depleted layer at a specimen's surface and
 766 around pores. b; Effect of a coating (light grey) on the density of vacancies within the specimen
 767 (darker grey). Continuous (dashed) line: without (with) coating.

768 Figure 7: Parallelepipedal γ' raft embedded within a homogeneously strained γ matrix. The
 769 interfacial network of a $\langle 100 \rangle$ edge dislocations is formed by reaction of matrix dislocations.
 770 As the signs of the Burgers vectors of matrix dislocations reaching the faces on different sides
 771 are different, a $[001]$ (resp. a $[100]$, a $[010]$) dislocation segments (red, resp. blue, green)
 772 can form only on (100) and (010) (resp. (010) and (001) , (100) and (001)) interfaces.

773 Figure 8: Experimental variation of $\delta_{\perp}$ following a drop of the applied stress from 165 MPa to
774 92 MPa during an in-situ test. (1090 °C). [24]

775 Figure 9: Dislocation densities within the rafts prior to a decrease of the temperature [25] black
776 squares, dots and diamonds corresponding to different specimens) and to a decrease in applied
777 stress (green squares [24]) and red dots [3] plotted vs. the perpendicular lattice mismatch. The
778 total dislocations densities (both populations with a strong majority of population 2
779 dislocations) measured post mortem on TEM thin foils (red circles) are scattered but are of the
780 same order of magnitude.

781 Figure 10: Evolution of the stresses (a), strain rates (b), and dislocation densities (c) during a
782 continuous loading with modified entry laws and dislocation velocities. d; Comparison with
783 experimental data [24]. (Assuming climb at the interface.)

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