Study of the deformation mechanisms in a Fe–14% Cr ODS alloy

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A B S T R A C T
In this present work, the plasticity of a rod bar of a 14% Cr ferritic ODS steel is examined through a multiscale approach based on both macroscopic and microscopic results. This bar was elaborated at CEA by powder metallurgy and consolidated by hot extrusion. The microstructure of the material has been characterized.

First, the tensile behavior of this material is studied in a wide range of temperatures. Thereafter, through in situ Transmission Electron Microscopy (TEM) straining experiments, dislocation/dislocation and dislocation/precipitates interactions are observed. The collapse of the tensile properties noticed from 400 °C can be explained by a change in the deformation mechanism. At lower temperatures, the hardening seems to be due to the precipitates, dislocations are pinned on oxides. At higher temperatures, the hardening role of the precipitates is still observed, but the dislocations seem to move in a more steady way, thermal activation of dislocations sources is observed and leads to formation of cavities at the grain boundaries.

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1. Introduction

The development of a new generation of reactors implies breakthrough in the field of materials. Particularly in sodium-cooled fast-neutron reactors (SFRs), the cladding materials have to face rough conditions in temperature, irradiation doses and solicitations.

In this context, considering an excellent dimensional stability under irradiation due to a ferritic matrix, and a high creep resistance thanks to the nano-dispersion of oxides, the oxide dispersion strengthened (ODS) steels are promising candidates as cladding materials [1]. At CEA, several studies are focused on these ODS materials.

Compared to those of austenitic steels, their tensile properties and their creep properties are highly improved. For instance, after 17,000 h of creep at 600 °C, the rupture strain is about 300–350 MPa for ODS alloys, whereas it is no more than 150 MPa for conventional steels [2].

Many studies have been carried out on nano-strengthened alloys, with a lot of microstructural observations, for different metallurgical treatments [3–6]. Besides, a wide range of mechanical tests – creep, tensile tests, and toughness for instance – have been done, under several experimental conditions (temperatures, stresses, strain rates etc.) [7–9]. However, a precise comprehension of the plasticity mechanisms is still needed, and only scarce data [10,11] coupling microscopic and macroscopic plasticity mechanisms are currently available on ODS alloys.

The deformation mechanisms in ODS are particularly linked to their complex microstructure (nano-oxide precipitates, high dislocations density, fine grain size etc.). To improve the comprehension of the plasticity mechanisms, analyses of the impact of the microstructural parameters on the mechanical behavior and observations of the plasticity mechanisms at a microscopic scale are then needed. To that respect, in situ Transmission Electron Microscopy (TEM) straining experiments which consist in dynamic observations of the microstructure under stress [10,12] are carried out.

The aim of this paper is then to give an insight on the plasticity of a Fe–14Cr ODS alloy, notably thanks to in situ TEM straining experiments. The microstructure of this Fe–14Cr ODS alloy is first finely characterized. On top of that, the mechanical properties of the material are studied. To reinforce the multiscale approach of this work, the dislocation motion is observed through in situ TEM straining experiments. Thus, first elements of the plasticity of ODS from room temperature to 600 °C, such as the modification of the dislocation motion, are evidenced.

2. Experimental

The oxide dispersion strengthened alloys are elaborated by powder metallurgy. A pre-alloyed powder of steel is mechanically alloyed to a fine powder of yttria in an attritor during 10 h, with a ball to powder ratio of 15. A rod bar of ODS steel with a ferritic matrix is obtained through a series of steps: the powder is sealed into a can and degassed at 300 °C under a vacuum better than 10−5 mbar, hot extruded at 1100 °C, then it is air-cooled and finally annealed at 1050 °C for 1 h and air-cooled at CEA.
The nominal composition of the Fe–14% Cr ODS listed as J05 is the following: (in weight%, iron is the balance) Fe–14Cr–1W–0.3Ti–0.3Y2O3.

The crystallographic orientation of grains was determined by X-ray diffraction (XRD) on samples electrolytically polished in a solution containing 90% of ethanol and 10% of perchloric acid, using a 4-circle goniometer D500 Siemens equipped with diode SiLi as detector. The measurements were made on the surface normal to the extrusion direction of the bar.

For the microstructural observations, classical TEM specimens and microtensile specimens for in situ experiments were mechanically grinded to less than 100 μm thick. Disks of 3 mm were punched for the TEM specimens and rectangles of 3 mm × 1 mm were cut using a diamond coated wire saw for the microtensile specimens. Both types of specimen were electrolytically jet polished with a solution containing 70% of ethanol, 20% of ethylene glycol monobutyl ether and 10% of perchloric acid at 5 °C. The observations were carried out on JEOL 2010 and Philips EM430 operating respectively at 200 and 300 kV.

Texture measurements were also carried out at the microscopic scale, using the ASTAR automatic crystal orientation and phase mapping system. It consists in collecting TEM micro-diffraction patterns (probe resolution around 20 nm) over a large sample area, and comparing them to a database of known crystal orientation [13].

Macro-tensile specimens are taken from the core of the bar with the gauge length parallel to the extrusion direction of the bar. Two geometries of specimens have been chosen: the first one with a gauge length of 11.5 mm and a square section of 2 mm × 2 mm and the second one with a gauge length of 15 mm and a circular section of 3 mm diameter. The tests were carried out from room temperature to 750 °C at strain rates of 7 × 10⁻⁴ s⁻¹ and 1 × 10⁻⁵ s⁻¹ on a hydraulic Instron machine equipped with a 50 kN load cell. Strain rate jump tests were carried out at room temperature and 600 °C, using the same geometry of samples as for tensile tests, in order to evaluate the strain rate sensitivity. The tensile specimens, once tested, have been cut along the extrusion direction. They were plated with Nickel to reduce as much as possible the edge effect induced by the coating on the fracture surface. To ensure good conductivity, gold has been deposited on the specimen by cathodic spraying.

3. Results

3.1. Microstructural results

All the investigated specimens present an obvious anisotropic microstructure. The grains are elongated along the extrusion direction: they are 1.5 μm long and 600 nm wide which leads to a grain aspect ratio of ~2.5 as illustrated in Fig. 1a. Observations of samples taken perpendicularly to the extrusion direction show a certain homogeneity in the grain size, with only a small dispersion in the size measurements.

The texture measurements by XRD confirm a (110) fiber texture parallel to the extrusion direction (Fig. 1b). Moreover, a high index of deformation texture is found for this specific orientation. This is probably a result of the elaboration process.

At TEM resolution, the crystal orientation mapping method points out crystallographic texture with a preferential orientation (110) (Fig. 1c), which is in agreement with the XRD measurements.

Due to the elaboration route and the heat treatments, the material has a high dislocations density close to 5 × 10¹⁶ m⁻² forming tangles. As already observed in ODS alloys [10,11], the dislocation Burgers vector is supposed to be ½ (111) as usual in bcc structures. The dislocations are straight and aligned along their screw component, indicating the existence of Peierls valleys. Moreover, the dislocations tend to be pinned on precipitates marked by the arrows in Fig. 2.

The nano-oxides are known to be responsible for the increase of the mechanical properties. To complete the microstructure characterization, the precipitation was analyzed both morphologically and chemically. This latter point is not tackled in this paper.

The observations revealed the presence of two types of precipitates: nanometric ones and coarse ones which size is spread from 15 nm to several hundreds of nm, in lower density. This second family of precipitates is not studied here, this work being focused on the small nano-oxides. They are said to be the more efficient to strengthen the alloy [14]. On conventional TEM thin foils, 420 precipitates have been counted yielding a density around 5 × 10¹⁰ m⁻². The average diameter of the nano-oxides is 2–3 nm (Fig. 3). As specified in [15] the interparticle distance thus calculated is λ = 100 nm. In the meantime, TEM analyses on extractive replicas of J05 were made. It consists in a selective dissolution of the matrix to retrieve the nano-oxides. With this method, a larger mean diameter for nano-particles is found. It is suspected that the smaller ones, mostly coherent with the matrix, are dissolved, hence a higher particle size. One advantage of observations on replicas is the better resolution of the nano-oxides. Plus, a higher density of nano-precipitates is found on replicas, and consequently a smaller inter-particle distance: λ = 24 nm [16,17]. Small Angle Neutron Scattering (SANS) analyses of the precipitation show two populations of small nano-precipitates, with an average diameter in the same order of magnitude as for TEM measurements, and another one twice larger. The density calculated through this technique is slightly higher than through TEM observation.

3.2. Mechanical results

The samples were tested in a wide range of temperatures and for several strain rates [18]. It appears that the mechanical properties, both yield stress and mechanical resistance, decrease slowly with the temperature up to 400 °C (1070–830 MPa). At higher temperature, a collapse of the properties is noticed: Rp0.2°C is only 300 MPa at 750 °C. A change in deformation mechanism is thus expected.

An astonishing result is that the faster the material is strained, the more ductile it is. A drastic loss of ductility for slower strain rates above 600 °C is observed. For instance, at 750 °C, the total elongation collapses from ~25% at 7 × 10⁻⁵ s⁻¹ to ~3% at 10⁻⁵ s⁻¹ (Fig. 4).

Considering the evolution of the properties, the idea of a change in deformation mechanism above 400–500 °C is reinforced.

In addition to that, a large increase in the total elongation is observed at 600 °C; it reaches its maximum at this temperature (35% at a strain rate of 7 × 10⁻⁴ s⁻¹ or 25% at 10⁻⁵ s⁻¹). To sort that out, strain rate jump tests are carried out at room temperature and 600 °C. They show a high strain rate sensitivity of the stress, S, at 600 °C defined as followed:

\[ S = \frac{\Delta R_{P0.2°C}}{\ln(\dot{\varepsilon})} \]  

where Rp₀.2°C is the yield stress of the material and ε is the considered strain rate. A variation of almost 100 MPa on the yield stress is observed when changing the strain rate from 7 × 10⁻⁴ s⁻¹ to 25% at 10⁻⁵ s⁻¹. At room temperature, the variation hardly reaches 10 MPa. These variations have also been calculated through conventional monotone tensile tests for different strain rates. The differential strain rate method, presented in [19], is more accurate to evaluate the effect of the strain rate, since the microstructure parameter can be considered as constant. Nevertheless, since the two techniques give the same strain rate sensitivity at 20 °C and 600 °C, we can reasonably...
assume that tensile tests results can be used to determine the strain rate sensitivity at all temperatures. The evolution of $S$ with the temperature is given Fig. 5. These tests have highlighted a peak of viscosity which may explain the peak of ductility. From room temperature up to 400 °C, the strain rate sensitivity is almost constant ($\approx 5$ MPa), while it shows a steep increase at 600 °C (23 MPa). At higher temperature, the strain rate sensitivity becomes again little-pronounced as at room temperature.

The observation of longitudinal cuts of specimens tested at 600 °C and 100 °C show a high density of porosities aligned along the extrusion direction as illustrated in Fig. 6. As it will be confirmed later on with in situ TEM straining experiments, they seem to be mostly located at the grain boundaries. On the contrary, for the sample tested at room temperature and 7 °C, no porosities are observed. Vickers hardness measurements were made along the gauge length on the longitudinal cuts. A reduction of 15% of the Vickers hardness (i.e. about 50 Hv) in the necking areas coincides with the appearance of the porosities. A difference of necking appears depending on the testing conditions: a strong reduction of section has been noticed, leading to a high necking coefficient for the sample tested at room temperature, whereas the gauge section does not vary that much for the sample tested at higher temperature and slower strain rate [18]. This result combined to SEM fracture observations confirm the evolution of the damage mechanisms above 400 °C/500 °C.

### 3.3. In situ TEM straining experiments

These experiments were carried out at 20 °C, 400 °C and 500 °C. Fig. 7 shows pictures extracted from a video recorded at room temperature. Upon straining, the dislocation labeled d (Fig. 7a) unpins and its screw segment grows upon straining (Fig. 7b). Fig. 7c illustrates its motion, in white the position of the dislocation at $t = 0$ s, and in black its position at $t = 0.40$ s. During the second sequence, starting at $t = 51$ s (Fig. 7d–f), the dislocation adopts a specific shape with an elongated segment, corresponding to the screw direction assuming that the dislocation Burgers vector is $b = 1/2 [1–11]$. Note the direction of the trace of the plane (011) at the sample surface (Fig. 7d), determined with the stereographic projection Fig 7g. This configuration results from the higher mobility of the edge segment as already observed in pure Fe [20,21]. Fig. 7h presents the configuration of the dislocation at $t = 158.44$ s. The pinning points are clearly noticed along the dislocation. Dislocation pinning–depinning process, not shown here is frequently observed. It can be also noticed that screw segments move jerkily between pinning points.

Up to 400 °C, intragranular dislocation mechanisms are noticed leading to homogenous deformation.

At 500 °C, a more intense dislocations activity is noticed close to the grain boundaries. Dislocations sources are observed in these particular areas. Fig. 8a and b shows two snapshots taken in an interval of 0.32 s, illustrating the motion of a dislocation close to grain boundary. The dislocation does not exhibit straight screw segments, obviously the lattice friction is negligible at such temperature. The dislocations motion appears to be steadier. Their pinning and depinning is evidenced. Indeed, although the nanoprecipitates cannot be clearly seen, their presence is evidenced by the bowing out of the dislocations when they become pinned.

Fig. 8c shows a sketch of dislocation configuration in Fig. 8a and b. More than 20 dislocation motions have been observed in this area indicating that a stable source operated near the grain boundary. This observation combined with a lower intragranular activity indicates that the deformation localizes at the grain boundary.

At higher deformation, cavitation and decohesion are clearly evidenced. Grain boundary decohesion can also be seen at larger scale (Fig. 8d).
4. Discussion

4.1. Mechanisms

The results obtained through the in situ straining experiments on the ODS Fe–14% Cr are consistent with the macroscopic results. Tensile tests underlined the stability of the mechanical properties, particularly yield stress and ultimate tensile stress, from room temperature to 400 °C. At higher temperatures, there is a collapse of the properties, and the material exhibits a high ductility at 600 °C [18].

From in situ TEM observations, it appears clearly that plasticity is controlled by intragranular mechanisms from room temperature to 400 °C. At higher temperatures, the role of grain boundary seems to be crucial. The presence of porosities aligned along the grain boundaries only in samples tested at high temperature emphasizes the intergranular character of the deformation at high temperature. The modification of damage mechanism – which can be an illustration of the mechanism of deformation – has already been pointed out in this material in [18]. The fracture surfaces are typical of ductile fracture with well defined dimples for temperatures below 400 °C. At higher temperatures, signs of intergranular...
decohesion are noticed, the dimples are smaller and are only located at the very periphery of the fracture surface. A similar behavior, according to the fracture maps, has already been noticed on the 14YWT alloy [22].

The damage of the material is more severe for high temperatures regime, and the grain boundaries seem to play a more important role on damage mechanisms than they used to at lower temperatures. Particularly the activation of sources and the
formation of cavities in these areas are only noticed from 500 °C. Consequently, it can be assumed that the grain boundary have an impact on the deformation mechanisms.

It can be noted that in the whole temperature range, nano-oxides act as efficient pinning points for the dislocations. The analysis of the thermal activation parameters – the strain rate sensitivity $S$ and the activation volume $V_a$ – can give clues to determine the mechanisms that control the deformation [23,24]. The activation volume of the J05 is calculated considering the formula below,

$$V_a = k_B T \frac{\partial \ln \dot{\varepsilon}}{\partial T} \approx k_B T M \frac{\Delta \ln \dot{\varepsilon}}{\Delta T}$$  \tag{2}

with $M$ the Taylor's factor equal to 2.5, $k_B$ the Boltzmann constant, $T$ the temperature, $\dot{\varepsilon}$ the strain rate, and $\sigma$ the macroscopic stress.

As already mentioned, it is estimated through both differential strain rate tests and strain rate jump tests. The evolution of the activation volume with respect to the temperature is plotted in Fig. 9. It increases slightly from $\sim 100b^3$ at 20 °C to $160b^3$ at 400 °C. At 600 °C, the activation volume drops to $65b^3$. Afterwards, it increases again but faster, up to 750 °C ($V_a = 310b^3$). According to the slopes, two regimes can be distinguished: one at low temperatures, and another one for higher temperatures. At room temperature, in J05, $V_a$ is close to the value obtained in MA 956 alloy [23]. In the Eurofer ODS alloy [19] two different regimes have also been identified, with similar values and evolution of the activation volume with the temperature. The change of slope and the sudden drop of $V_a$ between 400 and 600 °C probably stands for a change in mechanism of deformation that occurs at these temperatures. Between room temperature and 400 °C, several mechanisms which explain the measured activation volumes are expected to happen. First in situ TEM strain experiments at room temperatures revealed an important friction stress acting on screw dislocations. This phenomenon combined to the jerky dislocations motion are in agreement with a locking/unlocking mechanism as recently proposed in Fe [20,21]. The activation volume values determined thanks to tensile tests could be consistent with this thermally activated gliding mechanism. Indeed, the hardening role of the alloying element over the friction stress has been evidenced in Fe with a few% Si alloys [21,25] leading to $V_a$ in such alloys close to $100b^3$ at room temperature.

Secondly, localized fixed obstacles can be easily overcome at moderate temperatures. These obstacles can be solute impurities atoms (i.e. Cr, W, Ti atoms) or immobile dislocations (i.e. forest mechanisms). In both cases it yields to an activation volume:

$$V_a = Lbw$$  \tag{3}

With $L$ the distance between obstacles along the dislocation line and $w$ the obstacle width. For solute atoms [26]:

$$L = b \left( \frac{\mu}{kT} \right)^{1/3}$$

(4)

With $c$ the atomic fraction of the solute and $\tau$ the effective resolved stress. Taking $c = 0.01$ [23] yields at room temperature to $L = 20b$. Taking $w = b$ yields then to $V_a = 20b^3$ which is five times smaller than measured. Dislocation–solute atoms interaction can be ruled out as a controlling mechanism below 400 °C.

For the forest mechanism, the distance between obstacles is

$$L = \frac{1}{\sqrt{\rho}}$$

(5)

Taking again $w = b$ and $\rho = 5.10^{14} \text{m}^{-2}$ at room temperature yields to $V_a = 180b^3$ which is comparable to the experimental value.

The strong increase of the strain rate sensitivity and the activation volume above 400 °C, correlated to a more steady dislocation motion, indicates that other thermally activated mechanisms should operate. One strong candidate is the dislocation dragging of solutes atmospheres. Indeed, at such temperatures solute atoms can control the velocity of moving dislocations. In polycrystals Zr containing oxygen impurities, a strong activation peak around 400 °C has been unambiguously attributed to dynamical strain ageing [27]. This mechanism might explain the strong increase in the activation volume at 600 °C in our ODS.

4.2. Hardening models

In the following, a first approach is given to fit the evolution of the yield stress with the temperature. As the ODS alloys have a specific microstructure that confers them their interesting properties, it is important to evaluate the influence of each microstructural characteristic on the hardening of the material. The three considered contributions to the flow stress are the grain size, the dislocations density, and the precipitation.

For the first one, usually, the Hall–Petch effect is considered, where the induced stress decreases with the grain size. As suggested in [28]:

$$\sigma_{HP} = \frac{k}{\sqrt{D}} \text{ with } k = \frac{1}{5} \mu \sqrt{b}$$

(6)

where $\mu$ is the shear modulus, 89 GPa at room temperature, and $D$ the length of the glide system. Assuming that the glide system is made of the direction [111] and the (−1−1−2) plane, considering the higher Schmid factor, $D$ is estimated to be about 1000 nm. The values of the different contributions to the hardening for different temperatures are gathered in Table 1.

The second source of hardening is due to the interaction between dislocations. When a mobile dislocation interacts with another one which goes across its gliding plane, an induced stress can be defined with this formalism:

$$\sigma_{forest} = M_\alpha \mu b \sqrt{p_{\text{disloc}}}$$

(7)

where $\alpha$ is a factor that describes the efficiency of the hardening induced by the dislocations, $\alpha = 1/3$ [29], and $p_{\text{disloc}}$ is the dislocations density of the material determined earlier.

The third characteristic parameter of the microstructure is the dispersion of nano-oxides. Precipitates are playing an important role in the hardening of the material, pinning the dislocations. In order to estimate their impact on the hardening, an Orowan contribution is calculated through this equation [15]:

$$\sigma_O = \frac{M \mu b}{2 \pi \lambda} \left( \ln \left( \frac{D}{r_0} \right) + B \right)$$

(8)

where $\lambda$ is the inter-particle distance on a slip plane, estimated through TEM observations to 100 nm, $r_0$ is the cut of radius of the
core dislocations, assessed to be equal to \( b \) in our case, and \( \overline{D} \) is the average diameter of the precipitates. \( A \) and \( B \) are parameters that vary depending on the edge or screw character of the considered dislocation, for a screw dislocation, \( A = \frac{1}{15} \) and \( B = 0.6 \) [5]. The value of \( \lambda \) estimated during in situ TEM straining experiments is close to the one on thin foils. But it is suspected that not all the precipitates play a role in the hardening of the material.

From the in situ TEM straining experiment at 500 °C, the stress needed to overcome a precipitate can be estimated. The local stress acting on the dislocation can be calculated using the measure of the dislocation curvature just before unpinning [30,31]:

\[
\sigma_{\text{local}} = \frac{Ml_{b}}{4\pi R}\ln\left(\frac{R}{b}\right) \tag{9}
\]

With \( R \) the curvature radius, taking \( R = 30 \text{ nm}, M = 2.3 \), determined through the exact experimental conditions, \( \mu = 72 \text{ GPa} \) at \( T = 500 \text{ °C} \) and \( b = 0.248 \text{ nm} \) yields \( \sigma_{\text{local}} \approx 460 \text{ MPa} \). Because of thin foil effects, the internal stresses (i.e. forest hardening and the Hall–Petch contribution) can be neglected and thus the local stress is:

\[
\sigma_{\text{local}} = \sigma - \sigma_{\text{forest}} - \sigma_{\text{HP}} - \sigma_{\text{p}} \approx \sigma - \sigma_{\text{p}} \tag{10}
\]

where \( \sigma_{\text{p}} \) is the precipitate overcoming stress. Taking \( \sigma_{\text{Orowan}}(500 \text{ °C}) = 720 \text{ MPa} \) leads to \( \sigma_{\text{p}} \approx 260 \text{ MPa} \). Since this value is close to the Orowan stress calculated above, the precipitates can be passed by an Orowan mechanism. Considering the hypothesis of calculation, and the measurement uncertainty, the difference between \( \sigma_{\text{p}} \) and the Orowan stress might be due to a temperature effect and thermal activation.

Between room temperature and 400 °C, the yield stress can be described, in a simplistic approach, summing these 3 terms. Fig. 10 shows the experimental measured yield stress (circles) as a function of the temperature. The crosses correspond to the sum of the three hardening contributions assuming that the dislocations density is constant with the temperature. Indeed, observations on thin foils from specimen tested at 650 °C lead to the same dislocations density as at room temperature, and the work hardening is reduced at high temperature for these materials. Then, it is assumed that the only temperature dependence of the yield stress comes from the variation of the shear modulus with the temperature. Actually, this variation, which is linked to the energy factor of the dislocations [23], explains the softening of the material between room temperature and 400 °C, since the yield stress can be expressed as a linear combination of \( \mu \). Despite this hypothesis, a good agreement with the experimental data is found between room and 400 °C. This indicates that the hardening mechanisms are weakly thermally activated which is in agreement with the discussion above.

But, for higher temperatures, from 500 °C, the model does not fit the experimental curve anymore. The evolution of the yield stress diverges from the one of the shear modulus. It reinforces the idea that for higher temperatures it is likely to be thermal activation of the mechanisms responsible for the clearing of the precipitates. The damage at the grain boundaries of the material becomes severe for high temperatures range, thus, the observed softening of the material might also be a consequence of a damage process.

### 5. Conclusions

In this present work, the behavior of a rod bar of a Fe–14% Cr ODS steel elaborated at CEA has been mechanically characterized from room temperature to 750 °C. In situ TEM straining experiments have also been carried out at different temperatures. The main points are the following:

- The mechanical properties drop from 400 °C, with an important strain rate effect on these proprieties. The ductility of the material is limited at high temperature, and drops consequently when the strain rate is decreased.
- The J05 exhibits a high viscosity at 600 °C. It is related to the strain rate sensitivity that reaches a maximum at this temperature. It explains the peak of ductility.
- The hardening role of the precipitates is underlined at room temperature, with dislocations pinned on the nano-oxides. An Orowan mechanism is suspected from 20 °C to 400 °C.
- The role of the grain boundaries is becoming more important at higher temperature, with the activation of sources in these areas. For high temperature, thermal activation enables the dislocations to get through the precipitates more easily.
- The evolution of the activation volume with the temperature suggests 2 regimes. From 20 °C to 400 °C, experiments show an important friction stress, coherent with a locking/unlocking mechanism. The strong increase of \( V_a \) above 600 °C could be attributed to dynamical strain ageing.
- The deformation mechanism is modified with the temperature, with an intragranular character turning into intergranular for higher temperatures.
- Damage is more severe at high temperature than at low temperature, particularly with the formation of cavities leading to decohesion along the grain boundaries.

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