HIGH STRAIN RATE DUCTILITY IN UNIAXIAL TENSION : A REVIEW

G. Regazzoni, F. Montheillet

JOURNAL DE PHYSIQUE
Colloque C5, supplément au n°8, Tome 46, août 1985

HIGH STRAIN RATE DUCTILITY IN UNIAXIAL TENSION : A REVIEW

G. Regazzoni and F. Montheillet

Ecole des Mines de Paris, Centre de Mise en Forme des Matériaux (UA CNRS 852, GRECO Grandes Déformations et Endommagement), Sophia Antipolis, 06560 Valbonne, France

Résumé - Les problèmes spécifiques posés par l'essai de traction à grande vitesse sont d'abord examinés, ainsi que les différents paramètres mesurant la ductilité. Des résultats expérimentaux relatifs à un cuivre de haute pureté sont ensuite présentés pour illustrer l'accroissement typique de la ductilité dynamique avec la vitesse de déformation. Enfin, les principaux facteurs susceptibles de déterminer la ductilité sont successivement analysés. On en conclut que le principal phénomène est l'augmentation de la stabilité avec la vitesse de déformation, qui elle-même peut résulter de l'apparition d'une loi de comportement linéaire et/ou des effets d'inertie.

Abstract - The specific problems involved by high strain rate tension tests are first examined, as well as the various parameters describing ductility. Some experimental data relative to a high purity copper then illustrate the typical increase of dynamic ductility with strain rate. Finally, the main factors which are expected to determine ductility are reviewed. It is concluded that the dominant factor is the increase of stability with strain rate, which in turn may result from the occurrence of a linear flow rule and/or from inertia effects.

I - INTRODUCTION

Behaviour of materials at high strain rates has been the subject of an increasing number of studies during the last years. In particular, investigations on ductility, first essentially motivated by military applications (e.g. shaped charge jets), are now necessary in a lot of engineering applications. For instance, high strain rates are involved in metal-working processes, such as extrusion ($\dot{\varepsilon} \simeq 10^3$ s^{-1}), machining ($\dot{\varepsilon} \simeq 10^4$ s^{-1}), explosion or magnetic forming ($\dot{\varepsilon} \simeq 10^3$ s^{-1}). Safety problems are also concerned by high strain rate ductility, for instance shock absorbers in the car industry, resistance of nuclear vessels to overpressure, etc.

The domain of strain rates considered in the present paper, i.e. 10^3 s^{-1} $< \dot{\varepsilon} <$ 3.5×10^3 s^{-1}, is often referred to as "dynamic range" because inertia effects are expected to be non-negligible at such rates which involve short times and large accelerations, whereas the latter are very weak at low strain rates ("quasistatic range"). Other features contribute to characterize the dynamic range; some of them will be considered below, such as adiabatic heating, specific flow rule, etc. Two problems are to be distinguished for clarity:

(i) The comparison between ductility at low and high strain rates: according to Kawata et al./1/, ductility is larger at high strain rates than in the quasistatic range for fcc materials at room temperature, and conversely for bcc structures. It is also suggested that the behaviour of hcp metals is similar to that of fcc ones when the ratio c/a of the lattice parameters is close to the theoretical value 1.63, whereas it is analogous to that of bcc materials when c/a is far from 1.63.

(ii) The second problem, which will be considered in the present article, concerns the evolution of ductility within the dynamic range: there are very few references on the subject and they will be analyzed below.

In section 2, the methods for measuring ductility in uniaxial tension at high strain rates as well as some related problems will be examined. Section 3 will then present some typical data, concerning essentially the ductility of pure copper deformed at room temperature and 500 °C /2-5/. Finally, the main factors which are likely to determine the ductility level and its evolution in the dynamic range will be reviewed and discussed in Section 4.

II - EXPERIMENTAL METHODS AND PROBLEMS

Although only uniaxial tension tests will be dealt with in the following, it is worth to notice that some investigations on ductility at high strain rates have been performed using the method of expanding rings /6/. Dynamic tension tests may be carried out using a split Hopkinson bar /7,8/ or alternatively an impact testing machine /9,10/. Since the latter device was used in the experiments reported below, it will be described in greater detail.
Each test supplies two diagrams, namely the load (at the fixed extremity) vs. time and elongation vs. time curves. Interpretation of such experiments rises a number of specific difficulties, which have been discussed for instance in /11/. In particular, the strain, strain rate and stress are inhomogeneous at the beginning of the test (see section IV.1a). The local variables strain and strain rate can be measured by using a high speed camera, but at the price of very long and delicate experiments /10/. The overall strain rate can be deduced from the elongation vs. time diagram at any given time. In practice, however, each experiment is characterized by a nominal strain rate given by $\dot{\varepsilon} = \Delta L_F/(\Delta t_F L_0)$, where ΔL_F is the elongation at fracture, L_0 the initial gage length and Δt_F the duration of the test. Since the rate of elongation (i.e. the velocity of the mobile head of the specimen) is approximately constant and equal to $\Delta L_F/\Delta t_F$, the above equation gives the initial overall strain rate and it should be noted that the instantaneous overall strain rate decreases during the test. Moreover, since the load vs. time diagrams exhibit an initial peak, the onset of plastic flow is difficult to determine accurately, such that the determination of Δt_F and $\dot{\varepsilon}$ is still questionable.
Various parameters can be employed to characterize ductility (Fig.1):

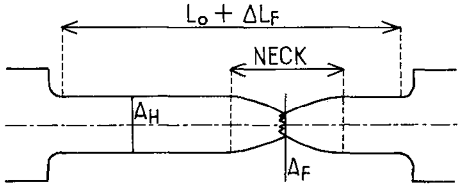

Fig. 1 - Schematic representation of a broken specimen.

(i) The total elongation at fracture $e_F = \Delta L_F/L_0$ has the drawback to include both the homogeneous and the necked part of the specimen, but its measurement can be made easily on the broken specimen.
(ii) The elongation at the maximum of the load vs. time diagram e_M. At this step of the elongation, the specimen cross-section is generally still (macroscopically) uniform, such that the (homogeneous) associated true strain $\varepsilon_M = \ln(1 + e_M)$ can be readily calculated. In a rate insensitive material, the latter is given by the Considère criterion $\varepsilon_M = \partial \ln\sigma/\partial \ln\varepsilon$ where $\sigma(\varepsilon)$ is the strain hardening law of the material. The value of e_M can be determined from the load and elongation diagrams.
(iii) By using a shadow projector, the average cross-section A_H of the homogeneous part of the specimen can be measured. From A_H it is possible to deduce an "elongation of the homogeneous part" e_H. It can be easily shown that $e_H = A_0/A_H - 1$, where A_0 is the initial cross-section. e_H is normally equal to e_M in the case of rate insensitive materials or greater than e_M due to strain rate effects /12/; in some cases, however, values of e_H lower than e_M have been observed /13/.

(iv) Finally, the length and volume of the neck can be determined with the shadow projector, as well as the cross-section at fracture A_F. When the latter is very reduced, it can still be measured by using the scanning electron microscope.
A full understanding of ductility (i.e. the necking and fracture processes) requires some microstructural investigations concerning for example the inhomogeneities of deformation at the grain scale and the fracture surfaces. The latter are related to damage, which has been measured in some cases from broken specimens /2/.

III - SOME TYPICAL DATA

The flow rule and ductility of OFHC and high purity coppers have been extensively investigated at various temperatures over the range of low and high strain rates /2-5/. As an example, the dependence of ductility on strain rate in the dynamic range is represented in Figs.2a and b in the case of a 99.999 % pure copper at 20 °C and 500 °C, respectively.
At room temperature ($T \approx 0.22T_m$), ductility remains almost constant at low strain rates, such that the dynamic data can be easily compared with the quasistatic ones. The values e_F = 53 % and e_H = 47 % measured at low strain rates are significantly lower than in the dynamic range. Within the high strain rate range, e_F and e_H increase continuously, the average slopes being nearly the same. e_M increases very slightly in the same interval.
On the other hand, at 500 °C ($T \approx 0.57\ T_m$) ductility exhibits large changes in the quasistatic range /2,3,14/: at very low strain rates ($\dot{\varepsilon} = 10^{-4}\ s^{-1}$), it is strongly reduced by intergranular brittleness; moreover, for $\dot{\varepsilon} \approx 5\times10^{-3}\ s^{-1}$ it goes through a maximum, which is attributed to the effect of dynamic recrystallization /2/. For these reasons the comparison between quasistatic and dynamic data is difficult. However, if $\dot{\varepsilon} \approx 5\times10^{-2}\ s^{-1}$, which falls after the above mentioned peak, is chosen as a reference strain rate for the quasistatic range, the corresponding ductility (e_F = 57 %, e_H = 42 % and e_M = 20 %) are again lower than in the dynamic range. The dependences of e_F, e_H and e_M within the interval of high strain rates are quite similar to those observed at 20 °C. However, the difference between e_F and e_H, which corresponds roughly to the elongation of the neck, is larger at 500 °C than at room temperature.
The cross-section A_F of the neck at fracture is very low at both 20 °C ($A_F/A_0 \approx$ 0.04) and 500 °C ($A_F/A_0 \approx 0.02$) and does not vary significantly over the dynamic range, which suggests that the growth of damage during straining is low. All the fracture surfaces are of the ductile type at both temperatures and whatever the strain rate /2,3,14/can be.
Qualitatively similar observations have been made on OFHC coppers (99.99 % Cu) /2-5/. Although the general trends, described above, are always present, the quantitative variations of ductility appear to be very sensitive to the nature and amount of impurities. Other investigations carried out on copper /15,16/, aluminum /17/ and austenitic steels /18/ generally show that ductility increases slowly with strain rate, although few data are available in the dynamic range at elevated temperatures. Steels with bcc structures /18,19/, as well as other bcc materials such as tantalum /13,20/, lead to more controversial results. Nevertheless, it seems that the most frequently observed behaviour within the dynamic range is that illustrated in Fig. 2a and b, i.e. a slow increase of ductility with strain rate. On the other hand, the overall ductility level at high strain rates can be lower than in the quasistatic range in some cases, e.g. tantalum at 20 °C /1,2,4,13/.
One of the main features of dynamic behaviour, which will be particularly relevant in the following discussion, is the flow rule of the material. A common characteristic which has been reported by several authors (for a review, see /2/), is the occurrence in the dynamic range of a linear relationship between the stress (at a constant strain) and the strain rate of the type $\sigma = \sigma_0 + \beta\dot{\varepsilon}$, where σ_0 and β are functions of strain. Such a flow rule has been first associated with the mechanism of dislocation damping, which was expected to be rate controlling in the dynamic range /21/. More recently, it has been proposed that the observed law could result from a combination of the thermally activated processes (e.g. the overcoming of the dislocation forest in copper) which are predominant at low strain rates and the mechanism of dislocation damping /2,22/. However, very recent results suggest that structure evolution with strain rate could be also a dominant factor /23/.

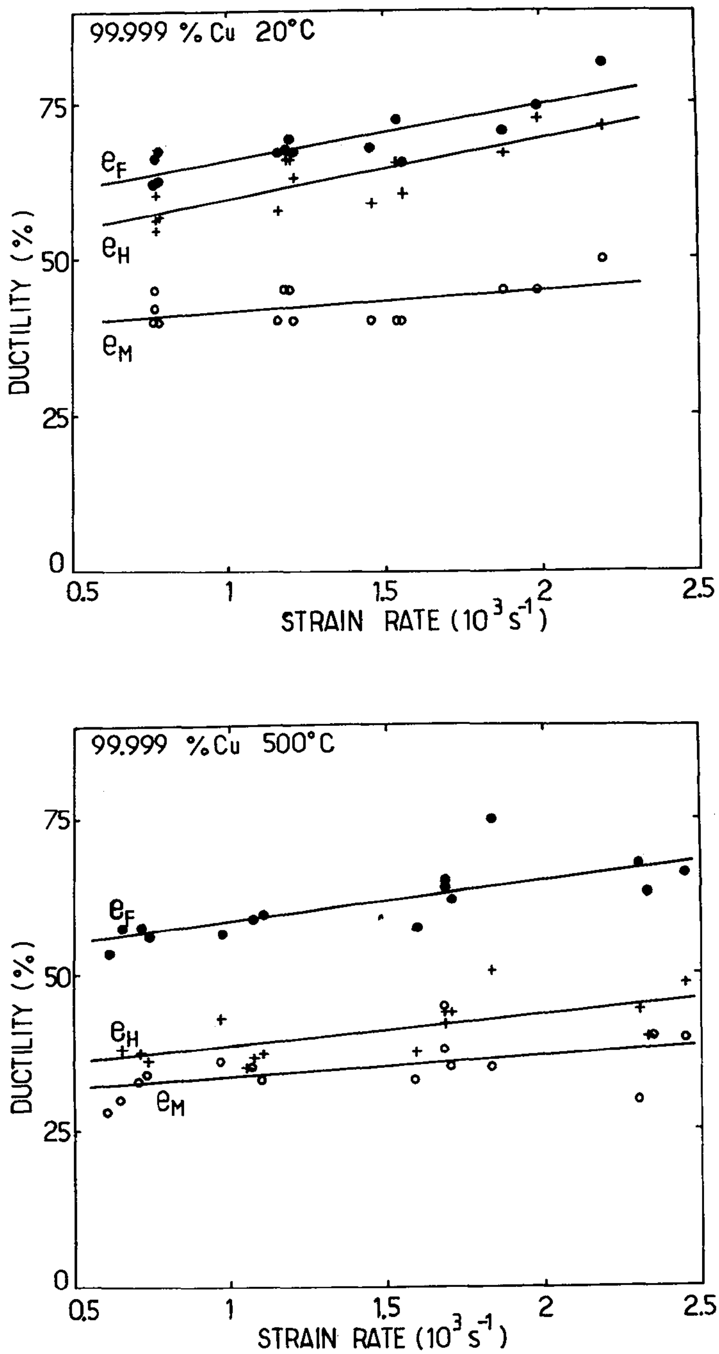

Fig. 2 - Strain rate dependence of the ductility of high purity copper at 20 °C (a) and 500 °C (b).

IV - REVIEW OF THE FACTORS DETERMINING DUCTILITY

The parameters which are likely to determine ductility during high rate tension can be attributed to three origins: (i) dynamic effects, viz. inertia effects and adiabatic heating, which are specific of deformation at high strain rate. Although they are not independent of the flow rule and microstructure of the material, these manifestations are essentially related to the conditions of the test; (ii) rheological effects, i.e. the influence of the material flow rule and (iii) metallurgical effects, including microscopic inhomogeneities and dynamic softening. Furthemore, each of the above parameters can act either on the development of damage during plastic straining (intrinsic ductility), on the elongation stability or possibly on both. However, in most cases, it is known that ductility is essentially controlled by the elongation stability, which in turn is determined by the material flow rule and by inertia effects (see below). On the other hand, metallurgical effects are expected to mostly affect the damage and fracture processes and will therefore have a minor influence on the overall ductility (except in a few cases where intrinsic ductility is the controlling parameter /2/). For clarity, the above various effects are now discussed in two parts, according to whether they are expected to decrease or to enhance ductility.

IV.1 - Factors decreasing ductility

a) Macroscopic inhomogeneities: at the beginning of a dynamic tension test, the strain and stress are inhomogeneous within the specimen /10/. This effect is due to inertia and more precisely to the finite velocity of elastic and plastic waves. It has been first investigated theoretically by von Karman and Duwez /24/ and then by several authors. In a recent paper, Regazzoni et al. /25/ have simulated numerically the tensile test of an initially cylindrical and homogeneous bar. They have shown that the inhomogeneity is reduced by strain hardening and strain rate sensitivity, as well as by an increase of the flow stress. Conversely, it is enhanced by an increase of the material density and a larger elongation rate or acceleration. As a consequence, at very high strain rates, fracture may occur at the vicinity of the impact head of the specimen before homogeneous deformation is reached. This phenomenon is commonly observed and restricts the strain rate range of interest of the dynamic tension test (about $3\times10^3\ s^{-1}$ in the case of copper).

b) Adiabatic heating: the duration of a dynamic tension test ranges from 0.1 to 1 ms such that it can be considered as adiabatic to a first approximation. Assuming that the elongation remains homogeneous, it is very simple to get a rough estimation of the temperature increase $\Delta\theta$ /2,11/: for a copper specimen strained to $\varepsilon = 0.4$ at $\dot{\varepsilon} = 3000\ s^{-1}$, $\Delta\theta = 31\ ^\circ C$ and $16\ ^\circ C$ at room temperature and $500\ ^\circ C$, respectively. This slight uniform increase of temperature is not expected to modify significantly the flow rule of the material, and the test could still be considered as isothermal. On the other hand, this holds no more when a neck is growing, since strain and consequently deformation heating then concentrate within a reduced length of the specimen. This effect has recently been taken into an account by Fressengeas and Molinari /26/ together with inertia effects. The calculations show clearly that adiabatic heating decreases ductility by a factor ≈ 0.2 when the data relative to a 304 type stainless steel deformed at room temperature are used. In addition, it should be noted that adiabatic heating promotes the formation of shear bands /27/.

c) Microscopic inhomogeneities: in this paragraph, we point out a few metallurgical factors that are known to appear at high strain rates. Although their relation with ductility is not yet clear, they are expected to decrease it. Experiments carried out on copper single crystals /28/ or polycrystals /29/ have revealed the occurrence of mechanical twinning at high strain rates. This deformation process, which is not observed in copper deformed under quasistatic conditions, is likely to induce local strain inhomogeneities and to increase the intergranular stresses. The latter in turn can promote brittle fracture by cleavage. It should be noted, however, that deformation twins have not yet been observed in copper specimens strained in uniaxial tension at the strain rates considered in the present paper. An other factor which could favour local inhomogeneities is the anisotropy of

single crystals. It has been observed by Chiem /28/ that single crystals of copper and aluminum submitted to a magnetic impulse exhibit a considerable anisotropy of strain. This should not be the case at low strain rates, according to the classical theories predicting the single crystal yield surface.
Among the above factors, macroscopic inhomogeneities and adiabatic heating are expected to decrease the stability of the elongation and therefore the uniform elongation e_H. Microscopic inhomogeneities, on the other hand, are more likely to promote damage initiation , i.e. to reduce intrinsic ductility.

IV.2 - Factors increasing ductility

a) Influence of inertia on necking: a complete study would require the analysis of the whole tension test, where the velocity prescribed to the moving extremity of the specimen increases from zero to a steady state value. During this first stage of elongation, which is about 10 µs long, material elements are submitted to very large accelerations, which on the one hand are responsible for the inhomogeneous deformation of the specimen and on the other hand can strongly influence the growth and further evolution of a preexisting defect. However, a numerical simulation of the whole test, extending from the initial state (no velocity prescribed at the sample extremity) to the stage where only the neck deforms would require very large computing times.

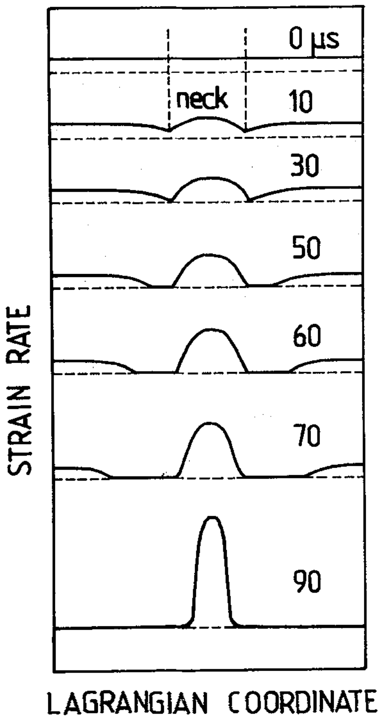

Fig. 3 - Typical evolution with time of the strain rate distribution in a tensile specimen with an initial defect (neck), showing the three stages of the instability process: case of a strain hardening rate insensitive material /32/. The origin of strain rate is represented by the dashed line.

Nevertheless, some computer calculations have been carried out recently to study the rate of growth of an initial imperfection in the case where a constant velocity is prescribed to the mobile end of the specimen. Taylor et al. /30/ have shown that inertia effects decrease the rate of localization within the specimen. This trend has been confirmed recently by Huang and Lee /31/ and by Fressengeas and Molinari /26/. Regazzoni et al. /25,32/ have investigated in greater detail the evolution of a specimen with an initial defect centered at the middle of the latter, starting at the precise time where the Considère criterion is verified, i.e. at the onset of macroscopic instability. They have shown that three stages can be distinguished (Fig.3), viz. (i) the strain rate increases in the defect while it decreases in the section adjacent to it; (ii) the "uniform" part is progressively unloaded starting from the neck to the ends of the specimen (Mott-Lee release waves); (iii) only the neck continues to deform until fracture occurs. During stages (i) and (ii) the "uniform" region undergoes a post-uniform strain due to inertia effects. Furthermore, the authors have shown that strain rate sensitivity also induces a post-uniform strain, which is of the same order of magnitude as the previous one (Fig.4). From the above results, it can be concluded that inertia effects increase ductility (precisely the uniform elongation) by delaying the neck growth and by controlling the unloading of the uniform part.
A simplified approach to the analysis of the whole test has been developed in /2/: the sample is divided into two parts, the one with the smaller cross-section ("neck")

being located at the fixed or at the impact head. The variations with time of the cross-sections in the "uniform part" and in the "neck" have been calculated by using a variational principle. The velocity of the moving extremity of the specimen was specified to increase from zero to the steady state value in 10 µs. It has been found that inertia reduces the growth of the defect when it is located near the fixed end of the specimen, and conversely when it is located near the impact head. The interpretation of this result is difficult because the model combines two different effects: on the one hand, the trend of the specimen to deform inhomogeneously at the beginning of the test, and on the other hand the effect of inertia on the growth of a defect. It was shown in IV.1a that the first one tends to limit ductility; hence it appears to be predominant when the defect is located near the mobile head of the specimen. Conversely, the second effect is a stabilizing factor as demonstrated above; it appears therefore that it is the major factor when the defect is located at the fixed end.

b) Influence of inertia on damage: it could be thought that damage can decrease with increasing strain rate, thus providing a better intrinsic ductility. Density measurements carried out on copper, however, do not clearly support this hypothesis /2/, or even suggest the opposite /33/.
Both the initiation and growth stages of damage could be influenced by high strain rate and the associated phenomena. Concerning the first one, no theoretical results are available to date, to the knowledge of the authors. However, the strain and stress inhomogeneities associated with hard inclusions can be expected to be reduced at high strain rates, due to the rapid increase of the flow stress at high strain rates, thus delaying the initiation of damage. Concerning the second stage, it has been shown /34/ that inertia effects would reduce the growth of spherical holes during plastic straining. It should be mentioned in addition that one of the effects of inertia is to develop a radial stress within the specimen /2,26/. It has been shown that the latter is maximum at the specimen axis and is always compressive during a test at constant strain rate ($\ddot{\varepsilon} = 0$). On the other hand, when ε is positive, the radial stress is tensile. This situation occurs at the mobile extremity at the beginning of the test and in the neck during the instability process. A tensile stress theoretically tends to increase the rate of growth of the cavities. However, the influence of the above radial stress, in particular when it is combined with the three-dimensional state of stress due to strain gradients (Bridgman effect) requires further investigation. It can be concluded that the net effect of high strain rates on damage during the tension test cannot be clearly predicted to date.

c) Flow rule of the material: it is well known that the constitutive equation of the material plays a predominant role in determining the elongation stability and thus ductility. In particular, stability increases strongly with the strain hardening exponent $n = \partial \ln\sigma/\partial \ln\varepsilon$ and with the strain rate sensitivity exponent $m = \partial \ln\sigma/\partial \ln\dot{\varepsilon}$.
Concerning strain hardening, the few experimental data are not very reliable due to the poor quality of the load-time diagrams obtained at high strain rates/2,13/. The measurements made on copper, however, display values of "n" very similar to that observed at low strain rates /4/. Nevertheless, an increase of "n" could be expected at high strain rates since dynamic recovery is less efficient in that range.
Turning now to the strain rate sensitivity, it should be noted that the linear stress-strain rate relation commonly observed at high strain rates, i.e. $\sigma = \sigma_0 + \beta\dot{\varepsilon}$, is associated with a parameter $m = \beta\dot{\varepsilon}/(\sigma_0 + \beta\dot{\varepsilon})$ which is strain rate dependent. It precisely increases with strain rate and thus enhances the elongation stability. This effect has been numerically investigated by Regazzoni et al. /25,32/. It has been clearly shown that the rate of neck growth is reduced when strain rate increases (Fig. 4). For copper, the net increase in uniform elongation at fracture has been found to be of the same order of magnitude that the increase resulting from inertia and to be consistent with the experimental results.
Finally, it could be argued that the employed m-values are questionable, since the latter are experimentally determined from tests conducted at various constant strain rates. In fact, rapid changes in strain rate can occur during the dynamic tensile test, for instance in the neck during the final stage of strain localization. For instantaneous rate changes, the appropriate value of strain rate sensitivity m_s

should be determined by strain rate jump tests, i.e. at constant structure. It is known that the m_s-values obtained this way are lower than the above m-values (see e.g. /35,36/). It is probable that the effective rate sensitivity should be intermediate between m_s and m /32/.

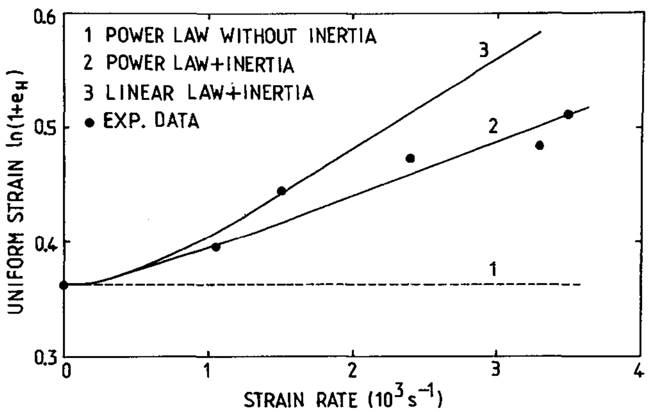

Fig. 4 - Strain rate dependence of the uniform (homogeneous) strain in tension, showing the respective influences of the flow rule and inertia effects /32/. The experimental data are relative to an OFHC copper (99.99 % Cu) strained at 20 °C.

d) dynamic softening: dynamic recovery is always present at low strain rates, even at temperatures much less than 0.5 T_m. It reduces the dislocation content of the material by the way of the annihilation of dislocations of opposite signs. Such a mechanism requires extensive climb and cross-slip of the dislocations. Since these thermally activated processes are time dependent, their efficiency at high strain rates is expected to be reduced due to the short duration of the tests.
There is no experimental evidence to date that dynamic recrystallization, which is present in low stacking fault energy metals at elevated temperature (e.g. copper at 500 °C) in quasistatic conditions operates at high strain rates as well. The strain where dynamic recrystallization starts (which nearly corresponds to the first peak of the stress-strain curve) is known to increase with $\dot{\varepsilon}$ /37/. Extrapolation of the available data to the range of high strain rates suggests that dynamic recrystallization cannot occur under those conditions, although some microscopic heterogeneities, such as mechanical twinning, could favour the germination of new grains. If they were present, both the softening processes above would reduce the dislocations concentrations and hence the local stresses within the material, thus increasing its intrinsic ductility.

V - SUMMARY

The main factors expected to determine ductility at high strain rates which were studied in the previous sections are reported in Fig.5. Their dynamic, metallurgical or rheological character, as well as their expected effects either on intrinsic ductility or elongation stability are also represented schematically.
From the above discussion, some general conclusions can be drawn:
(i) ductility mostly increases with strain rate in the dynamic range considered;
(ii) this is essentially due to an increased stability of the elongation process, i.e. to a slower development of the neck;
(iii) the latter may result alternatively from an increasing strain rate sensitivity associated with the occurrence of the linear stress-strain relationship, from inertia effects, or from both effects;
(iv) it is suggested that these conclusions hold within the dynamic range for both fcc and bcc structures.

<table>
<tr><th>DYNAMIC EFFECTS</th><th>METALLURGICAL EFFECTS</th><th>RHEOLOGICAL EFFECTS</th><th></th><th></th><th></th></tr>
<tr><td>MACROSCOPIC INHOMOGENEITIES
ADIABATIC HEATING</td><td></td><td></td><td>DECREASE</td><td rowspan="2">STABILITY</td><td rowspan="4">D U C T I L I T Y</td></tr>
<tr><td>INFLUENCE OF INERTIA ON STABILITY</td><td></td><td>FLOW RULE</td><td>INCREASE</td></tr>
<tr><td></td><td>MICROSCOPIC INHOMOGENEITIES</td><td></td><td>DECREASE</td><td rowspan="2">INTRINSIC DUCTILITY</td></tr>
<tr><td>INFLUENCE OF INERTIA ON DAMAGE</td><td>DYNAMIC SOFTENING</td><td></td><td>INCREASE</td></tr>
</table>

Fig. 5 - Table summarizing the main factors influencing ductility and their respective origins and effects.

ACKNOWLEDGEMENTS.

This work was supported by the CETAM (DTAT) under contract numbers 78-02012, 81-02049 and 83-02102. The authors acknowledge Dr. J.C. Giannotta for his contribution to the present article and Drs. J.P. Ansart and R. Dormeval from the Commissariat à l'Energie Atomique (Centre de Bruyères-le-Châtel) for their participation to the experimental program.

REFERENCES

/1/ Kawata, K., Hashimoto, S. and Kurokawa, K., in High Velocity Deformation of Solids, eds. K. Kawata and J. Shioiri, Springer Verlag, New York (1978) 1.
/2/ Regazzoni, G., Thèse de Doctorat d'Etat, Université de Grenoble (1983).
/3/ Regazzoni, G., Montheillet, F., Dormeval, R. and Stelly, M., in Deformation of Polycrystals: Mechanisms and Microstructures, eds. N. Hansen et al., Riso National Laboratory, Roskilde (1981) 343.
/4/ Regazzoni, G. and Montheillet, F., in Mechanical Properties of Materials at High Rates of Strain, Inst. Phys. Conf. Ser. N° 70, The Institute of Physics, London (1984) 63, Edit J. HARDING.
/5/ Giannotta, J.C., Regazzoni, G. and Montheillet, F., in Strength of Metals and Alloys, Pergamon Press, New York (1985) in press.
/6/ Grady, D.E., Kipp, M.E. and Benson, D.A., in Mechanical Properties of Materials at High Rates of Strain, Inst. Phys. Conf. Ser. N° 70, The Institute of Physics, London (1984) 315, Edit J. HARDING.
/7/ Albertini, C. and Montagnani, M., in Mechanical Properties of Materials at High Rates of Strain, Inst. Phys. Conf. Ser. N° 21, The Institute of Physics, London (1974) 22, Edit J. HARDING.
/8/ Lindholm, U.S. and Yeakley, L.M., Experimental Mech. 8 (1968) 1.
/9/ Kawata, K., Hashimoto, S., Kurokawa, K. and Kanayama, N., in Mechanical Properties of Materials at High Rates of Strain, Inst. Phys. Conf. Ser. N° 47, The Institute of Physics, London (1979) 71, Edit J. HARDING.
/10/ Dormeval, R. and Stelly, M., in Mechanical Properties of Materials at High Rates of Strain, Inst. Phys. Conf. Ser. N° 47, The Institute of Physics, London (1979) 154, Edit J. HARDING.

/11/ Regazzoni, G., Giannotta, J.C., Montheillet, F. and Dormeval, R., Cahiers du Groupe Français de Rhéologie 6 (1982) 205.
/12/ Ghosh, A.K., J. Engng Mater. Techol. (1977) 264.
/13/ Giannotta, J.C., Regazzoni, G. and Montheillet, F., International Conference on Mechanical and Physical Behaviour of Materials under Dynamic Loading, Les Editions de Physique, Paris (1985).
/14/ Giannotta, J.C., Regazzoni, G. and Montheillet, F., Journées d'Etudes du GAMI: Deformation des Métaux aux Grandes Vitesses, Paris (1983) in press.
/15/ Nadai, A. and Manjoine, M.J., J. Appl. Mech. 8 (1941) 77.
/16/ Sakui, S., Nakamura, T. and Nunomura, S., Proceedings of Third Japan Congress on Testing Materials (1961) 74.
/17/ Lindholm, U.S., Bessey, R.L. and Smith, G.V., J. Mater., JMLA 6 (1971) 119.
/18/ Meyer, L.W., Kunze, H.D. and Seifert, K., Shock Waves and High Strain Rate Phenomena in Metals, eds. M.A. Meyers and L.E. Murr, Plenum Press, New York (1981) 51.
/19/ Harding, J., J. Iron Steel Inst. 6 (1972) 425.
/20/ Hoge, K.G. and Mukherjee, A.K, J. Mater. Sci. 12 (1977) 1666.
/21/ Kumar A., Hauser, F.E. and Dorn J.E., Acta Metall. 16 (1968) 1189.
/22/ Follansbee, P.S., Regazzoni, G. and Kocks, U.F., in Mechanical Properties of Materials at High Rates of Strain, Inst. Phys. Conf. Ser. N° 70, The Institute of Physics, London (1984) 71, Edit J. HARDING.
/23/ Follansbee, P.S., Kocks, U.F. and Regazzoni, G., International Conference on Mechanical and Physical Behaviour of Materials under Dynamic Loading, Les Editions de Physique, Paris (1985).
/24/ von Karman, T. and Duwez, P., J. Appl. Phys. 21 (1950) 987.
/25/ Regazzoni, G., Johnson, J.N. and Follansbee, P.S., to be published.
/26/ Fressengeas, C. and Molinari, A., Acta Metall. 33 (1985) in press.
/27/ Costin, L.S., Crisman, E.E., Hawley, R.H. and Duffy, J., in Mechanical Properties of Materials at High Rates of Strain, Inst. Phys. Conf. Ser. N° 47, The Institute of Physics, London (1979) 90, Edit J. HARDING.
/28/ Chiem, C.Y., Thèse de Docteur Ingénieur, Université de Nantes, (1976).
/29/ Stelly, M., in Mechanical Properties of Materials at High Rates of Strain, Inst. Phys. Conf. Ser. N° 47, The Institute of Physics,London (1979) 252, Edit J. HARDING.
/30/ Taylor, J.W., Harlow, F.H. and Amsden, A.A., J. Appl. Mech. 45 (1978) 105.
/31/ Huang, Z.Q. and Lee, L.H.N., Int. J. Solids. Struct. 20 (1984) 897.
/32/ Regazzoni, G, Johnson, J.N. and Follansbee, P.S., Proceedings of the Symposium on Plastic Instability, Considère Memorial, Paris (1985) in press.
/33/ Giannotta, J.C., Microthèse de DEA, Ecole des Mines de Paris (1983).
/34/ Glennie, E.B., J. Mech. Phys. Solids 20 (1972) 415.
/35/ Mécking, H. and Kocks, U.F., Acta Metall. 29 (1981) 1865.
/36/ Perdrix C., Thèse de Docteur Ingénieur, Ecole des Mines de Paris, (1983).
/37/ Mc Queen , H.J. and Jonas, J.J., in Treatise on Materials and Alloys, ed. R.J. Arsenault, Academic Press, New York 6 (1975) 393.