Cold drawn steel wires—processing, residual stresses and ductility—part I: metallography and finite element analyses

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Received in final form 29 September 2005

ABSTRACT Cold drawing steel wires lead to an increase of their mechanical strength and to a drop of their ductility. The increase of their mechanical strength has long been related to the reduction of the various material scales by plastic deformation, but the mechanisms controlling their elongation to failure have received relatively little attention. It is usually found that heavily deformed materials show a tendency to plastic strain localization and necking. However, in this paper it is shown that, though the steel wires are plastically deformed up to strain levels as high as 3.5, a significant capability of plastic deformation is preserved in as-drawn wires. This apparent contradiction is resolved by the existence of residual stresses inside the wire. Finite element analyses have been conducted in order to show that residual stresses, inherited from the drawing process, are sufficient to produce a significant hardening effect during a post-drawing tensile test, without introducing any hardening in the local material behaviour. The main conclusion of this paper is that once the material has lost its hardening capabilities, residual stresses, inherited from the process, control the elongation of cold drawn wires. The finite element method allowed also the determination of the residual stress field that would lead to the best agreement between the simulated and the experimental stress strain curve of as-drawn wires.

Keywords EDXRD; elongation to failure; martensite; martensite transformation; residual stress.

INTRODUCTION

Cold drawn eutectoid steel wires are mainly employed by the cable industry for applications such as tyre reinforcement, suspension cables or springs. Cold drawing is applied in order to increase their mechanical strength. Cold drawing up to a strain of 3.5, for instance, increases their tensile strength up to 3000 MPa.1 At the highest plastic strain levels, their maximum strength is comparable to that of carbon fibres. Besides, steel wires are ductile, while carbon fibres are brittle. Ductility is an important parameter, on the one hand, since it allows one to bend or twist the wires permanently, which is required by various assembly processes (cabling for instance), and, on the other hand, since it provides resilience to the system.

Now, both ductility and strength are controlled by the drawing process. It is observed that tensile strength increases with the drawing strain level,7 while elongation to failure decreases significantly.

The mechanisms at the origin of the increase of tensile strength have long been discussed.3–5 As a result of a large plastic deformation, the material scales are drastically reduced, which leads to an increase of their mechanical strength.3–5 In the particular case of highly drawn eutectoid steel wires, their grain size (less than 1 μm) classes them in the group of ultrafine-grained materials, while the size of their lamellar substructure (less than 20 nm) classes them in the group of nanostructured materials. The refinement achieved by cold drawing at both scales is likely to contribute to the enhancement of their mechanical strength.

Besides, the mechanisms controlling their ductility have received relatively low attention. Recently, Zelin6 showed that the drop of elongation to failure of the wire is inherited from premature strain localization rather than from a reduction of the ductility of the material itself. The origin of that premature localization in tension is not completely understood. In,6 it is proposed that strain localization in tension might be inherited from the strain localization in shear bands that was shown to take place during
the drawing process. Strain localization in shear bands during processing has also been reported by various authors, in other ultrafine-grained materials and appear to be a general feature of these materials. Strain localization in tension could, therefore, be expected to appear at the very beginning of the tensile test, as a result of the existence of shear bands inherited from the drawing process. This is not the case. Cold drawn wires tested in tension display a significant hardening phase before plastic strain localization and failure occur. Moreover, the elongation to failure can be modified by applying mechanical treatments after cold drawing, such as straightening or low angle drawing, which are known to modify the residual stress state inside the wire. This indicates that residual stresses and elongation to failure are closely related.

The aim of the present paper is, therefore, to identify and discuss the mechanisms controlling the elongation to failure of cold drawn steel wires and to examine, in particular, the role of residual stresses.

RESIDUAL STRESSES AND FINITE ELEMENT ANALYSES

Material and processing

Tyre cords wires are obtained by cold drawing eutectoid steels. The drawing strain level (i.e. ln(Sf/Si) where Si is the initial section and Sf the final section) is commonly up to 3.5. During processing, the internal scales of the material are drastically reduced. These internal scales consist of: the diameter of the wire, the grain size and the size of the ferrite lamellae. At the beginning of the process, the steel is pearlitic with a lamellae thickness approaching 200 nm (Fig. 1(a)) and a grain size of 20 μm. After cold drawing up to a strain level of 3.5, the thickness of the ferrite lamellae is less than 20 nm (Fig. 1(b)), and the grain size is assumed to be less than 1 μm.

Simultaneously the morphology of grains is varying, but the evolution of the grain size and shape during cold drawing is not easy to determine in eutectoid steels, since grain boundaries are not easy to distinguish from lamellae (Fig. 1(b)). Therefore, a 0.1% C steel with an initial grain size of 20 μm was cold drawn in the same manner, in order to characterize the evolution of the grain morphology with cold drawing (Fig. 2(a) and (b)). In the initial microstructure, grains are equiaxed. After cold drawing, the morphology of grains was characterized in a longitudinal section (Fig. 2(a)) and in a cross section (Fig. 2(b)). It appears clearly that the evolution of the geometry of the grains and of the wire are not comparable. Grains are elongated in the drawing direction but they also appear to be curled about the wire axis in a cross section (Fig. 2(b)). This effect results from the inadequacy of the axisymmetric deformation imposed by the drawing process and of the cubic symmetry of the ferritic phase. The main consequence of this phenomenon is that the shortest dimension of grains diminishes during the drawing process with a higher rate than the wire diameter.

The evolution of the morphological texture of grains is associated with an evolution of the crystallographic texture. According to the literature, this texture is completely stabilized around a drawing strain of 1.8. In the stabilized configuration the texture consists primarily of a fibre texture with the normal to the (110) planes aligned with the drawing axis.

Ductility, mechanical properties and processing

First of all the evolution of the elongation to failure of the material with respect to the level of cold drawing was determined, both on a 0.7% and a 0.8% carbon steel (Fig. 3(a)). The trend is similar in both cases. The elongation to failure is rather high in the initial state (Δp% = 8%), because wires are annealed before the drawing
process. Immediately after the first die, the elongation to failure drops significantly ($Ap\% = 2\%$), due to the saturation of the material with dislocations inherited from the large plastic strain applied during the first step of the drawing process. Then the evolution of $Ap\%$ is somehow chaotic up to a drawing strain of 1.8, for which the texture is fully stabilized. Above this critical drawing level, the decrease of the ductility is moderate but regular, the elongation to failure is close to 2\%, for a drawing strain of 2 and goes down to 1\% for a drawing strain of 4.

In the same range (i.e. $2 < \ln(S_c/S) < 4$) the ductility in torsion $\gamma_p\%$ and the reduction in area $Z\%$ have been measured for increasing drawing levels from a drawing strain level of 1.8, from which the texture is stabilized (Fig. 3(b)). The reduction in area was calculated as: $Z = 2\ln(\phi/\phi_i)$, where $\phi$ is the diameter of the wire and $\phi_i$ the diameter measured after failure in the necking area. The ductility in torsion $\gamma_p\%$ was calculated as the number of turns to failure per unit length multiplied by the perimeter of the wire. It is observed that both the reduction in area and the ductility in torsion remain remarkably constant while the elongation to failure is divided by a factor two (Fig. 3(b)). Besides, both $\gamma_p\%$ and $Z\%$ are much larger than the elongation to failure (1\% < $Ap\%$ < 2\%, $Z\% = 60\%$ and $\gamma_p\% = 120\%$). This indicates that the intrinsic ductility of the material is not significantly modified by plastic deformation between $\varepsilon = 1.8$ and $\varepsilon = 4$.

In order to check this assumption, observations of the failure mechanisms have been performed in the necking area. Wires, drawn between strain levels of 0 and 3.5, were subjected to tensile tests. The fracture surface was observed using a scanning electron microscope. The failure is always typically ductile. At a drawing strain of 3.5 for instance, voids are well formed (Fig. 4), though the elongation to failure is less than 1\%. Their shape is curled and elongated (Fig. 4(a)), close to that of grain boundaries (Fig. 2). In order to better understand the formation of voids, the external surface of wires was mechanically polished, and the development of slip bands was studied.

Below a drawing strain of about 1.5, slip bands can either be constrained inside ferrite lamellae or may cross a whole colony of lamellae. Above a drawing strain of 1.5, the lamellar structure is no more a barrier for dislocations. For a drawing strain of 3.5 the interlamellar spacing is less than 20 nm (Fig. 1(b)) but slip band traces are continuous over domains that can be as large as 2 μm (Fig. 4(b)). Moreover, it was also confirmed that voids appear at the intersection between slip bands and grain boundaries.

Consequently, the drop of elongation to failure with the drawing strain level cannot be attributed to a more brittle character of failure, but is inherited from a premature strain localization in tension (Fig. 3(b)), which is consistent with the results of Zelin. Now the problem remains to understand first of all the origin of that premature strain localization, and secondly why a significant ductility is preserved at high drawing levels.

According to the condition for localization of Rudnicki and Rice, the plastic strain localization inside a wire subjected to a uniaxial tensile test occurs for a critical value of the hardening rate (in Ref. [13] $b_p = -E/4$ where $E$ is the Young’s modulus and $b_p$ the hardening rate). Therefore, the evolution of the hardening rate of the material with the drawing level is a key parameter to understand the evolution of the elongation to failure of the wire.

Consequently, the evolution of the mechanical strength with the drawing level was also characterized, using both post-drawing tensile tests and indentation tests.

It is observed, in agreement with previous results in the literature, that the reduction of the internal scales of the material is associated with a drastic increase of the mechanical strength of the wires. Cold drawing up to 3.4 leads, for instance, to an increase of the tensile strength by a factor 2.5 (Fig. 5(a)).

The same trend is obtained by using hardness tests in order to characterize the evolution of the mechanical strength with cold drawing (Fig. 5(b)). The Vickers hardness increases significantly with cold drawing. However, while the tensile strength is increased by a factor 2.5 by
Fig. 3 (a) Evolution of the elongation to failure versus drawing strain of a 0.7% C and a 0.8% C steels. (b) Comparison of the evolution of the reduction in area $Z\%$, the elongation to failure $Ap\%$ and the torsion ductility $\gamma_p\%$. The results were normalized by their value at a drawing strain level of 1.8, namely $Ap(\varepsilon = 1.8) = 2.2\%$, $\gamma_p(\varepsilon = 1.8) = 120\%$, $Z(\varepsilon = 1.8) = 60\%$

Fig. 4 Tensile test at room temperature of a 0.8% C steel wire, cold drawn at $\varepsilon = 3.5$. (a) scanning electron micrograph of the fracture surface, the failure is typically ductile. (b) slip band traces at the surface of the sample below the necking zone, voids formed at the intersection between a grain boundary and a slip line.

Fig. 5 (a) Consolidation of a 0.8% C and a 0.7% C steel by cold drawing. Tensile strength of wires versus drawing strain, (b) Vickers Hardness of a 0.8% C steel versus drawing strain, comparison of HV$_{0.1}$ and HV$_{0.2}$
cold drawing up to 3.5, HV0.1 increases by a factor 2.1 while HV0.2 increases by only a factor 1.7.

A last set of experiments provides the explanation of this apparent contradiction between the experimental results. Nanohardness tests were performed for three drawing levels, $\epsilon = 0$, $\epsilon = 2$ and $\epsilon = 4$. Five experiments have been conducted in each case, though only one is plotted in Fig. (6). The curves are very reproducible.

In the microhardness regime, on the right side of the graph, the hardness is not depth dependent. In that regime, a factor 1.7 is observed between the hardness in the initial state ($\epsilon = 0$) and after a cold drawing strain of 4. This is consistent with the evolution of HV0.2, displayed in Fig. 5(b).

Below that microhardness regime, hardness becomes strongly depth dependent. A very pronounced indentation size effect is observed when the indentation depth increases. These results are consistent with the fact that HV0.1 is always above HV0.2 (Fig. 5(b)), since the depth of the imprints performed using HV0.2 is larger than that performed using HV0.1.

The apparent contradiction in Fig. 5 can be explained by the existence of a softening effect at the onset of plasticity. During an indentation test, the indenter imposes the displacements at the surface, which makes possible the measurement of a large softening effect. On the contrary, during a tensile test, softening cannot be observed since softening leads to strain localization, necking and failure.

The softening effect during nanoindentation test is usually reported as the indentation size effect IDE, and is related to the motion of dislocations inside the grain. However, in the present case, and in particular for a drawing level $\epsilon = 4$, the typical volume plastically deformed during the indentation is far above the volume of one grain. For $\epsilon = 4$, in particular, the grain size is less than 500 nm, while the indentation depth is up to 1.5 $\mu$m and its contact area up to 7.5 $\mu$m $\times$ 7.5 $\mu$m. The volume of material plastically deformed during the indentation, calculated roughly as being above 2.5 times the volume of the imprint, should, therefore, contain more than 30 grains for an indentation depth of 0.5 $\mu$m and more than to 900 grains for an indentation depth of 1.5 $\mu$m. At this scale the material can, therefore, be considered as homogeneous. Therefore, in this particular case, the softening effect observed during the indentation test can, therefore, be considered as a softening effect of the polycrystalline material.

The existence of a peak and softening effect at the onset of plasticity is also consistent with industrial practice. As a matter of fact, mechanical treatments such as straightening or low angle drawing are applied after cold drawing in order to increase the ductility of the wires. Low angle drawing is analogous to skin passing, and straightening consists of bending the wire alternately, in order to impose an axial reverse plastic strain below the surface of the wire. In both cases, the bulk of the wire remains elastic, which avoids the development of strain localization. Below the surface, the plastic strain as imposed by the mechanical treatment is enough to go beyond the peak, which delays strain localization during subsequent tensile tests.

It is observed that these treatments increase the elongation to failure and reduce the maximum strength, which is consistent with the existence of a peak.

This effect is well known. The peak is usually attributed to the pinning of dislocations by the carbon solid solution, but such a mechanism does not explain the observed dependence of the phenomenon on the drawing level (Figs 5(c) & 6). The present experimental results indicate also that the peak is inherited from a scale effect.

This scale effect will not be discussed here, as it has been reported by numerous authors on ultrafine grains materials, that a tendency to necking is observed that is explained by the loss of the hardening capabilities of the material due to severe plastic deformation. For instance, recent results have been published by Wei et al. They have studied the behaviour of bulk ultrafine grain iron processed by ECAE under quasi-static and dynamic compression. After annealing, their material displays a hardening phase typical of bcc materials. In as-processed conditions, hardening disappears, and if ECAE is followed by rolling, softening is even observed in quasi-static and dynamic conditions.

To summarize in a few words, the appearance of a peak stress at the onset of plasticity when the grain size diminishes is consistent with the results of the nanohardness tests, with industrial practice, and explains well the drop of elongation to failure with cold drawing. Moreover, it is consistent with the general tendency of ultrafine-grained material to form shear bands.

However, during tensile tests conducted on as-drawn wires, a large hardening phase is always observed before necking (Fig. 7(a)). Moreover, the rate of hardening $H$
between the conventional plastic strain (i.e. 0.2%) and $Ap\%$, increases significantly when the drawing level increases and is directly proportional to $R_m$ (Fig. 7(b)). This result is apparently in contradiction with the existence of a peak inherited from the interaction between dislocations and the carbon solid solution and with the loss of hardening capabilities of ultrafine-grained materials processed by severe plastic deformation.

The existence of residual stresses, inherited from the drawing process, is expected to solve this contradiction.

**Residual stresses: finite element analyses**

The fact that the rate of hardening is proportional to the strength of the material leads one to think that this effect could be inherited from residual stresses. As a matter of fact, it is well known that residual stresses remain inside the wire after drawing. These residual stresses are known to modify the fatigue limit of cold drawn wires.\(^{17}\) The axial stress ($\sigma_{zz}$) is compressive at the centre and tensile below the surface.\(^{18}\) Therefore, the local threshold for plasticity is different at the centre and below the surface.

This effect is illustrated schematically in Fig. 8. This figure was constructed as follows for a drawing level equal to 3.5. At the local scale, the material is assumed to be elastic up to an axial stress level equal to the maximum strength as measured on wires, that is, 2770 MPa. Then the range of plastic strain over which softening may occur (around 2%) is evaluated from the observations of the consequences of mechanical treatment such as straightening. Finally, the amplitude of softening in tension was roughly evaluated with a rule of thumb using the amplitude of softening during a nanohardness test (Fig. 6). A residual stress level is assumed at the surface and at the centre. The stress–strain curve as determined on the wire (solid line) is the average of the local stress–strain curves at the surface (circle) and at the centre (squares).

During a tensile test, the applied stress is superimposed on the residual stresses (Fig. 8). Consequently, at the beginning of a tensile test, the whole wire is elastic and the slope of the average stress–strain curve is equal to the elastic modulus of the material. Then, the material at the surface is plastically deformed, while the centre...
remains elastic because the applied stress is superimposed on compressive residual stresses. Therefore, the slope of the stress–strain curve deviates from the elastic modulus and diminishes progressively, but remains positive though the material is softening at the local scale. This deviation increases progressively until the wire is plastically deformed up to its centre. Up to this point, hardening is observed at the macroscale on the average stress–strain curve (Fig. 8, solid line). Then softening appears at the macroscale also, which leads to plastic strain localization and necking.

It is shown using this schematic illustration that though the material has lost its hardening capabilities, hardening can be observed at the macroscale, up to the yield point of the centre of the wire. But from that point necking occurs and consequently the failure of the sample. Therefore, the order of magnitude of the axial residual stress at the centre of the wire can be evaluated as being the maximum strength (here 2770 MPa), minus the stress range at the centre. This range is the product of the maximum tensile strain (here εzz = 2.3%) by the Young’s modulus, if we assume that before \( R_m \), the centre of the wire is elastic. According to the present experiments (Figs 3(a) \& 5(a)), the order of magnitude of the axial residual stress at the centre of the wire should be close to \( \sigma_{zz} = -2000 \) MPa.

In conclusion, residual stresses are large enough to modify significantly the stress–strain behaviour of wires, and explain the apparent contradiction between Figs 7 and 6. Consequently, once the material has lost its hardening capabilities because of the large plastic deformation imposed by the process, the ductility of the wires is preserved by residual stresses inherited from the drawing process.

A finite element model was built so as to generate residual stress fields by wire-drawing, and to simulate a tensile test on the wire with residual stresses. These calculations were performed in order to determine an order of magnitude of the complete residual stress field inside the wire, namely \( \sigma_{zz}(r), \sigma_{\vartheta\vartheta}(r) \) and \( \sigma_{rr}(r) \).

**Hypotheses**

The problem was assumed to be axisymmetric (Fig. 9(1)). Non-linear geometry was employed. The contacts between the die and the wire (Fig. 9(2)), and between the grips and the wire (Fig. 9(3)) were considered to be unilateral hard contact. The elements are linear quadrilaterals. The computations were performed using Abaqus.

Concerning the constitutive behaviour of the material, rough approximations had to be made. First of all, the crystalline texture was neglected in the problem. Elasticity and plasticity were assumed to be isotropic. The von Mises criterion was employed for all the calculations. Since the crystal is BCC, the error due to this approximation is not expected to be particularly large.

Secondly, the initial hardening phase, that we aim to understand, can be inherited from residual stresses but also from material hardening itself. Two types of material hardening can be considered: effective hardening and kinematic hardening. Effective hardening corresponds to an increase of the dimension of the domain of elasticity of the material, while kinematic hardening (Bauschinger effect) corresponds to a displacement of this domain.

The large strains imposed during the drawing process are expected to saturate the effectivenes hardening of the material. Therefore, the initial hardening phase observed during a tensile test after drawing cannot be attributed to an effective hardening of the material. A slight isotropic hardening was nevertheless included in the model, whose slope (namely 800 MPa) was calculated using the slope of the consolidation curve around a drawing strain of 3.5 (Fig. 5(a)). With such a slope, the flow stress increases by only 8 MPa, when the plastic strain increases by 1%, which is roughly the value of the elongation to failure of the wire. This isotropic hardening remains negligible as compared with the tensile strength of wires (2770 MPa) at a drawing strain level of 3.5.
Besides this, kinematic hardening should also be considered. Kinematic hardening is inherited from the heterogeneity of plastic deformation within the material at various scales. The kinematic hardening at the scale of the wire, inherited from residual stresses due to the drawing process is the precisely object of this study. It does not have to be included in the material behaviour at the local scale, but should result from the existence of a residual stress field inherited from the heterogeneity of the plastic deformation inside the die.

Now kinematic hardening can also be been inherited from inter- and intra-granular heterogeneous plastic strain. Experiments are difficult because of the small diameter of the wires. However, due to the very small dimension of the grains, it is considered first of all that the diameter of the wires. However, due to the very small dimension of the grains, it is considered first of all that the diameter of the wires divided by the strain amplitude for which the kinematic hardening is saturated can be evaluated roughly as one burgers vector divided by the grain size D. This value is close to 0.02%, which is very small as compared with the plastic strain necessary to saturate the stress during a tensile stress strain curve (Fig. 7(a)). Therefore, no kinematic hardening was included in the constitutive behaviour of the material at the local scale.

Material softening is not included either in the simulations. As a matter of fact, it would not add significantly to the quality of the results (Fig. 8), while it would make the numerical resolution much more difficult.

As a conclusion, all the results of the simulations showed in the present paper include only a slight isotropic hardening. Only one material parameter remains to determine; the flow stress.

The key problem at this stage lies in knowledge of the material behaviour throughout the process. The industrial process is performed at a speed larger than 10 m.s\(^{-1}\). Consequently, the strain rate inside the die exceeds 10 000 s\(^{-1}\). In such conditions adiabatic heating appears. Adiabatic thermo-mechanical calculations were performed and it was calculated that the increase of the temperature is moderate (lower than 200° C). However, in these calculations, the wire was assumed to be cold at the entrance of a die. This is not the case in reality, because the wire goes consecutively through a large number of dies. Therefore, the temperature of the wire inside the last die depends on the cooling rate between two dies, which depends on numerous industrial factors. Various methods have been attempted to measure the temperature inside the last die, but until now this data remain uncertain.

Therefore, another approach to the problem was chosen. The material was assumed to obey two distinct constitutive behaviours. The first one was employed for the generation of residual stresses by wire drawing and the second one for the simulation of the tensile test. In both cases, the material is assumed to obey the von Mises criterion, with an isotropic hardening slope equal to 800 MPa. However, the flow stress is assumed to be different during the tensile test (\(\sigma_d;\) room temperature, low strain rate) and inside the die (\(\sigma_d;\) high temperature, high strain rate).

In situ measurements during the drawing process (Fig. 10) were performed in order to determine which flow stress \(\sigma_d\) should be employed for the simulation of the drawing process. For a given die angle, the output tension (\(T_d\)) on the wire is directly related to the flow stress of the material inside the die \(\sigma_d\) (here \(\sigma_d = 2 \cdot T_d/S\)), with the assumption that this flow stress is homogeneous inside the die.

Though the strain rate is very high during drawing (Fig. 10), the flow stress during drawing remains significantly below the flow stress as measured under quasi-static conditions (Fig. 10). Usually, for a given temperature, the effect of the strain rate is opposite, because of material viscosity. But in the present case, both the strain rate and the temperature increase when the wire goes through the die. The two effects are competing. The thermal effect appears to compensate the viscous effect.

Finally, for a drawing strain of 3.5, according to these experimental results, the flow stress employed for the generation of residual stresses by wire drawing was fixed at 1915 MPa. Other flow stresses have also been employed in order to test the sensitivity to this parameter (i.e. 500 and 2770 MPa).

Besides, the flow stress employed for the simulation of the tensile test with residual stresses, was fixed at 2770 MPa. Again, the sensitivity to this parameter was tested numerically. The results of these simulations are detailed the next section.
Results

In Fig. 11(a) are plotted the residual stress fields generated by the simulation of wire drawing using the von Mises criterion, an initial flow stress of 1915 MPa and a linear isotropic hardening with a slope equal to 800 MPa.

As expected, the axial stress (σzz) is compressive at the centre and tensile at the surface. Moreover, the axial residual stress at the centre of the wire is found to be equal to −2110 MPa, which is of the expected order of magnitude for a drawing strain of 3.5. At the surface the axial residual stress is positive and equal to 1500 MPa. The stress gradient inside the wire is found to be very large, that is, more than 7000 MPa per millimetre. The two other stress gradients, σrr and σθθ, are less pronounced, but in both cases the residual stress is compressive at the centre. The centre is subjected to a large compressive hydrostatic pressure equal to −1150 MPa.

In Fig. 11(b) the evolution of the axial stress field σzz(r) is plotted during a tensile test as simulated on the wire with the initial residual stress field as plotted in Fig. 11(a). Between an average stress applied on the wire comprised between 0 and 1054 MPa, the material behaviour is elastic. Above 1054 MPa, plastic strain occurs at the surface and the plastic zone invades progressively the whole section. At the saturation (⟨σzz⟩ = 2770 MPa), the stress profile appears to be reversed. The axial stress σzz is maximal at the centre. This reversal is due to the necking of the wire. These results confirm that residual stresses are wiped out at the saturation stress.

In Fig. 12(b) is plotted the comparison between the simulated stress–strain curves on the wire with the initial residual stress field as plotted in Fig. 11(a). Between an average stress applied on the wire comprised between 0 and 1054 MPa, the material behaviour is elastic. Above 1054 MPa, plastic strain occurs at the surface and the plastic zone invades progressively the whole section. At the saturation (⟨σzz⟩ = 2770 MPa), the stress profile appears to be reversed. The axial stress σzz is maximal at the centre. This reversal is due to the necking of the wire. These results confirm that residual stresses are wiped out at the saturation stress.

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Though the material behaviour introduced in the finite element analysis does not display hardening, a significant hardening phase is observed on the simulated stress–strain curve.

In the present case, the yield point at the centre corresponds to an average plastic strain of 0.7% which is in reasonable agreement with the experimental elongation to failure (0.9–1%) as measured for a drawing strain of 3.5 (Fig. 3(a)).

Now the second aim of this study was to evaluate an order of magnitude of the complete residual stresses inside the wire. For this purpose, the stress–strain curves obtained through the simulation is compared to the experimental one. The proper residual stress field is assumed to be that for which the agreement is the best. There are only two free parameters in the modelling that can be adjusted, namely the flow stress during drawing \(\sigma_d\) and the flow stress employed for the simulation of the tensile test \(\sigma_Y\). The role of each parameter is examined as follows:

First of all, the flow stress during drawing \(\sigma_d\) is fixed \((\sigma_d = 2770\,\text{MPa})\) while the flow stress employed for the simulation of the tensile stress is varied. In Fig. 12(a) the results of the FE analyses for \(\sigma_Y = 2770\,\text{MPa}\) and \(\sigma_Y = 4073\,\text{MPa}\) are compared with the experiment. The two numerical curves are exactly parallel, which indicates that the initial hardening phase is only dependent on the residual stress field. Besides, the saturation is reached in both cases for \((\sigma_{zz}) = \sigma_Y\). This is consistent with the fact that at the saturation, residual stresses are wiped out (Fig. 11(b)). Therefore, \(\sigma_Y\) is not a free parameter and should be adjusted to the experimental tensile strength \(R_m\).

Secondly, the flow stress employed for the simulation of the tensile test is fixed \((\sigma_Y = 2770\,\text{MPa})\) while the flow stress during drawing is varied. As expected, the hardening rate is strongly dependent on the level of the residual stresses. The saturation is reached for a plastic strain equal to 0.16% if \(\sigma_d = 500\,\text{MPa}\), equal to 0.7% if \(\sigma_d = 1915\,\text{MPa}\) and equal to 1.3% if \(\sigma_d = 2770\,\text{MPa}\). The best agreement is obtained for \(\sigma_d = 1915\,\text{MPa}\), which is the flow stress evaluated by in situ measurements during the drawing process (Fig. 10).

In conclusion, the residual stress field inside the wire is expected to be close to that plotted in Fig. 11(a), which was obtained by the simulation of wire drawing using a flow stress equal to 1915 MPa.

**CONCLUSIONS**

The main conclusions of this paper are as follows.

1. Cold drawing causes a drastic increase in the tensile strength of steel wires, and a pronounced reduction of their elongation to failure \(Ap\%\). However, the failure is always typically ductile with a pronounced reduction in area \((Z\% = 60\%)\). The reduction in area \(Z\%\) does not vary significantly with drawing strain.

2. Therefore, the drop in elongation to failure \(Ap\%\) during a tensile test corresponds to a premature necking of the wires, not to a more brittle character of failure. This premature necking is inherited from the loss of the material hardening capabilities due to the large plastic deformation applied during the wires processing.

3. Though the material plastically softens at the local scale (nanohardness tests), hardening is observed at the scale of the wire within the same range of plastic strain. Finite element analyses show that residual stresses inherited from processing are likely to preserve the hardening capabilities of the wire and thus its ductility. The surface is subjected to tensile residual stresses, while the centre is under compression. Consequently during a tensile test on the wire, the yield stress of the wire corresponds to the yield point at the surface, while the saturation stress of the wire (and necking) corresponds to the yield point at the centre. At this stage, an order of magnitude of the axial residual stress level at the centre of the wire can be determined using a rule of thumb.

4. In order to determine an approximation for the complete residual stress field inside the wire, the finite element method was employed in order to generate residual stresses (by the simulation of wire drawing) and then to simulate a tensile test on the drawn wire. The calculated residual stress field for which the simulated stress–strain curves provides the best agreement with the experimental stress–strain curve was determined. According to this approach, at the centre of the wire, the axial residual stress \(\sigma_{zz}\) would be around \(-2100\,\text{MPa}\) and the hydrostatic pressure around \(-1000\,\text{MPa}\).

5. The general conclusion of this paper is that plastic deformation induced by cold drawing does not reduce the intrinsic capability of deformation of the steel; on the contrary, it is enhanced. However, it leads to the loss of the hardening capabilities of the material, which leads to plastic strain localization in tension, necking and failure. However, once the material has lost its hardening capabilities, the elongation to failure of wires is preserved by residual stresses inherited from the drawing process.

**PROSPECTS**

The main prospects of this study are as follows. The complete residual stress field that provides the best agreement between the simulated and the experimental stress–strain curves was determined using the finite element method. It would be interesting at this
stage to compare these results with experimental measurements. Residual stresses are usually determined using X-ray diffraction or through the measurement of the wires elongation during the chemical dissolution of their surface. However, these methods provide only the measure of residual stresses at the surface, though we are mostly interested in the residual stress level at the centre. Therefore, the main prospect of this study is to determine the lattice parameter at the centre of the wire. This requires an energetic radiation. Besides, since the diameters of the wires are very small, and the stress gradient inside the wire is very large, a very small gauge dimension is also needed. This leads to the consideration of synchrotron radiations. These measurements have been performed and will be the object of a second paper.

In the present paper, the role of residual stresses on the wire elongation to failure was assessed. Also, it was shown that, provided that bounds are determined experimentally for the input parameters of the model, the finite element method allows a satisfactory prediction of the tensile stress–strain curve. Now, the logical prospect of this research is to analyse the role of industrial treatments such as low angle drawing and straightening, on both the tensile and torsion stress–strain curve.

Finally, it was shown that once the material has lost his hardening capabilities, the elongation to failure of wires is preserved by residual stresses inherited from the drawing process. This residual stress field depends on the flow stress during processing, which depends on various industrial parameters such as the strain rate and the temperature inside the die. The present results suggest, for instance, that an increase of the flow stress during processing would lead to an increase of the residual stress gradient and, therefore, to an increase of the elongation to failure. That increase of the flow stress could possibly be obtained by lowering the processing temperature. The feasibility and the effect of such a temperature control is also an interesting prospect for the future.

Acknowledgements

The authors would like to acknowledge the company Michelin for supporting this research and in particular E. Depraetere and P. Lesoeurs for fruitful discussions. The authors would also like to acknowledge Jean Luc Loubet (LTDS–Ecole Centrale de Lyon) for performing nanindentation tests and for fruitful discussions.

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