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IMPROVING THE FORMABILITY OF STAINLESS STEEL 321 THROUGH MULTISTEP DEFORMATION FOR HYDROFORMING APPLICATIONS

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ABSTRACT

Tube hydroforming (THF) is a well established process in the automotive industry and its application is being extended to the aerospace for manufacturing complex geometries. However, most of the alloys used in aerospace are high in strength and low in formability, which renders the application of THF more challenging. The objective of this paper is to present a method to increase the formability of an austenitic stainless steel. A multistep forming process was simulated through interrupted uniaxial tensile testing experiments to study the influence of the latter process on formability. The tensile test was divided into several deformation steps with a stress relief heat treatment after each forming step. The results indicated that the application of intermediate heat treatments considerably increased the formability of the stainless steel 321 alloy (SS321). Microstructure evolution as a function of deformation or heat treatment parameters was also investigated and revealed the formation of strain-induced martensite after the first deformation and heat treatment cycle without any deleterious effect on formability enhancement.

Keywords: aerospace alloys; stainless steel; multistep forming; formability; tube hydroforming.

AMÉLIORATION DE LA FORMABILITÉ DE L'ACIER INOXYDABLE 321 PAR DÉFORMATION MULTI-ÉTAPES POUR DES APPLICATIONS EN HYDROFORMAGE

RÉSUMÉ

L'hydroformage de tubes (THF) est un procédé bien établi dans l'industrie automobile et ses applications se sont étendues à l'industrie aérospatiale pour la fabrication de géométries complexes. Cependant, la plupart des alliages aéronautiques possèdent une grande résistance associée à une faible formabilité ce qui représente un défi en THF. L'objectif de cet article est de présenter une méthode d'amélioration de la formabilité d'un acier inoxydable austénitique. Un procédé de mise en forme multi-étapes a été reproduit à travers des essais de traction uniaxial interrompus pour étudier l'influence du procédé précité sur la formabilité. Les essais de traction ont été divisés en plusieurs tests de déformation suivis de traitements thermiques de relaxation des contraintes. Les résultats indiquent que l'application de traitements thermiques intermédiaires augmente considérablement la formabilité de l'acier inoxydable 321. L'évolution de la microstructure en fonction des paramètres de déformation ou de traitement thermiques a révélé la formation de martensite induite par déformation après le premier cycle de déformation/traitement thermique sans effet nuisible sur l'amélioration de la formabilité.

Mots-clés : alliages aéronautiques ; acier inoxydable ; mise en forme multi-étapes ; hydroformage.

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1. INTRODUCTION

Tube hydroforming is a forming process that uses a pressurized fluid, to plastically deform a given blank tube material into a desired shape. This technique presents many advantages compared to the traditional stamping and welding processes. It allows a lower weight to rigidity ratio as well as reducing the number of welds in an assembly which can considerably reduce the weight of the final product and also improve dimensional accuracy. Moreover, THF provides higher strength and surface quality in a part with a complex shape [1]. It is also well known for its reduced tooling and assembly costs [2]. With all these advantages, THF has become a widely used forming process in many industries such as automotive, marine, sanitary and electronic [3]. In the case of the aerospace industry, the application of THF is relatively new and very challenging. Indeed, due to the high strength of the materials used in aerospace and the complexity of the components, very high pressures are needed for hydroforming. On the other hand, the required tight tolerances and the limited formability of most aerospace alloys requires a very good understanding of material flow during the process to avoid premature failure or unacceptable thickness variations during THF.

The purpose of the present publication is on improving the hydroformability of tubes made of SS 321, which is one of the most widely used aerospace alloys. To this end, intermediate heat treatments have been developed and performed between multiple deformation steps. Interrupted tensile tests (ITT) were used as a mean to replicate multistep forming processes [4]. Tensile testing has the advantage of being well implemented in industry, standardized and relatively easy in its procedure and data interpretation. The application of a specific intermediate heat treatment on deformed specimens should lead to material restoration and consequently to mechanical properties improvement [5]. This paper focuses on the comparison of the formability performance and the microstructure to better understand microstructure evolution occurring along with formability enhancement.

2. EXPERIMENTAL PROCEDURE

2.1. Material

The austenitic SS 321 alloy used for this study is a chromium-nickel austenitic grade stabilized by titanium to minimize intergranular chromium carbide precipitation. Due to its composition, this alloy cannot be hardened by heat treatment [6].

Tensile specimens were machined according to ASTM-E8 standard using laser cutting from a 1 mm thick rolled sheet. The samples were loaded in the rolling direction. The latter was selected after prior testing which did not reveal any significant difference in the mechanical properties of specimens cut in the transverse or 45 degrees directions. The multistep forming/heat treatment experiments were carried out for two starting material conditions: as-received (AR) and stress-relieved (SR).

2.2. Mechanical Tests

Tensile tests were conducted at room temperature using a servo-hydraulic Materials Testing System-810 (MTS-810). Prior to testing, the width and thickness of each specimen were measured at three different locations along the 50 mm gauge length to determine the engineering stresses. Tensile tests were performed under displacement control mode with an equivalent strain rate of 0.25 s-1 until rupture.

As illustrated in Fig. 1, the tensile testing machine was equipped with both a laser video extensometer and the Aramis[®] digital image correlation (DIC) system. The laser video extensometer was placed behind the specimen to measure the global extension at the gauge length. The Aramis[®] (DIC) system is a non-contact optical deformation measurement system that is used to measure full 3D-strain fields. The system comprises of two CCD cameras, a trigger box and a high performance PC system. The two stereoscopic cameras of the system were placed in front of the specimen to capture local strains during testing. To measure strains with the DIC system, the surface of the gauge length on each specimen must be covered with a high-contrast

random speckle pattern. Thus, the front side of the specimen was painted with random speckles of black paint on a white background; while on the back side two pieces of laser tape were attached to mark the gauge length for the laser extensometer during testing. The elongation, tracked by the video-extensometer, was used to evaluate the deformation when plotting the global stress-strain curves thereby allowing stopping the test at any specific strain. The deformations captured by the DIC system were used to map the strain distribution over the specimen's gauge length area. The same data acquisition rate, 50 points per second, was used for the laser extensometer and the tensile load records. This allowed an easier and more accurate interpretation of the results. On the other hand, the DIC system triggered images at a frame rate of 3 images per second. At the end of the test, each DIC system image was computed and post-processed to extract the strain maps. The engineering stress was obtained by using the load data collected from the tensile testing machine during the test.



Fig. 1. Schematic set up for interrupted tensile testing.

2.3. Heat Treatments

Stress relieving heat treatments were performed under secondary vacuum at 982°C for one hour followed by controlled cooling at 19°C·min-1 down to 538°C and then air cooled to room temperature. The purpose of the stress relieving heat treatment (SR condition) is to increase the ductility of the material. It also stabilizes the microstructure against chromium carbide formation which may cause sensitization [7].

2.4. Interrupted Thermo-Mechanical Testing

Specimens were subjected to 1, 2 or 3 cycles of tensile testing up to the limit strain with a SR heat treatment performed between each tensile test. For each condition, final tensile loading was systematically performed up to rupture. The value of the limit strain was selected close to the onset of necking, which

was determined via the strain distribution measurement using the DIC system. For each specimen, the thickness was measured at the end of tensile testing. All sequences of tensile tests + SR treatment cycles were performed twice to validate the repeatability of the experiments.

2.5. Microstructure Evolution

The microstructures of the specimens in the AR, SR and cycles of deformation and SR treatments, were characterized by scanning electron microscopy (SEM) and electron backscatter diffraction (EBSD). The specimens were mechanically ground and then polished for 12 hours using a vibratory polisher (Vibro-Met) with colloidal silica. EBSD analyses were performed with a high-resolution Hitachi SU-70 field emission gun - scanning electron microscope (FEG-SEM) equipped with an Oxford-Channel 5 HKL acquisition system. The FEG-SEM was operated at 20 kV with a probe current of 14 nA. The specimens for the EBSD analysis were extracted from the center of the gauge length. Two different types of EBSD maps were acquired. The first one was performed over a large area of $1905 \times 1425 \,\mu\text{m}^2$ with a step size of 1 μm . The second one was focused over $635 \times 475 \,\mu\text{m}^2$ with a step size of $0.25 \,\mu\text{m}$. The larger maps covered a sufficient number of grains for grain size and shape statistical calculations. The second maps were used to estimate the percentage of the different phases present in the microstructure. Both martensite and austenite were indexed. In both cases, the percentage of non-indexed points within the EBSD maps is between 2 % and 8 % depending on the specimens.

3. RESULTS AND DISCUSSION

3.1. Strain Distribution Up To Rupture

The evolution of the major engineering strain distributed along the gauge length median axis during tensile testing of the AR condition is illustrated in Fig. 2. Here, the strain distribution is plotted at various levels of the global strain ε_{global} , i.e., 4 %, 25 %, 40 %, 47 %, 50 % and 51.5 %. Up to 25 % strain, the curve indicates a very uniform strain distribution along the specimen. At 40 % global strain, the distribution of the local strain was still relatively uniform and the difference between the maximum and the minimum strain values was about 5 %. However, beyond 40 % strain, although no local instability was noticed along the gauge length, the above difference exceeded 10 %, indicating that strain localization has taken place. Thus, 40 % strain was considered as the maximum strain that can be applied safely without the risk of strain localization. In the present investigation, in order to maximize the cumulative applied strain for a given number of deformation-heat treatment cycles, the 40 % strain value was selected as the incremental deformation applied to the specimens after each heat treatment.

3.2. Cumulative Stress-Strain Curves

The cumulative engineering stress-strain curves obtained from several interrupted tensile tests are shown in Fig. 3. The initial state of the material before the first tensile test was the AR condition. In Fig. 3, specimen A was pulled up to rupture. Specimen B was pulled up to 40 %, then unloaded and subjected to the SR heat treatment, and finally pulled up to rupture. Specimen C was subjected to 2 cycles of tension: loading up to 40 % — unloading — SR heat treatment two times, and then reloading up to rupture.

The mechanical properties of the specimens at each step of the process are summarized in Table 1. The maximum strain reached at each step as well as the cumulative strain reached at the end of the process is also provided. The cumulative strain was calculated by considering the new dimensions of the specimen after each deformation step as input values. Note that the limit strains applied were not always exactly 40 %: The deviation from the target value of 40 % is less than 1 % and is due to the time difference between the reading of the actual strain by the laser extensioneter and stopping of the test.



Fig. 2. Strain distribution along an AR tensile sample up to rupture.



Fig. 3. Cumulative stress-strain curves for 40 % limit strain starting with AR condition.

Sample	Forming path	0.2 % YS (Mpa)	Strain applied or strain at rupture (%)	Cumulative strain	UTS (MPa)
А	AR	226	51.5	51.5	586
В	AR	229	39.5	39.5	-
	AR + 40 % + SR	181	47.8	87	534
С	AR	223	39.1	39.1	-
	AR + 40 % + SR	183	40	79.1	-
	AR + 40 % + SR + 40 % + SR	176	40	119.1	535

Table 1. Mechanical properties of the sample subjected to 40 % limit strain in the AR condition.

As illustrated in Fig. 3, the application of the SR heat treatments allowed restoring the ductility of the material after each deformation step. Specifically, the elongation to rupture for the AR condition (specimen A), for one (specimen B) and two (specimen C) deformation-heat treatment cycles are similar: 40 to 47 %. It is worth noting that the cumulative strain before rupture in specimen C exceeded 110 % indicating a higher formability of the alloy achieved using the proposed approach. In fact, according to the stress-strain curves, the specimens after each deformation and heat treatment cycle appear to start over as a new material. It can be noted from Table 1 that the conventional 0.2 % yield strength (YS) decreased from one deformation step to the next. In fact, the major drop in YS (from 226 MPa to 182 MPa) occurred after the first deformation and heat treatment cycle. Specifically, this drop was 48 MPa (21 %) and 40 MPa (18 %) for specimens B and C, respectively. However, the decrease in YS was only 4 % after the second deformation-heat treatment cycle; falling from 183 MPa to 176 MPa for specimen C. The significant decrease in YS after the first deformation and is thus a residual effect of the pre-existing work hardening in the material.

To confirm that the YS drop after the first step originates from the mechanical state of the AR specimens, the SR heat treatment was applied to the AR specimens before tensile testing. These specimens were identified as A_{SR} , B_{SR} , and C_{SR} and their testing conditions, cumulative strains and mechanical properties are displayed in Table 2 and Fig. 4. The initial YS of specimens A_{SR} , B_{SR} , and C_{SR} is lower than that of specimens A, B, and C, confirming that the AR specimens were indeed in a work hardened state. As indicated in Fig. 4 and Table 2, the overall decrease in YS for the specimens B_{SR} and C_{SR} , after the first and/or second deformation-heat treatment cycle, does not exceed 15 MPa compared to the initial value (specimen A_{SR}).

Tables 1 and 2 also show that the same level of ultimate tensile strength (UTS) was reached in the last step for all cases. Specifically, the UTS does not significantly vary with either the starting material state or the number of deformation-heat treatment cycles.

Finally, in order to compare the global mechanical behaviour of the material during the multistep forming process, the hardening equations at each step were determined. These equations, which represent the mechanical behaviour of the material, are crucial inputs to the finite element models for the hydroforming process. There are several mathematical expressions generally used to describe the strain hardening behaviour of steels. In the case of SS 321, it has been shown [4,8] that the Swift equation [9], as shown in



Fig. 4. Cumulative stress-strain curves for 40 % limit strain starting with SR condition.

Sample	Forming path	0.2 % YS (Mpa)	Strain applied or strain at rupture (%)	Cumulative strain	UTS (MPa)
A_{SR}	SR	190	51.7	51.7	573
\mathbf{B}_{SR}	SR	190	40.1	40.2	-
	SR + 40 % + SR	176	56.6	96.7	560
C_{SR}	SR	180	40.4	40.4	-
	SR + 40 % + SR	169	40	80.4	-
	SR + 40 % + SR + 40 % + SR	166	39.9	120.3	535

Table 2. Mechanical properties of the sample subjected to 40 % limit strain in the SR condition.

Eq. (1), gives an suitable description of the material behavior for hydroforming applications:

$$\sigma = K(\varepsilon_0 + \varepsilon_P)^n \,, \tag{1}$$

with σ the true stress, K the strength coefficient, n the strain hardening coefficient, ε_P the plastic strain, and ε_0 a constant parameter that represents the strain hardening in the material prior to tensile testing [10]. The appropriate hardening coefficients were obtained using the least square method.

Thus, the data from specimen C_{SR} were used to calculate the strength coefficient K as well as the strain hardening coefficient n for each deformation step (Table 3). The results indicate that the strain hardening

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coefficients are relatively similar throughout the process. Hence, the formability appears not to be affected by the multistep process. It can then be concluded that the intermediate heat treatments did not affect the global hardening behaviour of the SS 321, but only restored its ductility after each cycle of tensile test.

$\frac{\text{Swift}}{\sigma = K(\epsilon_0 + \epsilon_p)^n}$	Specimen C_{SR} SR - 40 % - SR - 40 % - SR - R			
	Step 1	Step 2	Step 3	
K	1367.8	1348.1	1341.5	
n	0.5832	0.5770	0.5489	
ϵ_0	0.0276	0.0252	0.0209	

Table 3. Best fitted coefficients of the Swift hardening equation at each step of the multistep tensile test for C_{SR}.

3.3. Microstructural Analysis

The EBSD scan of the AR specimen is displayed in Fig. 5. The grey levels, observed from one grain to the other in Fig. 5(a), correspond to the EBSD diffraction index quality distribution. The AR specimen exhibits a uniform and equiaxed microstructure with a homogeneous grain size around 8 μ m. A small amount of martensite (about 4.5 %) was identified by EBSD in several grains of the AR specimen (Fig. 5b).



Fig. 5. EBSD maps of a As-received (AR) specimen without deformation: (a) diffraction quality index image reflecting the specimen microstructure; (b) phases identification image: austenite in dark gray and martensite in light gray.

SS 321 as an austenitic stainless steel is susceptible to form strain-induced martensite (SIM) as shown in Fig. 5(b). The martensite phase tends to appear heterogeneously in certain austenitic grains as a result of intense cold working [11,12]. This confirms that the as-received material was not fully annealed and the work hardening observed on tensile test results is due to the cold rolling process of the sheet. The SIM transformation is well documented particularly for metastable austenitic stainless steels with low stacking fault energy such as SS 304 or SS 316 [10,11,13–15]. The formation of SIM is the result of plastic deformation which leads to a phase transformation from austenite into martensite. The transformation is enhanced by high strains, low strain rates, low temperatures and lower nickel contents [16–18].

The relationship between the martensite formation and the formability is known to be a very complex phenomenon. In fact, in the case of uniaxial tension, Angel [17] and then Hecker [18] demonstrated that the ductility appears to depend on both the extent of martensite formation and the strain at which martensite forms. According to Rosen et al. [19] and confirmed by Talyan et al. [20], the ductility in metastable austenitic stainless steels is dominated by the formation and distribution of martensite over the tensile specimen and not by the total amount of martensite. There are two types of martensite formed sequentially from the austenite γ (FCC): martensite $\varepsilon \Box$ (HCP) then \Box martensite α' (BCC) [14,21,22]. The volume percent of ε martensite compared to α' is considered as negligible above approximately 15 % strain [19,23]. In the present study only α' martensitic, i.e., the stable phase, was detected by EBSD which is in agreement with the above references since the applied strain in this study is 40 %.

The EBSD maps of the SR specimens are displayed in Figs. 6–8. In Fig. 6, the SR specimen has undergone no deformation. The SR specimen in Fig. 7 has been deformed to 40 % strain and then SR heat treated again. Finally, the SR specimen presented in Fig. 8 has undergone 2 cycles of deformation and SR heat treatments. Figures 6(a), 7(a) and 8(a) represent the microstructure of the specimens whereas Figs. 6(b), 7(b) and 8(b) represent the distribution of the phases. The specimens that have undergone only the SR treatment (Fig. 6a) exhibit a uniform and equiaxed microstructure with average grain sizes very similar to that of the AR specimen presented in Fig. 5 (7.9 ± 0.5 μ m compared to 8.0 ± 0.2 μ m); however, with the difference that no martensite was found in the microstructure after the SR treatment indicating that the SR heat treatment is not only able to restore the mechanical properties of the material but also reverses the strain induced martensite into austenite without any effect on the grain size.



Fig. 6. EBSD maps of a stress-received (SR) specimen without deformation: (a) diffraction quality index image reflecting the specimen microstructure; (b) phases identification image: austenite in dark gray and martensite in light gray.

The microstructure of the "SR + 40 % + SR" specimen shows a combination of some large grains next to smaller ones (Fig. 7a). This indicates the possibility of recrystallization and/or grain growth due to the second SR heat treatment. In Fig. 8(a), after 2 cycles, i.e. "SR + 40 % + SR + 40 % + SR", the grains are significantly bigger and more heterogeneously distributed than in the previous case (Fig. 7a). Grain size distributions performed over 6670, 2150 and 2271 grains for the SR, "SR + 40 % + SR" and "SR + 40 % + SR + 40 % + SR" specimens, respectively are illustrated in Fig. 9. Such numbers of grains are sufficient for proper statistical analysis since, generally, a minimum set of 200 grains and a step size ten



Fig. 7. EBSD maps of a specimen that has undergone the sequence "SR + 40 % + SR": (a) diffraction quality index image reflecting the specimen microstructure; (b) phases identification image: austenite in dark gray and martensite in light gray.



Fig. 8. EBSD maps of a specimen that has undergone the sequence "SR + 40 % + SR + 40 % + SR": (a) diffraction quality index image reflecting the specimen microstructure; (b) phases identification image: austenite in dark gray and martensite in light gray.

times smaller than the grain diameter is recommended [24]. The grain size distribution confirms that the SR specimen is mostly composed of relatively small grains whereas the introduction of a deformation step ("SR + 40 % + SR") results in a wider distribution in size with a larger number of bigger grains. The analysis of the results in Fig. 9 also indicates that the average size of the grains becomes even bigger when a second deformation step is included. The proportion of small grains decreases when the number of steps is increased. Based on the above discussion, it can be stated that the level of applied prior deformation was high enough to onset some recrystallization and grain growth upon the application of the heat treatment.

The phases present in the specimens are compared in Figs. 6(b), 7(b), and 8(b). Figure 6(b) corresponds to the SR specimen and shows a fully austenitic microstructure. In Fig. 7(b), it can be seen that in the "SR + 40 % + SR" specimen, some martensite is still present in several grains even after the SR treatment.



Fig. 9. Grain sizes distribution measured statistically by EBSD for the three "SR", "SR + 40 % + SR" and "SR + 40 % + SR + 40 % + SR" on a population of 6670, 2150 and 2271 grains respectively.

Up to 7.5 % of martensite was quantified by EBSD showing that the second SR heat treatment was unable to remove all the martensite in the microstructure. However, in Fig. 8(b), the microstructure after 2 cycles of deformation and SR heat treatment is fully austenitic.

In order to confirm the amount of α' -martensite in the specimens identified by EBSD, some measurements were performed on the specimens using a Feritscope[®], a device used to measure the δ -ferrite content on austenitic stainless steels welds. The measurement is based on the ferromagnetism of the α' -martensite compared to that of austenite (paramagnetic). It has been reported that this method can be used as a direct and reliable way to measure the α' -martensite content [22,25]. Since the magnetic permeability of α' -martensite measured by the feritscope depends on the strain, it is necessary to convert the feritscope readings to actual α' -martensite through calibration curves [18,25]. According to Talonen et al. [25], the actual martensite α' content can be evaluated with the following equation:

Martensite
$$\alpha'$$
 content = 1.71 × Feritscope reading. (2)

The results in Table 4 indicate that the α' martensite content measured by EBSD and by the feritscope follow the same trend. The difference can be explained by the non-indexed points of the EBSD reading and the large step size used (0.25 microns) during EBSD mapping. It can also be explained by the fact that the Feritscope[®] estimates the volume of the material and not only a 2D plan as is the case for EBSD.

The Feritscope[®] results confirm that after one cycle of deformation and SR heat treatment, the totality of the SIM cannot be reverted to austenite. This finding is in agreement with those reported by other authors. For instance, Ghosh et al. [23] mention a residual volume percent of strain induced α' -martensite after annealing (800°C/1h) in an austenitic stainless steel with a composition very similar to the one investigated in the present study.

Specimens	Non-indexed	EBSD (%) Austenite	Martensite	Feritscope (%) α'-Martensite
AR (A)	5.9	89.6	4.5	3.83 ± 0.09
$SR(A_{SR})$	1.76	98.16	0.08	0.55 ± 0.01
$SR + 40 \% + SR (B_{SR})$	1.28	91.22	7.5	4.94 ± 0.12
$SR + 40 \% + SR + 40 \% + SR (C_{SR})$	1.03	98.92	0.05	0.61 ± 0.03

Table 4. Phase contents detected in different specimens using 2 methods: EBSD analysis and Feritscope®.

The presence of martensite after the second SR heat treatment in the specimen undergoing the "SR + 40 % + SR" sequence could indicate that 40 % strain corresponds to an excessive deformation level. It would then induce relatively stable martensite that could not be totally reverted into austenite for the specific combination of temperature and time used in the present work. In other words, for a given heat treatment (time and temperature) there may be a maximum level of strain that should not be exceeded otherwise the amount of martensite generated is too high to disappear after the same heat treatment. Interestingly, Grosse et al. [26] have reported that some residual martensite can be found in specimens that have undergone warm drawing and subsequent solution annealing (1h at 1040° C).

However, this threshold effect is not confirmed by the results obtained in the specimen after two cycles ("SR + 40 % + SR + 40 % + SR") as no more residual martensite is found after the last heat treatment. This suggests that the sole application of the SR heat treatment is not sufficient to completely remove residual martensite and other microstructural factors such as grain size and grain to grain misorientations have to be considered. According to Solomon et al. [21], the volume fraction of martensite formed is a function of the grain orientation and its relationship with its neighbors (i.e. local deformation conditions).

4. SUMMARY AND CONCLUSIONS

The objective of this paper was to present a method to increase the formability of the austenitic stainless steel 321 for hydroforming applications. A multistep forming process that consisted of a series of tensile testing with a subsequent softening heat treatment was studied. A limit strain of 40 % and stress relief heat treatments were applied to the tensile specimens. The applied stress relief heat treatment led to a major restoration of the properties, such that after each heat treatment step the yield stress of the material was recovered to its initial value. It was shown that with a judicious combination of forming and heat treatment cycles, very high levels of deformation can be reached through this approach. The importance of the initial state of the material was pointed out. Particularly, the as-received material was not in the fully annealed state and the microstructure contained some residual martensite. This martensite phase, which is most likely strain induced, may be removed by an appropriate heat treatment. However, some remaining martensite was observed after the SR heat treatment in the "SR + 40 % + SR" specimen (but not in the "SR + 40 % + SR + 40 % + SR"). Hence, a deformation threshold of 40 % (not to be exceeded) cannot alone fully explain the stabilization of some strain induced martensite. Additional work is ongoing to better understand the governing mechanisms of martensite stabilization during thermo-mechanical processing. This is of particular importance as the limit strain applied at each step has to be optimized to ensure complete microstructural recovery during multistep forming processes.

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REFERENCES

- 1. Koç, M. and Altan, T., "An overall review of the tube hydroforming (THF) technology", *Journal of Materials Processing Technology*, Vol. 108, No. 3, pp. 384–393, 2001.
- Zhang, S.H., "Developments in hydroforming", *Journal of Materials Processing Technology*, Vol. 91, Nos. 1–3, pp. 236–244, 1999.
- 3. Lang, L.H., Wang, Z. R., Kang, D. C. and Yuan, S. J., "Hydroforming highlights: sheet hydroforming and tube hydroforming", *Journal of Materials Processing Technology*, Vol. 151, Nos. 1–3, pp. 165–177, 2004.
- 4. Anderson, M., *Tube Hydroforming of Aerospace Alloys : Material Characterization Methods*, Master's Thesis, École de technologie supérieure, Montreal, Canada, 2010.
- 5. Golovashchenko, S. and Krause, A., "Improvement of formability of 6xxx aluminum alloys using incremental forming technology", *Journal of Materials Engineering and Performance*, Vol. 14, No. 4, pp. 503–507, 2005.
- 6. Allen, D.K., Metallurgy Theory and Practice, American Technical Society, Chicago, IL, USA, 1969.
- 7. Chandler, H., *Heat Treater's Guide : Practices and Procedures for Irons and Steels*, 2nd ed., ASM International, Materials Park, OH, USA, 1995.
- Saboori, M., Champliaud, H., Gholipour, J., Gakwaya, A., Savoie, J. and Wanjara, P., "Analytical evaluation and finite element simulation of material characteristics in Tube hydroforming of aerospace alloys", *Proceedings* of the International conference on advances in materials and processing technologies (AMPT), Paris, France, pp. 148–151, October 24–27, 2010.
- 9. Swift, H.W., "Plastic instability under plane stress", *Journal of the Mechanics and Physics of Solids*, Vol. 1, No. 1, pp. 1–18, 1952.
- 10. Datsko, J., Material Properties and Manufacturing Processes, Wiley, New York, USA, 1966.
- 11. Lecroisey, F. and Pineau, A., "Martensitic transformations induced by plastic deformation in the Fe-Ni-Cr-C system", *Metallurgical and Materials Transactions B*, Vol. 3, No. 2, pp. 391–400, 1972.
- 12. Olson, G. and Cohen, M., "A general mechanism of martensitic nucleation: Part I. General concepts and the FCC \rightarrow HCP transformation", *Metallurgical and Materials Transactions A*, Vol. 7, No. 12, pp. 1897–1904, 1976.
- 13. Lula, R.A., Parr, J.G., and Hanson, A., *Stainless steel*, American Society for Metals, Metals Park, OH, USA, 1989.
- 14. Mangonon, P. and Thomas, G., "Structure and properties of thermal-mechanically treated 304 stainless steel", *Metallurgical and Materials Transactions B*, Vol. 1, No. 6, pp. 1587–1594, 1970.
- 15. Smaga, M., Walther, F., and Eifler, D., "Deformation-induced martensitic transformation in metastable austenitic steels", *Materials Science and Engineering A*, Vols. 483–484, pp. 394–397, 2008.
- 16. Peterson, S., Mataya, M., and Matlock, D., "The formability of austenitic stainless steels", *JOM Journal of the Minerals, Metals and Materials Society*, Vol. 49, No. 9, pp. 54–58, 1997.
- 17. Angel, T., "Formation of martensite in austenitic stainless steels", *The journal of the Iron and Steel Institute*, Vol. 177, No. 5, pp. 165–174, 1954.
- Hecker, S., Stout, M., Staudhammer, K. and Smith, J., "Effects of Strain State and Strain Rate on Deformation-Induced Transformation in 304 Stainless Steel: Part I, Magnetic Measurements and Mechanical Behavior", *Metallurgical and Materials Transactions A*, Vol. 13, No. 4, pp. 619–626, 1982.
- 19. Rosen, A., Jago, R., and Kjer, T., "Tensile properties of metastable stainless steels", *Journal of Materials Science*, Vol. 7, No. 8, pp. 870–876, 1972.
- 20. Talyan, V., Wagoner, R., and Lee, J., "Formability of stainless steel", *Metallurgical and Materials Transactions A*, Vol. 29, No. 8, pp. 2161–2172, 1998.

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- 21. Solomon, N. and Solomon, I., "Deformation induced martensite in AISI 316 stainless steel", *Revista de Metalurgia*, Vol. 46, No. 2, pp. 121–128, 2010.
- 22. Talonen, J., Nenonen, P., Pape, G. and Hanninen, Н., "Effect of strain rate on the strain-induced γ α' -martensite transformation and mechanical properties of \rightarrow austenitic stainless steels", Metallurgical and Materials Transactions A, Vol. 36, 2, No. pp. 421-432, 2005.
- Ghosh, S., Mallick, P., and Chattopadhyay, P., "Effect of reversion of strain induced martensite on microstructure and mechanical properties in an austenitic stainless steel", *Journal of Materials Science*, Vol. 46, No. 10, pp. 3480–3487, 2011.
- 24. Humphreys, F.J., "Nucleation in recrystallization", *Proceedings of the Second Joint International Conferences on Recrystallization and Grain Growth, ReX and GG2, Société Française de Métallurgie et de Matériaux*, Annecy, France, Trans Tech Publications Ltd, pp. 107–116, August 30-September 3, 2004.
- 25. Kumar, B., Mahato, B., Bandyopadhyay, N. and Bhattacharya, D., "Influence of strain-induced phase transformation on the surface crystallographic texture in cold-rolled-and-aged austenitic stainless steel", *Metallurgical and Materials Transactions A*, Vol. 36, No. 11, pp. 3165–3174, 2005.
- Grosse, M., Kalkhof, D., Niffenegger, M. and Keller, L., "Influencing parameters on martensite transformation during low cycle fatigue for steel AISI 321", *Materials Science and Engineering A*, Vol. 437, No. 1, pp. 109–113, 2006.