Publication IV

M. Ivanchenko, Y. Yagodzinskyy, and H. Hänninen. 2009. Effect of plastic deformation on anelastic mechanical losses in multicomponent substitutional austenitic alloys. Materials Science and Engineering A, volumes 521-522, pages 121-123.

Materials Science and Engineering A 521–522 (2009) 121–123

Contents lists available at ScienceDirect

Materials Science and Engineering A

journal homepage: www.elsevier.com/locate/msea

Effect of plastic deformation on anelastic mechanical losses in multicomponent substitutional austenitic alloys

M. Ivanchenko*, Y. Yagodzinskyy, H. Hänninen

Laboratory of Engineering Materials, Helsinki University of Technology, P.O. Box 4200, FI-02015 TKK, Finland

ARTICLE INFO

Article history:
Received 18 July 2008
Accepted 22 September 2008

Keywords:
Austenitic stainless steel
Internal friction
Hasiguti peaks
Dislocation–point defect interaction
Modulus recovery

ABSTRACT

Complex two-component peak of anelastic mechanical losses was observed in multicomponent substitutional austenitic alloys after prior plastic deformation. In AISI 316L, AISI 316LN, AISI 310 and AISI 304 austenitic stainless steels, this peak is situated in the wide range of temperature from 200 to 400 K. Characteristics of the observed peak for different thermo-mechanical treatments were studied employing low-carbon AISI 316L austenitic stainless steel. It is shown that the low-temperature component (peak P1 with maximum at 250 K) has a relaxation origin, while the high-temperature component (peak P2 located at 340–360 K) represents a transitional process developing in a deformed austenitic stainless steel during linear heating. The activation enthalpy and the pre-exponential factor of the relaxation time of peak P1 are 0.48 ± 0.06 eV and $10^{-11\pm1}$ s, respectively. Effects of prior cold deformation and heating rate on the observed anelastic mechanical losses are also described. The origins of the peaks are discussed in terms of interactions of dislocations with point defects generated by plastic deformation.

1. Introduction

Austenitic stainless steels (SS) are attractive materials for the applications where good corrosion properties are requested together with high toughness and good formability. Thus, it is important to know the change in the mechanical properties of the austenitic stainless steels after prior plastic deformation. The knowledge of mobility of dislocations and their interactions with lattice point defects is required for understanding the mechanisms of creep as well as of static and dynamic strain aging. Necessary information about the state of dislocations, interstitials and lattice defects and their possible interactions can be obtained by means of mechanical loss spectrometry better known as internal friction (IF). The changes in the relaxation spectra of austenitic stainless steels with the amount of plastic deformation and the presence of interstitials in the solid solution have been studied by many authors; for a brief review see Ref. [1]. Most of the studies have been focused on the IF peak located in the vicinity of 630 K (at frequency ≈1 Hz) established by Rozin and Finkelshtain in 1953 [2]. Only few works are available on the IF peaks located in the temperature range from 100 to 500 K, which are observed in the deformed austenitic stainless steels [3–8]. In metastable austenitic SS the IF peaks located in this temperature range may be attributed to strain-induced α' and ε martensites [5–7]. However, IF peaks in the low-temperature range have also been ascribed to deformed stable austenite, where their origin may not be explained by the presence of other phases [4,7,8]. In spite of the systematic observation of the internal friction peaks in different austenitic multicomponent alloys in the temperature range from 100 to 500 K at frequencies around 1 Hz, there is still no clear explanation of their nature. The aim of the present study is to verify the presence of the IF peaks in different grades of austenitic stainless steels and to characterize these peaks in cold-deformed stable low-carbon AISI 316L austenitic stainless steel.

2. Experimental

Four types of austenitic stainless steels AISI 316L, AISI 316LN, AISI 310 and AISI 304 were used in this study. Chemical compositions of the studied steels are shown in Table 1. All the studied materials were supplied as plates in annealed condition. The specimens for prior straining were cut from the plates by electric discharge machining as samples of width 6–8 mm and thickness 0.5–2 mm. The 5 or 20% tensile strain of the samples was performed at room temperature using a servohydraulic 100 kN MTS 810 test machine at strain rate of 10^{-3} s^{-1}. The level of strain was controlled by an MTS 632.12C – 20 extensometer. The temperature dependencies of the internal friction, Q^{-1}, were measured using the method of free decay of resonance oscillations by an inverted torsion pendulum over the temperature range from 100 to 550 K with the amplitude of the torsion deformation of about 10^{-5}. The natural frequency of the pendulum was varied in the range from 1 to 3 Hz. In most of the measurements the applied heating rate (HR) was 2 K/min. In order

* Corresponding author. Tel.: +358 9 451 3539; fax: +358 9 451 3537.
E-mail address: mykola.ivanchenko@tkk.fi (M. Ivanchenko).

0921-5093/$ – see front matter © 2009 Elsevier B.V. All rights reserved.
doi:10.1016/j.msea.2008.09.093

Table 1
Chemical composition (in wt.%) of the studied steels.

Grade	Cr	Ni	Mo	Mn	C	N
AISI 316L	16.4	10.5	2.0	0.98	0.02	–
AISI 316LN	17.0	11.0	2.1	1.5	0.02	0.176
AISI 310	25.54	19.09	0.24	0.88	0.049	0.036
AISI 304	18.5	10.9	0.22	0.73	0.048	–

to examine whether the studied peaks had relaxation or transitional origin, different HRs in the range of 1–4 K/min were applied. The specimens for internal friction measurements with typical size of about 0.5 mm × 1.5 mm × 35 mm were cut by a tabletop disc-saw and polished with 1200 grit emery paper.

3. Results

The temperature dependence of the anelastic mechanical losses, Q^{-1} (i.e., internal friction), of four different grades of austenitic SS (AISI 316L, AISI 316LN, AISI 304 and AISI 310) are shown in Fig. 1. IF spectra were obtained after prior tensile strain of the steels to 20% of elongation. Observed spectra have two well-defined maxima for the strained materials, forming a complex two-component peak with low- and high-temperature components at about 250 and 350 K called here peak P1 and P2, respectively.

Fig. 2 shows the internal friction of AISI 316L SS for various values of prior tensile strain and for the as-supplied materials. The amplitude of the observed two-component IF peak increases in

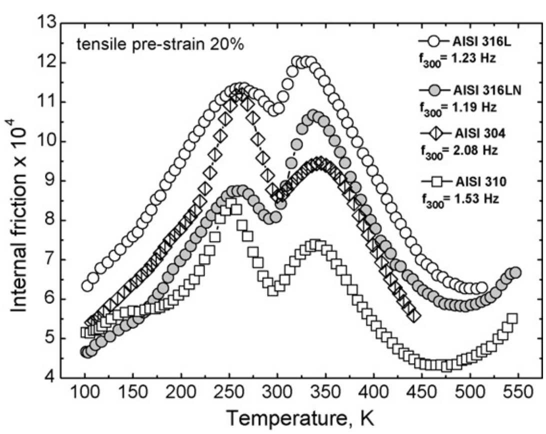

Fig. 1. Temperature dependence of internal friction for different grades of austenitic stainless steels after prior plastic deformation in tension to 20% strain at room temperature and at a strain rate of 10^{-3} s^{-1}.

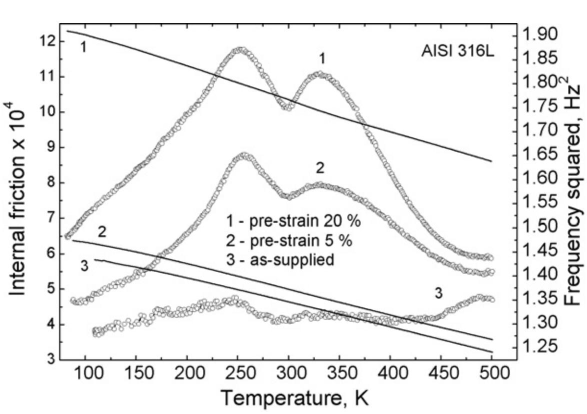

Fig. 2. Change of the peaks in AISI 316L austenitic stainless steel plastically pre-strained in tension at room temperature and at a strain rate of 10^{-3} s^{-1}.

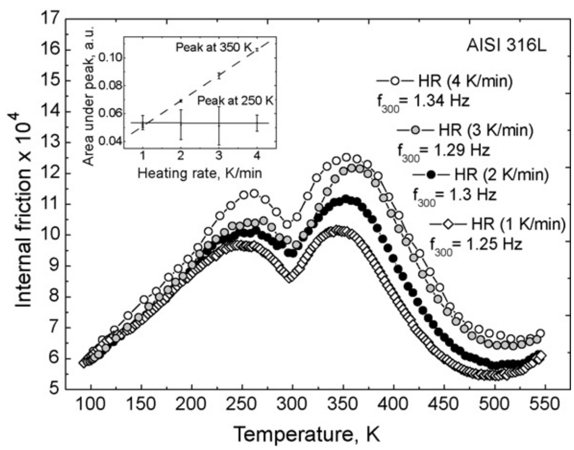

Fig. 3. Temperature dependence of internal friction of AISI 316L steel at different heating rates. The inset shows the dependence of the area under fitted peaks on heating rate for the low- (dashed line) and the high-temperature peak (solid line).

magnitude with increasing the amount of prior tensile strain and only minor humps were evident for the as-supplied material.

In order to study whether the observed peaks had relaxation or transitional origin, different heating rates in the range of 1–4 K/min were applied. The obtained internal friction spectra and dependencies of the areas under the peak on the value of HR are shown in Fig. 3. To calculate the areas under the peak of IF spectra, the background was assumed to be a straight line from 100 to 500 K and was subtracted. The spectra were then fitted with Gaussian peaks with correlation coefficient $R \geq 0.995$. The low-temperature component (P1) was fitted with two Gaussian peaks with centres of gravity in the vicinity of 190 and 250 K and the sums of the areas of fitted peaks were taken. The high-temperature part (P2) was fitted with single Gaussian peak with maximum around 355 K. The obtained results show that P1 has a relaxation origin, while P2 represents a transitional process taking place during linear heating.

Samples of 20% pre-strained AISI 316L austenitic SS were measured several times in the temperature range from 100 to 550 K without dismounting the sample to evaluate the reproducibility of the observed two-component IF peak. Fig. 4 shows the changes in IF spectrum and frequency squared, which is proportional to the shear modulus, during the first and second measurement runs. A small defect in the shear modulus was observed during the first run over the temperature range of P2. In the second run, the high-temperature component (P2) disappears almost completely, while

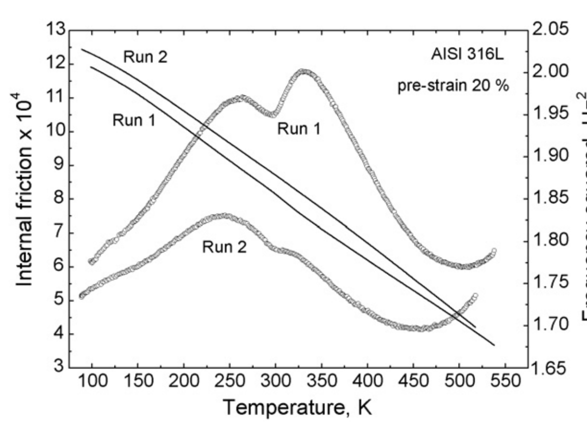

Fig. 4. Temperature dependence of internal friction and squared natural frequency for AISI 316L austenitic stainless steel after tensile pre-strain of 20% at room temperature and at a strain rate of 10^{-3} s^{-1}. Comparison between data of the first and second measurement runs.

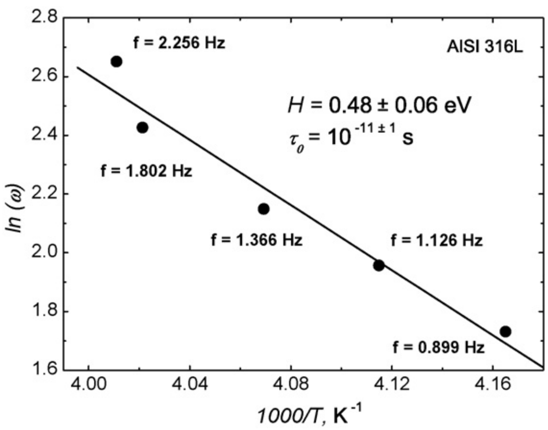

Fig. 5. Arrhenius plot for the low-temperature peak P1, obtained after removal of transitional high-temperature peak P2 by annealing at 550 K for 100 min.

the low-temperature component (P1) decreases and its peak maximum shifts towards lower temperature. The values of the shear modulus in the second run were higher than in the first. In subsequent additional runs the IF spectra and the values of frequency squared remained the same as in the second run.

According to the results concerning the HR and the stability of the peaks, the low-temperature component of the double peak (P1) appears to be a relaxation process. To obtain the activation enthalpy and the pre-exponential factor for the process, P2 was removed by annealing the material at 550 K for 100 min and the measurements at different frequencies were carried out by changing the inertia members of the pendulum. The Arrhenius plot for peak P1 is shown in Fig. 5. The activation enthalpy and the pre-exponential factor of the relaxation time are 0.48 ± 0.06 eV and $10^{-11\pm1}$ s, respectively.

4. Discussion

The observation of an IF peak with two well-defined components in all the studied grades of steels provides the evidence of a common behavior of deformed austenitic stainless steels. The studied peaks are caused by plastic deformation, as they are not observed in the as-supplied annealed condition of the materials and increase in amplitude with the amount of plastic deformation.

From the obtained results it is evident that only P1 has a relaxation origin. Most probably, peak P1 is similar to Hasiguti peaks [9]. This peak is stable and is present in all the repeated runs with the same value of amplitude after P2 is annealed out. The activation enthalpy and the pre-exponential factor of the relaxation time (0.48 ± 0.06 eV and $10^{-11\pm1}$ s, respectively) are in good agreement with previous results obtained by Igata et al. for cold-worked AISI 304L austenitic stainless steel [3,7]. The position of the peak observed by Igata et al. [3] calculated for an 1 Hz frequency would be 265 K, which is close to the position of P1 reported in this study. The relaxation processes described by Igata et al. [3,7] and Hasiguti et al. [9] were associated with the interactions of dislocations with point defects. The model suggested by Hasiguti et al. is based on stress-assisted unpinning of dislocations from point defects. The relaxation maximum appears, when the frequency of thermal unpinning coincides with the external vibrational frequency. As the temperature position, activation energy and pre-exponential factor of the relaxation time of P1 are very close to those for the peaks described by Igata et al. and Hasiguti et al., we assume that the same mechanism is involved in the present relaxation processes.

The disappearance of peak P2 accompanied by the recovery of the shear modulus during the first measurement run of the prestrained material is evidence of some transitional process taking place within the material during heating. This process may be related to the annihilation of the point defects. As a matter of fact, measurements of resistivity changes during isochronal annealing in the temperature range from 373 to 533 K exhibit a recovery stage in irradiated AISI 304 steel, attributed to annihilation of point defects [10]. A recovery stage, which does not involve a rearrangement of dislocations, was found for cold-worked AISI 304 and AISI 316 austenitic SS at around 580 K [11]. The activation energy for this process was 57 kJ mol^{-1} (0.59 eV) for AISI 304 and about 80 kJ mol^{-1} (0.83 eV) for AISI 316 austenitic SS. These obtained values are close to the enthalpy of vacancy diffusion in FCC metals [12]. Based on these facts, we suggest that peak P2 is attributable to the migration of point defects to the dislocation cores. At temperatures around 500 K, when the mobility of interstitial impurities increases, the pinning of dislocations by carbon or nitrogen atoms may also occur. The pinning of dislocations by impurities should lower their mobility and may result in a decrease in the amplitude of P1 as observed in the second run. This hypothesis is supported by the increase in the shear modulus taking place in the repeated measurement runs, when P2 is not present. The kinetics of the P2 decay and the recovery of the shear modulus during annealing will be discussed elsewhere.

5. Conclusions

IF peak with two well-defined components, which depend on the amount of plastic deformation, is observed in all the studied grades of austenitic stainless steels. The low-temperature component of the observed peak has relaxation origin with characteristics similar to Hasiguti peaks. Thus, it appears to represent a dislocation–point defect interaction effect. The high-temperature component of the peak has transitional origin and is probably related to the annihilation of point defects introduced by the plastic deformation at the dislocation cores.

Acknowledgements

This study has been supported by SAFIR 2010 (Finnish Research Programme on Safety of Nuclear Power Plants) and Graduate School on New Materials and Processes.

References

[1] H. Numakura, K. Kashiwazaki, H. Yokoyama, M. Koiwa, J. Alloys Compd. 310 (2000) 344–350.
[2] K.M. Rozin, B.N. Finkelshtain, Rep. Acad. Sci. USSR 91 (1953) 811–814.
[3] N. Igata, H.B. Chen, K. Miyahara, T. Uba, J. Phys. Col. C5, Tome 42 (1981) 193–198.
[4] M. Ivanchenko, U. Ehrnstén, V. Nevdacha, Y. Yagodzinskyy, H. Hänninen, Proceedings of 7th International Conference on High Nitrogen Steels, Ostend, Belgium, September 19–22, 2004, GRIPS media GmbH, Bad Harzburg, Germany, 2004, pp. 641–649.
[5] V.R. Baraz, S.V. Grachev, L.D. Rolschikov, Steel USSR 2 (1972) 670–672.
[6] J. Talonen, H. Hänninen, Metall. Mater. Trans. A 35A (2004) 2401–2406.
[7] N. Igata, H.B. Chen, K. Miyahara, Scripta Metall. 16 (1982) 1039–1042.
[8] A.H. Heuer, F. Ernst, H. Kahn, A. Avishai, G.M. Michal, D.J. Pitchure, R.E. Ricker, Scripta Mater. 56 (2007) 1067–1070.
[9] R.R. Hasiguti, N. Igata, G. Kamoshita, Acta Metall. 10 (1962) 442–447.
[10] M.S. Anand, B.M. Pande, R.P. Agarwala, J. Nucl. Mater. 58 (1975) 117–118.
[11] S.W. Yang, J.E. Spruiell, J. Mater. Sci. 17 (1982) 677–690.
[12] J.A. Slane, C. Wolverton, R. Gibala, Metall. Mater. Trans. A 35A (2004) 2239–2245.