Texture evolution of warm-rolled and annealed 304L and 316L austenitic stainless steels

D Lindell
1 Researcher, Swerea KIMAB AB, Box 7047, 165 07 Kista, SWEDEN
E-mail: david.lindell@swerea.se

Abstract. The brass-to-copper rolling texture transition is observed during warm rolling austenitic stainless steels. In the current paper austenitic stainless steels 304L and 316L have been subjected to warm rolling at 700°C to 90 % reduction. The evolution of microstructure and texture during subsequent annealing has been studied using dilatometry and electron backscatter diffraction. Recrystallisation texture for 304L was primarily cube with some retained rolling texture while 316L only had retained rolling texture. The different behaviour between the two steels is believed to originate from differences in molybdenum content.

1. Introduction
Stainless steels have attractive combinations of mechanical and chemical properties that can be tailored to fit many different type of applications. Highly fluctuating cost of alloying elements has increased the interest for ferritic grades and lean duplex grades. Lean duplex grades are usually benchmarked against austenitic grades where they have an advantage when it comes to specific weight. This trend towards a more diversified stainless steel portfolio has reinforced the motivation to understand the delicate interplay between microstructure and various types of corrosion. This in turn can lead to new grades that are optimised not only in terms of chemistry but also by the microstructure. One example is certain high temperature grades for which the grain size is a balance requirement in terms of creep strength and corrosion resistance. A more futuristic view includes the concept of grain boundary engineering. Steels with over 80 % coincidence lattice boundaries (CSL) and a consequent impressive resistance against intergranular corrosion have been demonstrated [1, 2]. Further, crystallographic effects on localised corrosion [3, 4] and corrosion in sulphuric acid [5] have been observed. The aim of the current paper was to use a processing rather than an application point of view to explore the possibilities of texture and microstructure control of warm rolled material. Stainless steels have low-to-medium stacking fault energy so the rolling texture is usually of the brass type [6-8]. However, Hu [8, 9] has demonstrated that AISI 304 undergoes the brass-to-copper texture transition during warm rolling. The warm rolling approach is appealing because recrystallisation textures of materials with copper-type rolling texture can be very sensitive to processing history and chemistry.

2. Experimental
Two 26 mm plates of 304L and 316L were received from Outokumpu Stainless in Degerfors, Sweden. The chemical composition of the steels is given in Table 1. In order to refine the grain size prior to warm rolling, the 26 mm thick plates were cold rolled to 50 % reduction and annealed at 850°C for 1 h. The now 12 mm thick plates were cut to pieces of 20 x 8 x 12 mm³ thereafter reheated to 600, 700 and 800°C for 15 min and immediately warm rolled in six passes down to 90 % nominal reduction. The surface temperature after these six passes was about 50-80°C lower than after reheating. For 304L, the copper rolling texture was pronounced in all three strips. However the traces of brass texture were largest for 600 ºC and some degree of recrystallisation was observed for the 800 ºC specimen so the 700 º C specimen was selected for further studies. For comparison, the 700 ºC version of 316L was also selected. The actual rolling reductions of these two strips were 91 and 88 % for 304L and 316L respectively. The evolution of microstructure and texture during annealing was studied using dilatometry and EBSD. The annealing temperatures were 800, 900 and 1140°C and the...
annealing times (time at annealing temperature) were 3, 10, 30 and 120 s. The heating and cooling rate was in all cases 50 °C s⁻¹. EBSD maps comprising at least 30k points with coarse step size were collected for texture analysis and higher magnification maps were collected for grain size determination. Because no standard lubrication could be used during rolling the surfaces consists of a sheared zone, which was excluded from further analysis. Orientation distribution functions were calculated using the serial expansion method using the software Salsa within the HKL software package using $L_{\text{max}} = 22$ and 3° Gaussian spread. Grain sizes extracted from the EBSD were determined using the linear intercept method along the rolling direction with a critical disorientation of 10°.

Table 1. Heat analysis of the steels investigated (wt%).

|       | Cr | Ni | Mn | Mo | Cu | Si | C  | N  |
|-------|----|----|----|----|----|----|----|----|
| 304L  | 18.3 | 8.1 | 1.5 | 0.5 | 0.4 | 0.3 | 0.021 | 0.07 |
| 316L  | 17.5 | 11.1 | 1.2 | 2.3 | 0.5 | 0.4 | 0.014 | 0.07 |

3. Results and discussion

3.1 Evolution of microstructure

The recrystallisation fraction and average grain size as a function of annealing cycle is given in Table 2. 304L recrystallises at 800 °C, even though not fully at the longest time investigated, while 316L does not recrystallise at all within this time frame. Both steels recrystallise at 900 °C with a grain size of ~4 µm when fully recrystallised. It is often stated that the relative high recrystallisation temperature of the stainless steels make it impossible to achieve grain sizes below about 10 µm [10, 11] without special treatments such as martensite reversion. Alloying for this kind of purpose is a balancing act between the effect of alloying on recrystallisation temperature and on grain boundary mobility. The chemistries used here are by no means optimised in this respect, however it demonstrates that at least half of the typically stated 10 µm can be achieved in practise with simple means. The effect of alloying on grain boundary mobility is most clearly seen at the highest annealing temperature, 1140°C. The Mo alloyed 316L had a grain size of 20 µm after the annealing for 120 s at 1140°C while 304L had a grain size of 54 µm after the same treatment.

Table 2. Average grain size and recrystallisation fraction as a function of annealing cycle.

| Annealing temperature (°C) | Annealing time (s) | Degree of recrystallisation (%) | Average grain size (µm)** | Degree of recrystallisation (%) | Average grain size (µm)** |
|---------------------------|-------------------|---------------------------------|---------------------------|---------------------------------|---------------------------|
|                           |                   | $\Sigma_3$ included | $\Sigma_3$ excluded | $\Sigma_3$ included | $\Sigma_3$ excluded |
| 800                       | 3                 | 0 -                | -                      | 0 -                | -                      |
| 800                       | 10                | 5 1                | 1                      | 0 -                | -                      |
| 800                       | 30                | 18 1              | 1                      | 0 -                | -                      |
| 800                       | 120               | 65 2              | 3                      | 0 -                | -                      |
| 900                       | 3                 | 87 2              | 3                      | 5 1                | 2                      |
| 900                       | 10                | 100 3             | 5                      | 55 2               | 3                      |
| 900                       | 30                | 100 5             | 10                     | 81 2               | 3                      |
| 900                       | 120               | 100 7             | 12                     | 100 3              | 4                      |
| 1140                      | 3                 | 100 7             | 12                     | 100 5              | 8                      |
| 1140                      | 10                | 100 9             | 16                     | 100 6              | 11                     |
| 1140                      | 30                | 100 13            | 22                     | 100 8              | 13                     |
| 1140                      | 120               | 100 28            | 54                     | 100 11             | 20                     |

*Assumed that the deviation from 99 % hit rate originate from deformed material only.

**Critical disorientation = 10°.
3.2 Evolution of recrystallisation texture

ODF sections for 304L with constant $\phi_2 = 0^\circ$ and $\phi_2 = 45^\circ$ are given in Figure 1. The as-rolled texture (a) shows typical copper rolling texture features with a pronounced fraction of the copper orientation $\{112\}<111>^C$, along with S $\{123\}<634>^S$ (not shown in these sections) and brass $\{110\}<112>^B$, which shows some spread towards the Goss orientation $\{110\}<001>^G$. At the early stage of recrystallisation (b) the copper and brass components are weakened while the Goss orientation and all other orientations along $\phi_1 = \phi_2 = 0^\circ$ are strengthened. Towards the end of recrystallisation and during the early grain growth stage (c-e), two maxima along $\phi_1 = \phi_2 = 0^\circ$ start to form at $\Phi = 0^\circ$ (cube orientation) and $\Phi \approx 30^\circ$.

Further grain growth (f-g) results in strengthening of the cube orientation, but there are still remains of the rolling texture even after excessive grain growth. Here the evolution of texture is described on basis of the grain size, however there are also indications that a higher temperature for a given final grain size facilitates cube texture formation. 900$^\circ$C-120s and 1140$^\circ$C-3s resulted in the same average grain size but the latter had a sharper cube component, see Figure 1e-f.

A comparison between the recrystallisation texture of 304L and 316L is given in Figure 2, where the ODFs are plotted for $\phi_1 = \phi_2 = 0^\circ$. 316L had, compared to 304L, a stronger Goss component in the deformation texture. During grain growth in 316L, the Goss component weakens and a maximum at $\Phi \approx 30^\circ$ is formed. In contrast to 304L further grain growth does not result in the formation of cube. The difference between the two steels, and the fact that rolling texture was retained in 304L, is believed to be due to differences in grain boundary mobility by solute drag effects, mainly from molybdenum. Even the 304L used in this work had 0.5 wt% Mo, which may explain why components in the rolling texture are retained. The work by Hu [8, 9] shows that this can be avoided, unfortunately no information regarding the Mo content in their work could be found. Another explanation that cube was not formed is that insufficient number of nuclei is retained from earlier processing steps. This explanation is however not likely because the number of cube oriented grains was very similar for 304L and 316L.

---

a) As warm rolled at 700$^\circ$C to 91% reduction.

b) 800$^\circ$C ; 30 s – 18 % recrystallised, GS = 1 $\mu$m. The ODF is calculated from the recrystallised grains i.e. not the full dataset.

c) 800$^\circ$C ; 120 s – 65 % recrystallised, GS = 3 $\mu$m. The ODF is calculated from the recrystallised grains i.e. not the full dataset.

d) 900$^\circ$C ; 10 s – fully recrystallised, GS = 5 $\mu$m.
e) 900ºC ; 120 s – fully recrystallised, GS = 12 µm.

f) 1140ºC ; 3 s – fully recrystallised, GS = 12 µm.

g) 1140ºC ; 30 s – fully recrystallised, GS = 22 µm.

Figure 1. Examples of ODF sections for $\phi_2 = 0^\circ$ (left) and $45^\circ$ (right) for 304L. For the partially recrystallised material given in a and b the orientation functions are calculated on basis of the recrystallised fraction only. The contour steps are all unity.

Figure 2. Plot of the ODF for constant $\phi_1 = \phi_2 = 0$.

5. Conclusions

Microstructure and texture evolution of 304L and 316L during annealing after warm rolling have been studied. Recrystallisation proceeds with a strengthening of the Goss component that spreads along the $\phi_1 = \phi_2 = 0^\circ$ section ($<100> // RD$). During growth, 304L develops cube texture but with some rolling texture retained, while for 316L the rolling texture is retained without any formation of cube texture. To increase the strength of the cube texture the chemistry has to be further optimised, the current belief is that molybdenum should be avoided in order to do so.

Acknowledgments

The Hugo Carlsson foundation is greatly acknowledged for the financial support of this work. Prof. Rachel Pettersson and Prof. Bevis Hutchison are greatly acknowledged for reading the script and coming up with ideas.

References

[1] Michiuchi M, Kokawa H, Wang Z J, Sato K and Sakai K 2006 Acta Materialia 54 p 5179
[2] Shimada M, Kokawa H, Wang Z J, Sato Y S and Karibe I 2002 Acta Materialia 50 p 2331
[3] Shahyari A, Szpunar J and Omanovic S 2009 Corrosion Science 51 p 677
[4] Krishnan S, Dumre J, Bhatt S, Esther T, Akinlabi T and Ramalingam R 2013 International Journal of Mechanical, Industrial Science and Engineering 7 p 246
[5] Lindell D and Pettersson R 2014 Submitted to Materials and Corrosion
[6] Chowdhury S G, Das S and De P K 2005 Acta Materialia 53 p 3951
[7] Kumar B R, Mahato B, Bandopadhyay N R and Bhattacharya D K 2005 Materials Science and Engineering A 394 p 296
[8] Hu H 1974 Texture 1 p 233
[9] Hu H and Goodman S R 1984 Material Science and Engineering 67 p 143
[10] Schino A D and Kenny J M 2003 Materials Letters 57 p 3182
[11] Tomimura K, Takaki S, Tamimoto S and Takanaga Y 1991 ISIJ International 31 p 721