1. Introduction

Among the various softening processes recrystallization plays a major role due to its effect on the characteristic microstructural length scale which affects macroscopic properties like strength e.g. through the Hall–Petch relation. Therefore, the development of high nitrogen duplex steels appears very attractive, because in addition to specific improvements resulting from the coupling between ferrite and austenite and from nitrogen enriched austenite, the phase boundaries strongly confine grain growth. In a previous paper the hot and cold deformation of duplex stainless steels (DSS) was addressed. The aim of this study was to elucidate how recrystallization takes place during annealing of both a salt bath furnace and a conventional furnace. During annealing the rolled microstructure changes to a more isotropic morphology without pronounced coarsening of the constituent phases. Depending on the initial cold rolling texture the ferrite phase undergoes recovery or recrystallization, but recovery dominates the softening of the bcc phase. Despite of the layered microstructure the active softening mechanisms of ferrite are the same as known from bulk single phase materials and obviously are negligibly influenced by the presence of the austenite phase.

KEY WORDS: duplex stainless steel; recrystallization; recovery; texture; microstructure.

2. Experimental

The composition of the five investigated steel grades is given in Table 1. These duplex stainless steel grades were commercial alloys except for their different thermomechanical histories. DSS 1 was an extruded and round forged rod material 30 mm. DSS 2 and DSS 3 were hot rolled rod materials with rectangular profiles 45×12 mm² which were produced by alternating the rolling plane (horizontal and vertical) during the hot rolling schedule. DSS 4 was a hot rolled plate of 6 mm thickness, and DSS 5 served as a test alloy and was only available as a slab of an ingot. Hence, the latter material was only cold rolled, but in the same manner as the other steel grades.

Subsequent to hot and cold rolling the samples were annealed either in a salt bath furnace at 1 100°C or in a conventional furnace at 1 050°C and subsequently water quenched. While the high heating rate of the salt bath annealing allowed softening to proceed without being influenced by precipitates, the conventionally annealed samples experienced some phase transformations and precipitates of α-phase, χ-phase and CrN nitrides because the temperature versus time profile T(t) crossed TTT curves. The annealing temperatures of 1 050°C and 1 100°C were chosen to enable long annealing times without the influence of pre-
Precipitates by homogenization because in this temperature range the volume fraction of the α- and γ-phase stayed almost constant.

The microstructure development was investigated by optical microscopy and TEM. All textures were determined from the normal direction (ND) with Co-Kα1 radiation (λ=1.79 Å) by means of incomplete pole figures (5°/H11015 dα{111} dγ{110}) for the γ-phase and two [(200)_α, (112)_α, (113)_α] for the α-phase in back reflection mode on a fully automated texture goniometer. By use of a position sensitive detector (PSD) it was possible to record and deconvolute the {111}_γ and {110}_α pole figures despite of the peak overlap owing to the close interplanar spacing (d_{111}^γ=d_{110}^α).

### 3. Results

In Figs. 1 through 4 the microstructure evolution during annealing in a salt bath furnace at 1100°C after 60% cold rolling is presented. The 60% deformed sample was chosen for a demonstration of the microstructure development for practical reasons, meanwhile the microstructure development after 90% rolling was revealed to be the same. Figure 1 shows the state of incipient annealing where locally
grains appear in shear bands and slip bands but the cold rolled morphology with the typically elongated, alternating layered structure of austenite and ferrite is conserved. After 10 s annealing primary recrystallization had progressed in the austenite phase (light areas) and left a fine grain structure with only some unrecrystallized remains of deformed austenite (Fig. 2). The grains of the ferritic phase (dark areas) already extended through the full thickness of the layer. Hence, in the ferrite phase primary softening and subsequent grain growth seemed to have already occurred. During the following 100 s anneal grain growth progressed also in the austenite phase and a similar grain arrangement as in the ferrite phase was reached (Fig. 3). The result was a kind of “bamboo-structure” for each phase where the grain boundaries ran perpendicular to the phase boundaries, and the grains extended through the full height of the respective layer. However, the alternate austenite–ferrite layer arrangement formed during cold rolling remained unchanged. Annealing of 2000 s caused a change in the microstructure both with regard to phase arrangement and morphology (Fig. 4). The bamboo-like structure of both phases turned to a kind of “pearl structure”, i.e. a more globular grain structure and phase morphology. This conversion continued and became more pronounced at longer annealing times, as sketched in Fig. 5. The microstructure showed only little tendency to coarsen during annealing despite the high annealing temperature.

Annealing in a conventional furnace required a longer time to reach such morphology due to the lower temperature (1050°C) and owing to the retarding effect of precipitates on recrystallization and grain growth. Because of the lower heating rate and thus the longer time intervals spent in the precipitation sensitive states the steels DSS 2 and 4 showed strong precipitation and phase transformation for annealing times of 200 to 500 s. Invariably the softening in the ferrite phase progressed much faster than in the austenite phase (Fig. 6).

As discussed in Ref. 1) the cold rolling textures of the austenite phase in the different investigated duplex steels were very similar. Correspondingly, the annealing textures did not differ substantially from each other. Surprisingly, the annealing textures (Figs. 6(a), 6(b)) did not show the typical brass recrystallization (BR) orientation \{236\}-(385) well known for rolled and recrystallized single phase austenite with comparable stacking fault energy (Fig. 7(c)). Instead remnants of the rolling texture, especially the Goss orientation \{011\}-(100) were retained and additionally the D orientation \{11 11 8\}-(4 4 11), a twin to the Goss orientation, developed. The typical recrystallization texture of a comparable single phase material with similar stacking fault energy is given in Fig. 7(c). It is obvious that in spite of similar cold rolling textures of the austenite phase of the DSS and comparable single phase materials\(^1\), the annealing

![Fig. 5. Sketch of the microstructure development during salt bath (top row) and conventional (bottom row) annealing (a) austenite, (g) austenite.](image)

![Fig. 6. TEM micrograph of a specimen annealed 40 s in a conventional furnace. The ferrite (a) is already fully recovered while the austenite (γ) still has a deformed microstructure.](image)

![Fig. 7. Annealing textures of the austenite phase of 90% cold rolled DSS 2 (a) and DSS 4 (b) (after 2000 s at 1100°C in a salt bath furnace) and a single phase austenite with comparable SFE (c).](image)
textures differ conspicuously.

As expected from the differences in the cold rolling texture the annealing textures of the ferrite phases of the various steels differed from each other. While DSS 1 (Fig. 8(a)) showed almost the same texture as a single phase ferritic steel (Fig. 7(c)), the ferrite textures of the other DSS as represented by DSS 4 (Fig. 8(b)) were clearly different. In these cases the \(\alpha\)-fiber was retained during annealing, and the \(\gamma\)-fiber developed less strongly especially at the \([111]_{/112}[/111]_{/112}\)-orientation. In contrast, DSS 1 and the single phase ferritic steel exhibited a \(\gamma\)-fiber with strong intensities at the \([111]_{/112}\)-orientation (Figs. 8(a), 8(c)).

The effect of increasing degree of cold rolling (30, 60 and 90\%\) on the annealing textures of the ferrite phases is shown in Fig. 9 for DSS 1 and in Fig. 10 for DSS 4. The \(\gamma\)-fiber intensities grew during annealing with increasing rolling deformation. The strength of the \([111]_{/112}\)-orientation changed markedly in the annealed condition after about 60\% cold rolling, i.e. after a critical deformation the \([111]_{/112}\)-orientation started to dominate the annealing texture. In DSS 4 a weak increase of the \([111]_{/112}\)-orientation was noticed with increasing rolling deformation (Fig. 10).

4. Discussion

4.1. Microstructure Evolution

In the austenite phase the described microstructure development during annealing was obviously due to primary recrystallization (Fig. 2(b)) and the nucleation, the development of primary recrystallization and subsequent grain growth were easy to recognize. A bamboo-like structure developed almost instantly upon annealing between 3 and 10 s annealing time in the ferrite phase. From the micrographs it
was not obvious which softening mechanism controlled this fast evolving phenomenon. Both strong recovery, i.e. a recrystallization in-situ, or primary recrystallization or grain growth could have caused this development. The fast evolution of the bamboo-like structure is indicative of strong recovery (recrystallization in-situ), since static recrystallization is preceded by an incubation period due to recovery controlled nucleation. Moreover, recovery reduces the driving force for recrystallization. The TEM and texture measurements strongly support the hypothesis that annealing of the ferrite phase was dominated by recovery.

During further annealing the microstructure changed both with regard to phase arrangement and phase morphology while the microstructural length scale and phase fractions remained essentially stable (Fig. 4). The conversion of the “bamboo structure” to a “pearl structure” was already evident after short annealing times at grain junctions (Figs. 4, 5). At triple junctions comprised of one grain boundary and two phase boundaries the flat layered phase structure broke up into smaller pieces by penetration of the other phase along the grain boundaries (Fig. 11). Evidently, this process was initiated by any adjustment of the contact angles at triple junctions of phase and grain boundaries. The subsequent development towards a more globular structure and break-up of the phase lamella was driven by a reduction of the total interfacial energy due to Ostwald-ripening (Fig. 11). When the lamella thickness was sufficiently small, e.g. in the 90% rolled microstructure, this adjustment was not attained before break up of the lamella. The same phenomenon was reported for thin films, which break up by grooving. This evolution of microstructure is diffusion controlled whereby the microstructure dimension dramatically influences the kinetics of the process and thus, the annealing time to reach this “pearl structure”. At higher rolling degrees the microstructure dimensions strongly decreased and thus provided a higher density of fast diffusion paths by means of grain and phase boundaries. Also the necessary diffusion length was decreased.

During annealing in a conventional furnace (1050°C) DSS 2 and 4 temporarily underwent massive precipitation and ferrite to austenite phase transformation. In brief, the precipitating chromium-rich \( \sigma \)- or \( \chi \)-phase depleted the ferrite matrix of ferrite stabilizing elements and thus promoted transformation to austenite. In DSS 4, which had the same microstructure dimensions as DSS 2 but with much higher nitrogen content (0.4 %) no precipitates formed and no transformation occurred, since nitrogen inhibited precipitation and thus, transformation. On the other hand DSS 1, with a lower nitrogen content, did not cause precipitation or transformation. In this case the microstructure dimensions were much larger and thus diffusion proceeded much more slowly. Before precipitates could develop the temperature in the material had already reached 1050°C, which is slightly higher than the homogenization temperature, where the phase fractions are stable. Consequently, DSS 5 which had the largest austenite phase fraction and grain size and additionally the highest nitrogen content, was not affected by precipitation during slow heating. This stresses the impor-
tant influence of the microstructure dimensions on diffusion controlled microstructure development, e.g. liability to precipitation and phase transformation. After phase break-up the further ripening progressed very slowly because it became controlled by slow volume diffusion after the complete separation of the phase areas.

4.2. Annealing Textures

In contrast to the deformation texture of austenite in DSS and in single phase steel the texture development during annealing of the austenite phase in the DSS resulted in a texture different from single phase austenite (Fig. 7(c)), i.e. the recrystallization process must have been affected by the topology of the deformed microstructure. In single phase materials with low SFE like brass or single phase austenitic stainless steels the BR-orientation (236)/(385) typically develops during recrystallization (Figs. 7(a), 7(b)). This is commonly explained by nucleation and annealing twinning in the shear bands of these materials and subsequent growth selection of grains with 40° (111) orientation relationship with respect to the brass (rolling) orientation.

In concordance with expectation resulting from the observed difference of deformation texture in ferrite the ferrite annealing textures of the various DSS differ from each other. Meanwhile these texture evolutions are known after annealing of typical single phase ferritic steels. It is well known that the active recrystallization mechanisms and thus, the annealing texture depends on the deformation texture. In ferritic steels the (001) fiber or at least a strong (111) fiber will be realized by a low grain boundary mobility e.g. 

4.3. Effect of Microstructure on Texture

The texture measurements and especially the TEM investigations, which revealed that subgrains constituted the ferrite bamboo segments, substantiated that recovery dominated the softening and thus the annealing texture of the ferrite phase of DSS 4 (Figs. 9, 10). The low intensity at (111)/(112) besides other poorly developed orientations indicated the suppression of static recrystallization during softening. The recovery controlled annealing texture of the ferrite phase of the DSS strongly resembled the annealing texture of single phase ferritic steels and did not reveal any influence of the concurrent second phase.

The major reason for dominating recovery during annealing of the ferrite phase is the high SFE of bcc ferrite, which prevents dislocation dissociation and the efficiency of thermal activation on dislocation mobility which strongly promotes recovery. Additionally, and this is a second phase effect, the ferrite phase segments are surrounded by the austenitic phase and thus, consist of only a small number of ferrite next neighbor grains. This means that the nucleation sites for recrystallization are reduced because grain boundaries contiguous to ferrite grains are rare. The retarded nucleation and the reduced dislocation density due to recovery essentially suppressed recrystallization.

5. Conclusions

The annealing behavior of cold rolled DSS was found to be affected by the presence of the two phases. In spite of the fact that the deformation mechanisms, respectively the deformation textures did not differ conspicuously from those of single phase materials, the environmental conditions for the softening mechanisms had changed.

The microstructural phase arrangement in DSS markedly affects the nucleation conditions for recrystallization. For the austenite phase the absence of shear bands after cold rolling engenders the development of different annealing texture orientations compared to fcc single phase material. Moreover, suppression of growth selection disfavors the development of a BR texture. Hence, annealing texture development of the austenite phase in DSS is markedly different from annealing texture evolution in austenitic single phase material.

The high SFE of ferrite and the layered grain morphology strongly favor recovery over recrystallization during annealing. The characteristic microstructural length scale of the phase topology in DSS remained very stable even after long annealing times. The deformation induced pancake-like microstructure strongly affects diffusion by providing...
high diffusivity paths along grain and phase boundaries and thus controls diffusion controlled processes like precipitation and Ostwald ripening. Since the microstructural dimensions are controlled by deformation the degree of deformation markedly influences the annealing behavior of DSS apart from its effect on recrystallization.

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