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[Lim Jaehan](#) , Hwang Soonhong , Lee Sangwon , [Jang Byounglok](#) *

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Article

Analysis of Mechanical Properties SCR420 Low Alloy Steel According to Heat Treatment Before Cold Forging

Jaehan Lim ^{1,2}, Soonhong Hwang ¹, Sangwon Lee ¹ and Byounglok Jang ^{3,*}

¹ Hyundai Steel R&D Center, Dangjin 1480, Republic of Korea; babyatom3@hyundai-steel.com or babyatom3@inha.ac.kr

² Dept. of Materials Science and Engineering, Inha University, Incheon 22212, Republic of Korea

³ Dept. of Materials Processing and Engineering, Inha Manufacturing School, Incheon 21999, Republic of Korea

* Correspondence: jang.bl@inha.ac.kr

Abstract: This study developed a new heat treatment method, normalizing & stress relief (N.S.R), to increase productivity compared to spheroidizing annealing (S.A). The influence of different microstructures resulting from these heat treatments was investigated in cold-forged steel. Despite a shorter heat treatment time, the mechanical properties of the N.S.R alloy were found to be similar to those of the S.A alloy. The factors influencing the mechanical properties of the experimental alloys were analyzed using the Hall-Petch equation, and the predicted values closely matched the measured strength of hyper-eutectoid steels. The primary factors affecting mechanical properties were microstructure and dislocation density. In the case of the S.A alloy, the microstructure exhibited lower strength due to the spherical cementite structure. In contrast, the N.S.R alloy had lower strength because of a reduced dislocation density. This was achieved by stress-relief heat treatment below the A_1 temperature after phase transformation. Based on the mechanical properties, cold forging simulations showed that the effective stress during cold forging of the N.S.R alloy was similar to that of the S.A alloy.

Keywords: cold forging; cementite; spheroidizing annealing; SCR420 steel; automotive commercial components

1. Introduction

In response to global warming, the Paris Agreement to keep the temperature increase to well below 2°C above pre-industrial levels came into force in 2016. In response, major countries around the world are introducing carbon border taxes and adopting net-zero policies, where the sum of carbon savings and emissions is zero in 2050 [1]. As a result, the automotive industry is making various attempts to comply with greenhouse gas reduction policies [2–4]. In particular, in the field of automotive drive components, forging and heat treatment processes are being changed to improve material recovery and reduce carbon emissions. Forging is the process of manufacturing automotive parts by applying heat and pressure to a metal to plastically deform it. Depending on the temperature, forging is categorized into hot forging and cold forging. [5,6]. Hot forging heats the workpiece above its recrystallization temperature, resulting in excellent formability and productivity. However, it has disadvantages such as thermal cracking, poor dimensional accuracy, and surface quality [7].

On the other hand, cold forging is a forging process that takes place at room temperature, so the dimensional accuracy of the alloy is very good. However, the material must be heat treated before forging to ensure ductility [8]. In particular, the cold forging process is carried out after forming a ductile microstructure because a large deformation occurs at room temperature at a high rate during forging. However, S.A, which improves cold forging, requires a long time at A_1 temperature or higher

to obtain ductility of the material, so it is not efficient in terms of forging productivity and emitting a large amount of CO₂ gas [9–11].

In this study, we proposed a new N.S.R eco-friendly heat treatment process that can effectively shorten the heat treatment time while maintaining the same mechanical properties as S.A heat treated materials. To confirm the effectiveness of N.S.R. heat treatment, the test alloy was selected SCR420 steel, which is widely used in automotive parts, and the microstructure analysis and mechanical properties were compared according to the heat treatment method [12].

The Hall-petch model was used to investigate why the mechanical properties of S.A and N.S.R alloys are similar despite the reduced heat treatment time. The effective stresses in cold forging of S.A. and N.S.R. heat treatments were also compared using by AFDEX, a finite element analysis software for plastic processing.

2. Materials and Methods

2.1. Development of N.S.R Heat Treat Ment

In industry, spheroidizing annealing to achieve ductility in materials are typically held near A₁ temperature for at least 20 hours and then cooled [9,13,14]. This heat treatment is difficult to secure productivity and emits a lot of CO₂ gas, so it needs to be improved from a long-term perspective. In this study, we developed a new heat treatment method (N.S.R) that can compensate for the shortcomings of spheroidizing annealing, and the schematic diagram of the heat treatment is shown in Figure 1.

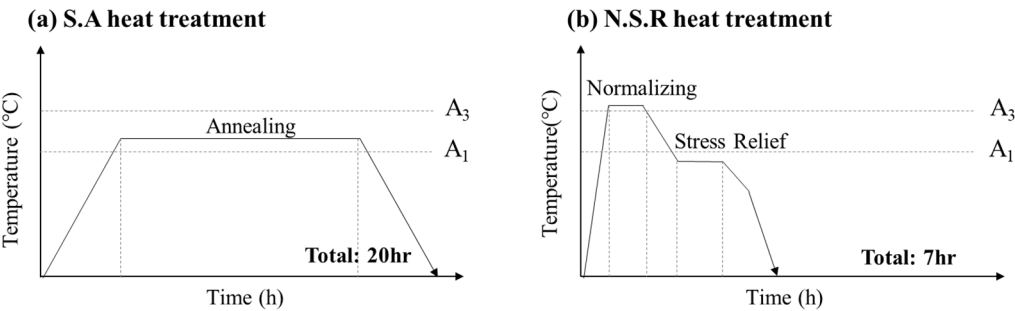


Figure 1. Schematic of heat treatment (a) S.A, (b) N.S.R.

N.S.R heat treatment is a heat treatment method that applies a combination of normalizing and stress relief heat treatment, and the heat treatment time is shortened to 7 hours by controlling the heat treatment temperature and cooling rate. In order to improve the strength and microstructure homogenization of the alloy, it was cooled to below A₁ temperature after maintaining a certain time in the austenite region above A₃ temperature. The alloy was then heat treated and annealed below A₁ temperature to minimize the residual stresses generated during the phase transformation from γ (austenite) $\rightarrow$ α (ferrite) + θ (cementite).

N.S.R improves productivity by about 70% compared to conventional S.A. heat treatment, which can significantly reduce manufacturing costs and carbon emissions. In this study, S.A and N.S.R heat treatments were performed based on SCR420 material with the composition as shown in Table 1. The SCR420 material utilized in this study is Φ 60 bar steel from hyundai-steel.

Table 1. Chemical composition of SCR420 (wt.%).

Chemical composition	C	Si	Mn	Cr
SCR420	0.22	0.25	0.80	1.20

2.2. Microstructures and Mechanical Properties

To analyze the microstructure of the experimental alloys before and after heat treatment, the vertical cut surfaces of the $\Phi 60$ rolled bar and heat treated material were mechanically polished with silicon carbide paper (#100~#2000) and diamond suspension at $1\mu\text{m}$ level and etched with a 3% solution of Nital. Microstructural analysis measurements were analyzed using an OM (Olympus Gx51), SEM-EDS (JEOL, 7100F) microscope. The ferrite fraction was measured according to ASTM E562, and the remaining volume fraction other than the measured ferrite fraction was considered as pearlite.

The cementite spacing of the experimental alloys was measured by applying the mean free path (MFP) method. For quantitative analysis of MFP, at least 10 SEM photographs at $\times 20,000$ magnification were taken for each condition. Based on this, the mean random spacing obtained was divided by 2 to obtain the mean true spacing as shown in Eq. (1) [15].

$$S_r = \frac{3.14 \cdot D}{2n} \quad (1)$$

Where S_r is the average distance between cementites, D is the diameter of the measurement area, and n is the number of cementites measured in that area. Grain size and dislocation density were measured using EBSD (Oxford Symmetry) mounted on the FE-SEM. Grain size was measured at a magnification of $\times 500$, step size of $0.2\mu\text{m}$, tilt of 0% , and accelerating voltage of 20 kV . Dislocation density was measured at $\times 1000$ magnification, step size $0.2\mu\text{m}$, slope 80% , and acceleration voltage 20 kV .

To evaluate of the mechanical properties of S.A and N.S.R alloys, samples were taken from the $1/4$ point of the cross-section of the $\Phi 60$ bar and evaluated in tensile and compression tests at room temperature. The tensile specimens were evaluated using a tensile testing machine (Zwick/Roell, Z250) using a proportional test specimen, KS 14A (diameter 6 mm , marking distance 30 mm). The tensile test was performed at 0.009 mm/s up to yield and 0.241 mm/sec during tension.

The compression test was performed using a compression tester (Zwick/Roell, Z600) with a 15 mm compression specimen of $\Phi 10$. The compression test was performed at a speed of 3.3 mm/s to compress the specimen to 60% of its cross-sectional area.

2.3. Analysis of Cold Forging Formability

In order to validate the cold forging process of N.S.R alloy, a commercial forging analysis was conducted on the input shaft, an automotive drive part. The input shaft cold forging process consists of three processes: forward extrusion, backward extrusion, and flange upsetting, and the molds and models for each process are shown in Figure 2.

The finite element model of the input shaft is made of SCR420 $\Phi 60$ bar steel cut into 300 mm length, and the molds of each process are modeled at a $1:1$ scale and meshed with Tetrahedral Brick Element. AFDEX, a commercial forging finite element analysis software, was then used to analyze the forming of each process. Stress-strain curves for the cold forgeability evaluation of S.A and N.S.R alloys were based on room temperature tensile test results. In addition, young's modulus, poisson's ratio, density, and thermal expansion coefficient were calculated by J mat. pro, a thermodynamic simulation program, and their properties are shown in Table 2.

To avoid direct friction between the mold and the workpiece during cold forging, a phosphating coating treatment is applied to the workpiece, and the Coulomb friction coefficient between the forging mold and the workpiece is set at 0.1 , and elastic deformation analysis is performed [16].

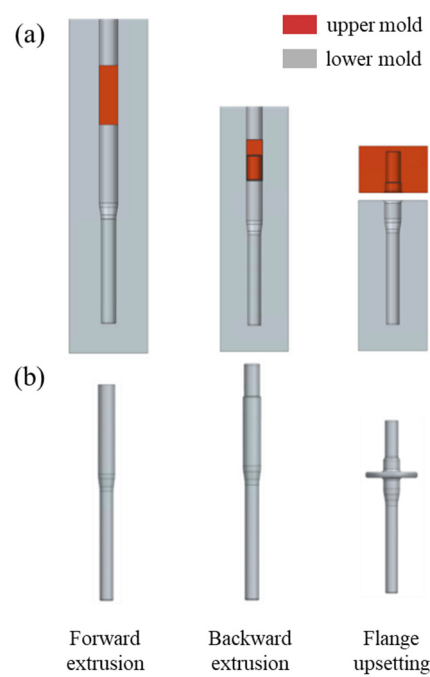


Figure 2. Finite element Model by 3stage process (a) Mold, (b) Input shaft-part.

Table 2. Properties of experimental alloys.

Properties	Unit	Value
Young's modulus	GPa	210
Poisson's ratio	-	0.3
Density	g/cm ³	7.85
Thermal expansion coefficient	1/°C	1.2E-05

3. Results and Discussion

3.1. Microstructures and Mechanical Properites

Figure 3 shows microstructure measurements of as-rolled and heat-treated steels. Figure 3a shows the microstructure of the as-rolled material of SCR420 steel grade, which has a mixed ferrite and pearlite microstructure. In addition, Figure 3. (e) shows the micrograph of pearlite at ×30,000 magnification and confirms that it has a fine lamella structure of cementite.

S.A alloy exhibited a typical spherical cementite texture as shown in Figure 3b [17]. As shown in the schematic diagram in Figure 4, S.A heat treatment causes the layered cementite in the pearlite to segment above the initial A₁ temperature. Then, with prolonged heat treatment, some of the carbons are dissolved and residual cementite remains. In the subsequent heat treatment and cooling stages, the undissolved residual cementite acts as a nucleation site to grow spherical cementite [11,18]. On the other hand, the microstructure of N.S.R. heat-treated alloy has a mixed organization of layered and spherical cementite as shown in Figure 3g. This is because the cementite dissolved at A₃ and above grows into spherical cementite as it is cooled to below A₁ temperature, and the growth of cementite as it is maintained at temperatures below A₁ results in a mixed lamellar and spherical microstructure.

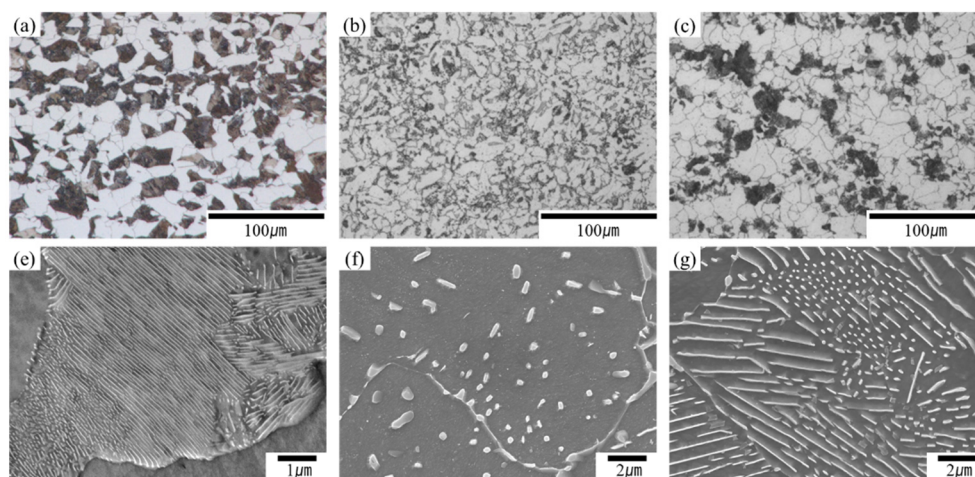


Figure 3. Microstructure measured by OM (a) As-rolled, (b) S.A, (c) N.S.R alloys, by SEM (e) As-rolled, (f) S.A, (g) N.S.R alloys.

The ferrite fraction of the experimental alloys was measured base on the as-rolled and heat treatment conditions. The ferrite fraction was measured to be 55% for the as-rolled condition, 80% S.A and 74% N.S.R, respectively. The main reason for the difference in ferrite fraction between as-rolled and heat-treated methods is the cooling rate and cooling start temperature.

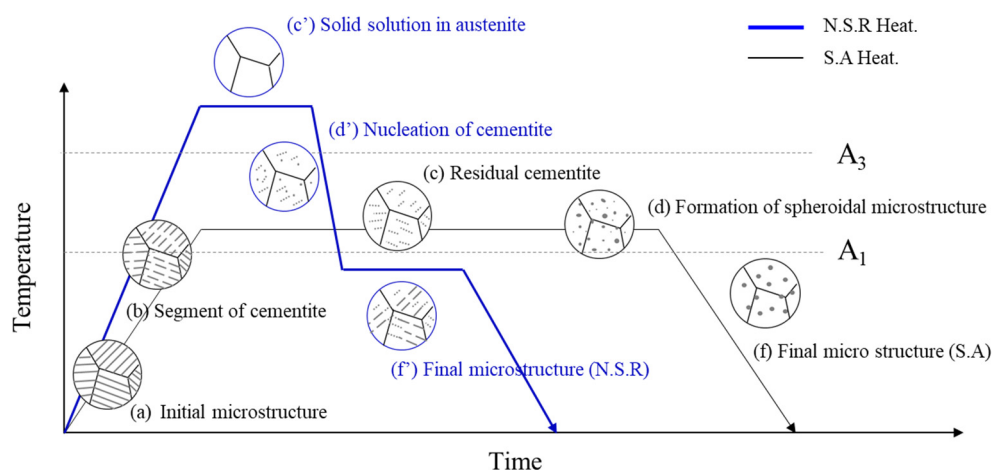


Figure 4. Schematic of microstructure by heat treatment stage (black line) S.A, (blue line) N.S.R.

Slower cooling rates increase the fraction of ferrite, while faster cooling rates increase the fraction of pearlite. This is due to the shortening of the shorter distance that the dissolved carbon atoms in the austenite can travel [18]. In other words, the accumulation of carbon concentration at the austenite/ferrite interface makes it easier to form pearlite at high temperatures, and the interlamellar distance of cementite becomes smaller. In general, the cooling rate of $\Phi 60$ bar steel after rolling is measured to be about $0.5\sim 1^\circ\text{C/s}$ for air cooling and about 0.1°C/s for furnace cooling. Therefore, the pearlite fraction of as-rolled with the fastest cooling rate is the highest.

In addition, when the cooling start temperature is lower than A_3 , as in S.A. heat treatment, austenite and cementite coexist in the region, so the shape of cementite is affected by the heat treatment temperature [11,13]. Therefore, even if the cooling rate of S.A. alloy and N.S.R. alloy is the same at 0.1°C , the ferrite fraction is different.

The tensile and compressive strength results of the S.A and N.S.R alloys are shown in Figure 5 and Table 3, respectively.

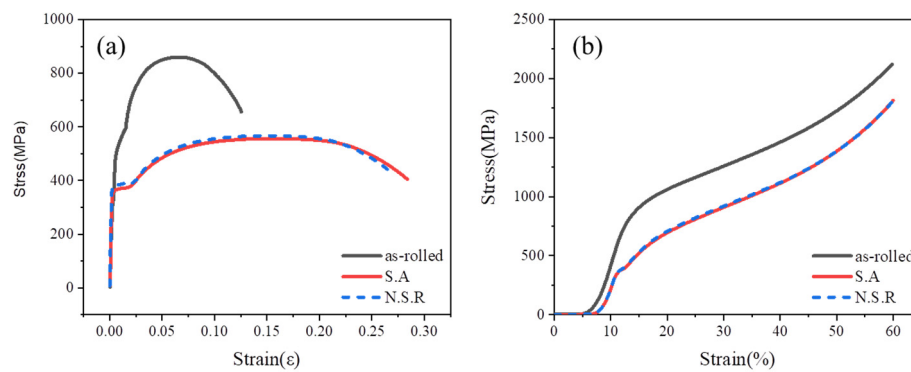


Figure 5. Result of mechanical properties at the room temperature (a) Tensile (b) Compressive stress-strain curve.

The tensile test results showed that the yield strength of as-rolled was 565 MPa and the ultimate tensile strength was 842 MPa. In comparison, the yield strength of S.A was 377 MPa, and the tensile strength was 556 MPa. Finally, the yield strength of N.S.R was measured at 382 MPa and the tensile strength to be 567 MPa.

In addition, the compressive strength of the as-rolled material at 60% cross-sectional area compression was measured to be 2,119 MPa, while the compressive strengths of the S.A and N.S.R heat-treated alloys were 1,810 MPa and 1,815 MPa, respectively. Despite the different heat treatment methods, the tensile and compressive properties were found to be similar.

Table 3. Mechanical properties of experimental alloys.

Mechanical properties	Tensile Properties			Compressive Stress (MPa)		
	Y.S ¹ (MPa)	U.T.S ² (MPa)	Elongation (%)	20%	40%	60%
as-rolled	565	842	17%	1,170	1,520	2,119
S.A	362	556	30%	805	1,169	1,810
N.S.R	382	567	28%	810	1,174	1,815

¹ Y.S: Yield Stress, ² U.T.S: Ultimate Tensile Stress.

3.2. On the Strength Mechanisms

In general, the mechanical properties of pearlite steels can be described by the Hall-Petch relationship as shown in Eq. (2) [19].

$$\sigma = \sigma_o + K_y \lambda^{-\frac{1}{2}} \quad (2)$$

Where σ is the yield strength σ_o is the friction stress, K_y is the Hall-petch constant and λ is the interlamellar spacing of pearlitic steels. However, it is difficult to express hypo-eutectoid steels in Eq (2), because they have a dual-phase structure of ferrite and pearlite. Therefore, various studies have been attempted to quantify the strength of steels with a dual phase of ferrite-pearlite steels as shown in Eq. (3), including those by O'Donnelly, Reuben, Baker et al. [20].

$$\sigma = \sigma_f + K_y \lambda_p^{-\frac{1}{2}} \quad (3)$$

Where is σ_f the friction stress of pure ferrite, K_y is the constant, λ_p is the mean ferrite slip distance in the composite and λ_p is defined as Eq. (4). [20].

$$\lambda_p = V_p(S_r - T_r) + V_a + d_a \tag{4}$$

Where V_p and V_a are the pearlite and ferrite fractions, respectively, d_a is the ferrite grain size. S_r is the interlamellar spacing, T_r is the mean cementite thickness, expressed by Eq. (5) [21].

$$T_r = \frac{0.15 \cdot Carbon(wt\%) \cdot S_r}{V_p} \tag{5}$$

Based on the above equations, this study expresses the yield strength of hypo-eutectoid steels as Eq. (6) and (7) below.

$$\sigma_m = K_y[V_p(S_r - T_r) + V_a d_a]^{\frac{1}{2}} \tag{6}$$

$$\sigma_y = \sigma_f + \Delta\sigma_m + \Delta\sigma_{ss} + \Delta\sigma_{dis} + \Delta\sigma_{particle} \tag{7}$$

Where σ_f is the ferrite strength, $\Delta\sigma_{ss}$ is the solid solution hardening, $\Delta\sigma_{dis}$ is the dislocation hardening, and $\Delta\sigma_{particle}$ is the precipitation hardening. However, the experimental alloys used in this study have a very low contribution to precipitate formation, so the $\Delta\sigma_{particle}$ effect was excluded

Initially σ_f was used as $\sigma_f= 70$ MPa for pure Fe [22]. K_y is a material constant that varies with heat treatment conditions. Therefore, experimental values were applied, and K_y was calculated from the yield zone slopes of the experimental alloys. As shown in Table 4, K_y values for the experimental alloys were measured as: as-rolled 601 MPa/ $\mu\text{m}^{1/2}$, S.A 411 MPa/ $\mu\text{m}^{1/2}$, N.S.R 430MPa/ $\mu\text{m}^{1/2}$. The as-rolled K_y values were within the typical range of 440 to 760MPa/ $\mu\text{m}^{1/2}$ for ferrite-pearlite steels [23].

Table 4. K_y value of experimental alloys.

K_y	as-rolled	S.A	N.S.R
unit : MPa/ $\mu\text{m}^{1/2}$	601	411	430

S.A and N.S.R's K_y values were found to be similar to the K_y value (420 MPa/ $\mu\text{m}^{1/2}$) obtained from spheroidizing heat treatment in a similar carbon range studied by L. Anand's. This phenomenon of decreasing K_y value with slower cooling rate is due to the transformation of lamellar structure cementite into spheroidal form and distribution within the grains [23,24]. Therefore, the K_y values of S.A and N.S.R alloys were lower than that of as-rolled.

The S_r of the experimental alloys was calculated using Eq. (1) to obtain the average true spacing of cementite, and T_r was calculated based on Eq. (5). S_r was respectively measured as 0.12 μm for as-rolled, 3.6 μm for S.A and 1.3 μm for N.S.R. T_r was measured 0.01 μm for as-rolled, 0.43 μm for S.A, and 0.19 μm for N.S.R alloys.

The average grain size has a significant influence on the strength of metallic materials because high-angle grain boundaries (HAGBs) inhibit the movement of dislocations [25,26]. d_a is the result of using EBSD, and the average HAGB sizes were measured as 7.42 μm for as-rolled, 6.55 μm for S.A, 6.75 μm for N.S.R alloys, as shown in Figure 6.

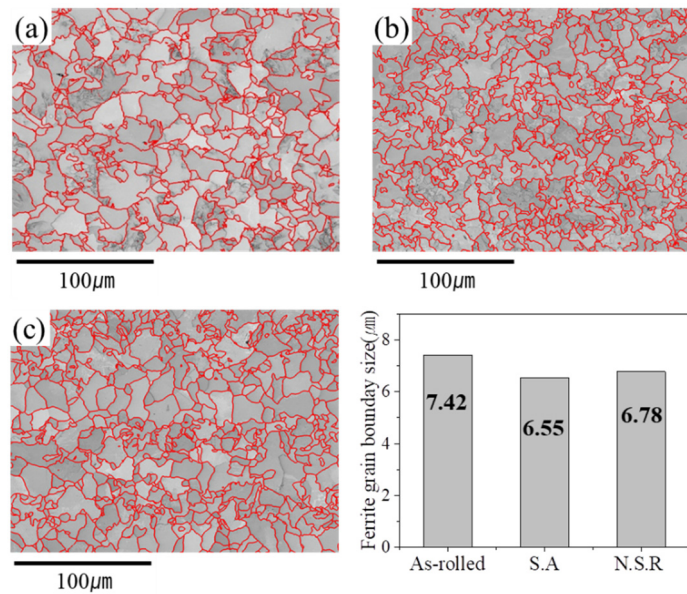


Figure 6. Analysis of grain boundary size of experimental alloys by EBSD (a) as-rolled, (b) S.A, (c) N.S.R alloys.

The strength of $\Delta\sigma_m$ utilizing grain size and cementite parameters, was measured to be 293 MPa for as-rolled, 167 MPa for S.A and 185 MPa for N.S.R alloys, respectively.

The strengthening mechanism for the chemical component can be explained as shown in Eq. (8) below.

$$\Delta\sigma_{ss} = 47[\text{Si}] + 30[\text{Mn}] + 31[\text{Cr}] \quad (8)$$

K.K. Ray reported that Mn and Si increase by 30 and 47 MPa per wt%, respectively [27]. Gutierrez also reported that Cr increases by 11 and 31MPa per wt% [28].

Thus, the effect of solid solution strengthening by composition was calculated to be 82.6MPa. Another strengthening mechanism, dislocation strengthening, is expressed by Taylor's equation, Eq. (9). [26].

$$\Delta\sigma_{dis} = \alpha G b \sqrt{p} \quad (9)$$

Where α is the constant (0.38), G is the shear modulus (81GPa), b is the burgers vector (2.48Å), and p is the dislocation density [17,29,30].

The dislocation density of the experimental alloys was measured using Geometrically Necessary Dislocation (GND) density maps as shown in Figure 7, and the dislocation densities were measured as follows as: as-rolled $1.97 \times 10^{14} \text{ cm}^{-2}$, S.A $0.95 \times 10^{14} \text{ cm}^{-2}$, N.S.R $0.64 \times 10^{14} \text{ cm}^{-2}$. $\Delta\sigma_{dis}$ was calculated to be 107MPa for as-rolled, 74MPa for S.A, and 61MPa for N.S.R alloys, respectively.

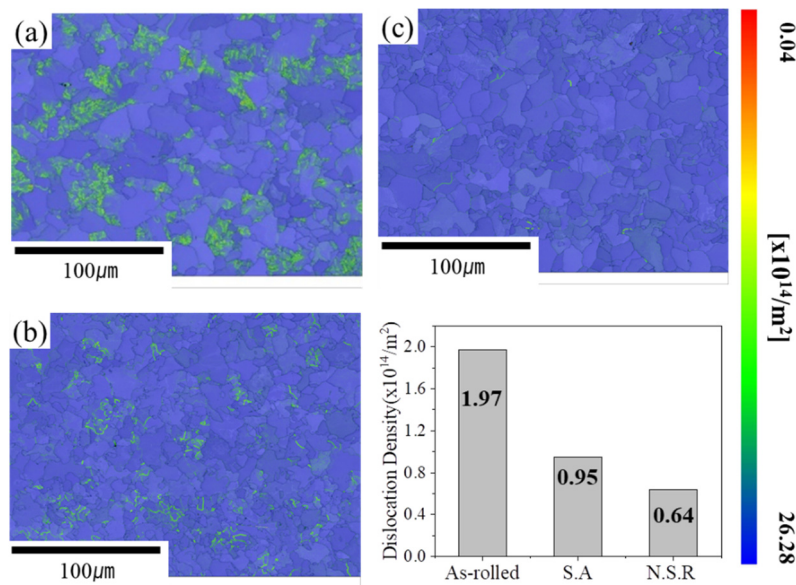


Figure 7. EBSD GNDs distribution maps (a) as-rolled, (b) S.A, (c) N.S.R alloys.

The sum of the predicted and experimentally measured values for each strengthening mechanism is shown in Figure 8.

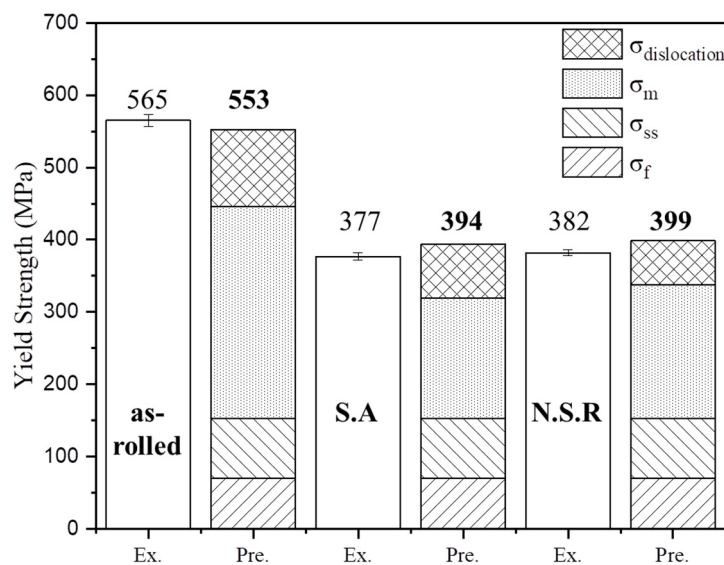


Figure 8. Comparison of the contributions of each strengthening mechanism to the as-rolled, S.A, N.S.R alloys (Ex: Experimental, Pre: Predicted).

The yield strength was predicted using parameters based on the heat treatment process, and the predicted total strength was measured to be 553 MPa for as-rolled, 394 MPa for S.A, and 399 MPa for N.S.R alloys. These values are similar to the experimentally measured values, with a deviation of about $\Delta 17$ MPa. The difference between the measured and predicted yield strengths for the S.A and N.S.R alloys was the same, at $\Delta 5$ MPa. This indicates that the predicted values for each parameter affecting the yield strength were calculated correctly.

The difference in strength between the S.A and N.S.R alloys was $\Delta 18$ MPa, depending on the microstructure variables. However, the reason why N.S.R alloys exhibit similar strength to S.A materials is due to the reduction in dislocation density. Compared to as-rolled, the dislocation density of S.A and N.S.R alloys was reduced by 52% and 68%, respectively. This resulted in a 13 MPa

reduction in the dislocation-induced strengthening effect of the N.S.R alloy compared to the S.A. alloy.

In general, when steel passes through the A_1 eutectoid temperature, a phase transformation dislocation is formed during the γ (austenite) $\rightarrow \alpha$ (ferrite) + θ (cementite) phase transformation.

In the case of the S.A alloy, as shown in Figure 1a, heat treatment proceeds above the A_1 temperature to form austenite and spherical cementite, and the dislocations disappear. However, dislocations are generated by the newly grown pearlite during cooling [11,13]. On the other hand, in the case of the N.S.R alloy, stress relief heat treatment is carried out below the A_1 temperature after the phase transformation is completed, so the dislocation strengthening effect is lowered as the dislocations disappear.

Figure 9 shows the microstructure and GND density of the as-rolled material at high magnification. The dislocation density is low in the ferrite region, while it is high in the pearlite region. These results demonstrate that dislocations are formed as a result of phase transformation.

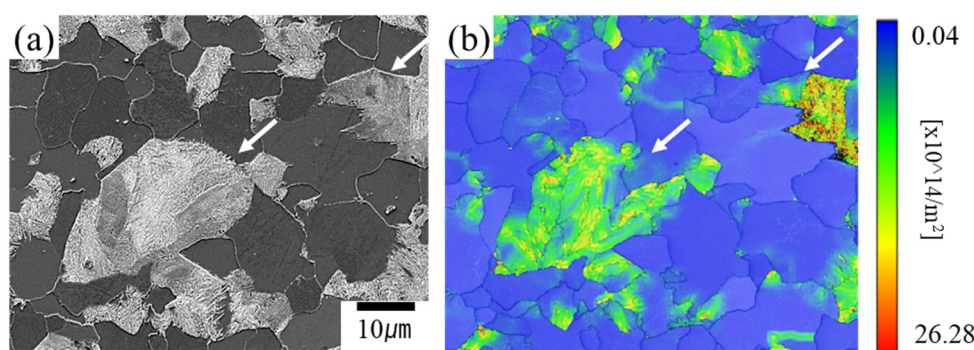


Figure 9. Microstructure of as-rolled (a) SEM image, (c) GND Map.

3.3. Analysis of Cold Forging Formability

The purpose of the effective strain and stress analysis is to compare the influence of the mold in the continuous production process of automotive parts based on the heat treatment method. Figure 10 shows the results of analyzing the cold forging process of the input shaft using the mechanical properties of the experimental alloys when S.A and N.S.R heat treatments are applied, which include (a) effective strain and (b) effective stress. Both S.A. and N.S.R. alloys exhibited similar strain rates in the flange upsetting process with the highest compression ratio.

In addition, the maximum effective stress at the flange was 874 MPa for both S.A. and N.S.R alloys, with no significant difference observed in the extrusion process. Therefore, it was confirmed that the maximum stress of the N.S.R alloy during the forging process is similar to that of the S.A alloy. This similarity is attributed to the comparable mechanical properties of the two experimental alloys. Based on these results, the application of the N.S.R heat treatment method in the automotive industry is expected to exhibit cold forging properties similar to those of the S.A heat treatment.

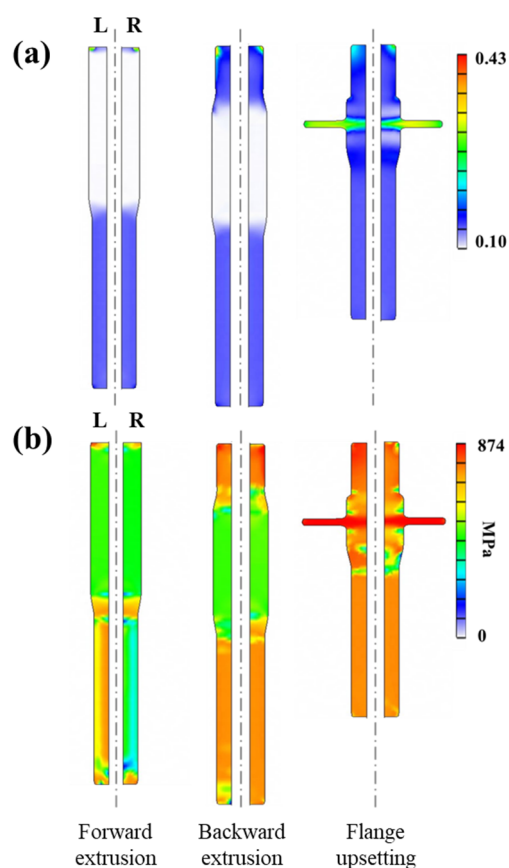


Figure 10. Finite element results of cold forging of experimental alloy, left: S.A, right: N.S.R (a) effective strain, (b) effective stress.

4. Conclusions

In this study, an eco-friendly N.S. heat treatment method with mechanical properties similar to those of the S.A heat-treated material was proposed. Changes in microstructure and mechanical properties were observed as a function of the heat treatment method. The mechanical properties were investigated using the Hall-Petch equation, and the cold workability was predicted through finite element analysis. The following conclusions were drawn:

1) N.S.R heat treatment takes 7 hours, compared to S.A heat treatment, which takes more than 20 hours, reducing the heat treatment time by approximately 65%.

2) As a result of S.A. and N.S.R heat treatment of SCR420 steel grades, the tensile strengths were 556 MPa and 567 MPa, respectively, and the strength when compressing 60% of the cross-sectional area was 1,810 MPa and 1,815 MPa, respectively, showing similar mechanical properties.

3) The yield strength of the predicted and experimental alloys was calculated using the Hall-Petch formula, and the yield strengths of the S.A and N.S.R materials showed similar values with a difference of $\Delta 5$ MPa. The difference between the actual yield strength and the predicted yield strength was $\Delta 17$ MPa, confirming the validity.

4) The morphology of cementite in SCR420 steel changed depending on the heat treatment method. In the case of as-rolled before heat treatment, the cementite has a fine lamellar structure, while in the case of the S.A alloy, the cementite has a spherical shape. In the case of N.S.R, the cementite exhibits both a lamellar structure and spherical shape. The influence of these microstructures on the yield strength was calculated to be 229 MPa, 169 MPa, and 186 MPa, respectively, with the microstructure of the S.A alloy showing the lowest effect.

5) The reason why the N.S.R alloy exhibits similar mechanical properties to the S.A alloy is related to the dislocation density. In the case of the S.A alloy, dislocations are formed by phase

transformation during cooling after the heat treatment is completed. However, in the case of the N.S.R, dislocations are eliminated due to the stress relief heat treatment process that occurs after the phase transformation is completed.

6) Compared to the as-rolled material, the dislocation density of the S.A and N.S.R alloys was reduced by 52% and 68%, respectively, and the strengthening effect due to dislocations was calculated to be 74 MPa and 61 MPa, respectively. In other words, the reason why the N.S.R alloy has similar mechanical properties, despite the reduced heat treatment time, is due to the dislocation density.

7) The elastic deformation analysis of the cold forging process for automotive input shafts showed that the effective stresses of the S.A alloy and N.S.R alloy differed by up to $\Delta 24$ MPa, with the effective stresses during forging being equivalent.

8) Based on these results, it is expected that the N.S.R heat treatment can be applied as a pre-heat treatment before cold forging automobile parts, contributing to the reduction of carbon emissions.

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