Hot Deformation Post-Imprint of Spray-Formed Nb-Containing High-Speed Steel Based on Hot Processing Maps

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Date: 2023-03-18T00:00:00+00:00

Abstract

A novel M3:2 high-speed steel was prepared by combining the alloying strategy of partial vanadium substitution with niobium and spray forming rapid solidification technology. The hot deformation behavior of the alloy was investigated at temperatures of 950–1150°C and strain rates of 0.001–10 s^{-1}. Based on the experimentally obtained true stress–true strain curves, a hot processing map was established using the Dynamic Material Model (DDM), combined with kinetic analysis and microstructural observation. The processing map was divided into four regions: plastic instability zone (>1 s^{-1}), low strain rate zone (0.001 s^{-1}), low deformation temperature zone (<1000°C), and processing safety zone. The mechanism of crack formation in the low strain rate and low deformation temperature zones was discussed in detail, which led to the determination of the hot processing window as 1050–1150°C at 0.01–0.1 s^{-1}. To achieve a microstructure characterized by fine grains, granular carbides, and uniform distribution, the optimized hot deformation parameters were identified as 1150°C and 0.1 s^{-1}. Following forging and heat treatment, the spray-formed niobium-containing M3:2 high-speed steel demonstrated superior hardness, bending strength, and other mechanical properties compared to powder metallurgy high-speed steel of equivalent composition.

Full Text

Preamble

Vol. 29 No. 7
CHINESE JOURNAL OF MATERIALS RESEARCH
July 2015

Hot Deformation of Spray-Formed Nb-Containing High Speed Steel —A Study Using Processing Map

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Supported by National Basic Research Program of China No. 2011CB606303.

Manuscript received December 31, 2014; in revised form February 28, 2015.

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Abstract

The hot deformation behavior of as-spray-formed Nb-containing AISI M3:2 high speed steel has been investigated by compression tests at temperatures ranging from 950–1150°C and strain rates from 0.001–10 s^{-1} with 50% reduction. Processing maps were developed according to the principles of the Dynamic Material Model. It was found that the flow curves exhibited the classic shape of dynamic recrystallization (DRX)—rising to a peak, followed by softening to a steady state. The hot working process of the steel can be carried out safely in the domain of (Td: 1050–1150°C, $\dot{}$: 0.01–0.1 s^{-1}). To obtain microstructures with fine grains and uniform distribution of fine granular carbides, the hot working process should be carried out at 1150°C and a strain rate of 0.1 s^{-1}. Flow instability occurred when strain rates exceeded 1 s^{-1}. After proper hot working and heat treatment, the hardness and bending strength of the spray-formed Nb-containing M3:2 high speed steel reached 67 HRC and 3467 MPa, respectively.

KEY WORDS metal materials, spray forming, high speed steel, hot deformation, processing map, niobium

Introduction

M3:2 high speed steel (domestic grade CW6Mo5Cr4V3), developed by increasing the carbon and vanadium contents of AISI M3 steel, is widely used for manufacturing cutting tools such as broaches, formed milling cutters, turning tools, and drills for machining medium-strength steels, high-temperature alloys, and other difficult-to-machine materials. It also serves as an excellent cold work die steel. Compared with M3, M3:2 high speed steel offers higher hardness, wear resistance, and good toughness. To further improve the hardness and wear resistance of high speed steels, research has been conducted on substituting niobium—a strong carbide-forming element—for part of the vanadium. Due to the low cooling rate of conventional casting processes, Nb-containing high speed steels tend to form coarse eutectic structures during solidification. At high niobium contents, coarse MC carbides can even precipitate directly from the melt, limiting performance improvements and hindering the development and application of Nb-containing high speed steels. The emergence of powder metallurgy provided new opportunities for developing Nb-containing high speed steels. Thyssen Edelstahl developed two Nb-containing powder metallurgy high speed steels, designated TSP1 and TSP8, which have been patented and commercially applied. However, powder metallurgy involves complex processes, multiple steps, and high costs, restricting its application range and limiting the widespread adoption of Nb-containing high speed steels. In contrast, high speed steels produced by spray forming exhibit fine equiaxed grains, no macro-segregation, and small, uniformly distributed eutectic carbides, showing microstructural characteristics similar to powder metallurgy high speed steels. Moreover, compared with powder metallurgy, spray forming offers simpler processes, fewer steps, and lower costs.

Processing maps have proven effective for optimizing hot working parameters and have been applied to materials including magnesium alloys, titanium alloys, nickel-based superalloys, steels, and aluminum alloys. While studies on the hot deformation behavior of spray-formed high speed steels and the establishment of constitutive equations have been reported, the application of processing maps to investigate the hot deformation behavior of high speed steels remains rare. Liu et al. studied the hot deformation behavior of powder metallurgy M3 high speed steel through isothermal compression and established its processing map. However, these studies focused on microstructural changes within the safe processing zone, without addressing microstructural evolution and crack formation mechanisms in the plastic instability zone, low strain rate region, or low temperature region.

This work investigates the microstructural evolution and crack formation mechanisms during deformation of a spray-formed Nb-containing M3:2 high speed steel (developed independently by our research group with patent number CN 102605263) under various deformation temperatures and strain rates. A processing map for this alloy was established to guide the selection of actual hot working parameters, providing a solid foundation for promoting the practical application of this newly developed alloy.

1.1 Sample Preparation and Property Testing

Commercial W6Mo5Cr4V2 steel was used as the master alloy. During melting in a medium-frequency induction furnace, intermediate alloys including C, W, Mo, V(50%, mass fraction)-Fe, and Nb(65%)-Fe were added. After the molten steel composition was homogenized by electromagnetic stirring, it was poured into a preheated tundish at 1560°C. The melt then passed through a guide nozzle into the atomization chamber, where it was atomized by nitrogen gas at 0.46 MPa pressure. The droplets flew through the atomization gas until reaching the deposition substrate at a distance of 450 mm, forming a deposit with a diameter of 180 mm and height of 70 mm. The composition of the prepared Nb-containing spray-formed M3:2 high speed steel (mass fraction, %) was: C 1.31; W 6.10; Mo 4.90; Cr 4.48; V 2.75; Nb 0.50.

Cylindrical specimens with a diameter of 8 mm and length of 12 mm were machined from the deposit for hot compression testing. Hot compression tests were conducted on a Gleeble-1500 thermal simulator at deformation temperatures of 950, 1000, 1050, 1100, and 1150°C, with strain rates of 0.001, 0.01, 0.1, 1, and 10 s^{-1}, and a deformation degree of 50%. Specimens were heated to 1180°C at 10°C/s, held for 3 minutes, cooled to the deformation temperature at 3°C/s, held for another 3 minutes, and then deformed. Deformed specimens were immediately water-quenched. The deformed specimens were mechanically ground and polished, etched with 8% nitric acid alcohol solution, and examined using a ZEISS SUPRA 55 field emission scanning electron microscope.

Based on the processing map, hot working parameters were selected for forging the deposit. To avoid generating large stresses, the forged material was buried in sand for slow cooling after forging. An infrared thermometer was used throughout the forging process to control the temperature range and ensure that forging occurred in the recrystallization region.

Samples were taken from the forged bar, sealed in vacuum quartz tubes filled with Ar, and heat-treated. The heat-treated samples were used for testing hardness, bending strength, and impact toughness. The heat treatment process was: 900°C for 90 min, furnace cooling + 1200°C for 20 min, oil quenching + 560°C for 60 min, air cooling (repeated 3 times).

Hardness was measured on a TH320 Rockwell hardness tester, with seven points measured on each sample; the maximum and minimum values were discarded and the average was taken as the hardness value. Bending strength was tested by three-point bending according to GB/T 228-2002, with sample dimensions of 5 mm × 5 mm × 35 mm and a span of 30 mm. Impact tests were conducted on a JB-30B testing machine using unnotched samples with dimensions of 10 mm × 10 mm × 55 mm. All four long surfaces were precision ground to ensure high surface finish.

1.2 Principles and Construction of Processing Maps

The Dynamic Material Model (DDM) treats a workpiece undergoing hot deformation as a power dissipation unit. At any instant, the total power input P is dissipated through two complementary routes: energy dissipated by plastic deformation (G) and energy dissipated by microstructural changes (J). According to the principle of power partitioning:

$$P = \sigma\dot{\varepsilon} = G + J = \int_0^{\dot{\varepsilon}} \sigma \, d\dot{\varepsilon} + \int_0^{\sigma} \dot{\varepsilon} \, d\sigma$$

where σ is the true stress and $\dot{\varepsilon}$ is the strain rate. The strain rate sensitivity exponent m is defined as:

$$m = \frac{d \ln \sigma}{d \ln \dot{\varepsilon}}$$

From the above equation, the partitioning of total power between plastic deformation and microstructural changes depends on the m value. The relationship between stress and strain rate can be expressed as:

$$\sigma = K\dot{\varepsilon}^m$$

where K is a material constant dependent on temperature and strain. According to this model, when the true stress-true strain relationship satisfies the above equation, the energy dissipated by microstructural changes (J) can be expressed as:

$$J = \int_0^{\sigma} \dot{\varepsilon} \, d\sigma = \frac{m}{m+1}\sigma\dot{\varepsilon}$$

When $m = 1$, the material is in an ideal linear dissipative state, and J reaches its maximum value $J_{\max}$. For non-linear dissipative states, a dimensionless parameter η is defined to characterize the efficiency of power dissipation during hot deformation:

$$\eta = \frac{J}{J_{\max}} = \frac{2m}{m+1}$$

The power dissipation map is constructed based on the variation of energy dissipation efficiency η with temperature and strain rate.

The Dynamic Material Model is established on the basis of rigorous irreversible thermodynamics. Prasad proposed a dynamic continuum criterion and an instability criterion for material flow based on principles of irreversible thermodynamics and the separability of power dissipation:

$$\xi(\dot{\varepsilon}) = \frac{\partial \ln[m/(m+1)]}{\partial \ln \dot{\varepsilon}} + m < 0$$

By plotting $\xi(\dot{\varepsilon})$ as a function of deformation temperature and strain rate, a flow instability map can be obtained. Superimposing the instability map onto the power dissipation map yields the processing map at a given strain condition. For dynamic analysis of different regions in the processing map, the power-law creep equation is used:

$$\dot{\varepsilon} = A\sigma^n \exp\left(-\frac{Q}{RT}\right)$$

where A is a material constant, $n(= 1/m)$ is the stress exponent, Q is the activation energy, R is the gas constant, and T is the absolute temperature.

2.1 Original Microstructure

The as-deposited microstructure of the spray-formed Nb-containing M3:2 high speed steel is shown in [Figure 1: see original paper]. The deposit exhibits equiaxed grains with an average size of 50 m. Discontinuous lamellar M2C eutectic carbides and irregular blocky MC carbides are distributed along grain boundaries. Additionally, some micropores are present in the as-deposited microstructure. Both the lamellar eutectic carbides and micropores adversely affect the properties of high speed steel, necessitating subsequent thermomechanical processing to improve carbide morphology and distribution and eliminate the inherent microporosity.

2.2 True Stress-Strain Curves

Typical true stress-true strain curves of the alloy under different deformation conditions are shown in [Figure 2: see original paper]. The flow stress is significantly influenced by deformation temperature and strain rate. Under high strain rate conditions, the flow stress at different temperatures exhibits typical dynamic recrystallization characteristics: rapid increase during initial deformation due to work hardening, followed by a peak stress and subsequent softening to a steady state. This occurs because softening mechanisms (such as dynamic recovery and dynamic recrystallization) counteract the hardening effect of work hardening. When work hardening and softening reach equilibrium, the flow stress becomes stable. At lower strain rates (0.001 s^{-1}), the flow stress curves at different temperatures show steady-state characteristics with multiple peaks. This is because dislocations must accumulate to a certain level to form recrystallization nuclei. At low strain rates, dislocation accumulation is relatively slow, and the dislocation accumulation from deformation during one dynamic recrystallization cycle is insufficient to initiate new recrystallization within the recrystallized region. The process involves two competing mechanisms: work hardening in unrecrystallized and recrystallized regions, and softening from ongoing recrystallization. Softening dominates initially, followed by hardening, leading to periodic peaks in the flow stress curve.

The flow stress decreases with increasing temperature. Higher temperatures increase atomic vibration amplitude, weakening atomic bonding and reducing deformation resistance. Additionally, elevated temperatures facilitate dynamic recrystallization nucleation and enhance grain boundary mobility, accelerating softening. Consequently, flow stress decreases with temperature at constant strain rate. Conversely, flow stress increases with strain rate at constant temperature because higher strain rates generate more dislocations and suppress softening processes like dynamic recovery and recrystallization.

2.3 Dynamic Analysis

[Figure 3: see original paper] shows the fitted relationships between stress, strain rate, and temperature in different regions of the processing map at a strain of 0.6, demonstrating good fitting accuracy. The stress exponent is 5.88 in the high-temperature region (1050-1150°C) and 7.32 in the low-temperature region (950-1000°C). According to the power-law creep equation, the activation energy for deformation can be calculated from:

$$Q = R\left[\frac{\partial \ln \sigma}{\partial (1/T)}\right]_{\dot{\varepsilon}}$$

The calculated activation energy is 489 kJ/mol in the range of 1050-1150°C and 0.001-0.1 s^{-1}, and 609 kJ/mol in the range of 950-1000°C and 0.001-0.01 s^{-1}. The significant differences in stress exponent and activation energy indicate that different mechanisms control plastic deformation in these regions.

2.4 Processing Map and Deformation Microstructures

The processing map for the spray-formed Nb-containing M3:2 high speed steel at a true strain of 0.6, constructed based on the Dynamic Material Model, is shown in [Figure 4: see original paper]. Raj and Gandhi used atomistic models to interpret microstructural changes in different regions of processing maps and established boundary conditions for safe processing zones. The deformation mechanisms in safe zones include dynamic recovery, dynamic recrystallization, and superplasticity, while instability zones exhibit intergranular fracture, wedge cracking, ductile fracture of hard particles, and adiabatic shear band formation.

However, their safe zones only considered defect avoidance without microstructural optimization. Prasad suggested that for low stacking fault energy materials, recrystallization typically occurs at 0.7-0.8Tm (where Tm is the melting point) with strain rates of 0.1-1 s^{-1}, and the maximum energy dissipation efficiency ranges from 0.3-0.35.

Based on these considerations and microstructural analysis, the processing map at a true strain of 0.6 can be divided into four regions:

(1) Plastic Instability Region (Region 1): Shown as the shaded area in [Figure 4: see original paper], this region spans 950–1150°C at strain rates of 1-10 s^{-1}. [Figure 5: see original paper] shows microstructures after deformation in this region. At low deformation temperatures (950°C), instability manifests as localized plastic deformation and shear bands. The degree of localization increases with strain rate, slightly widening the shear bands. At higher temperatures, adiabatic heating during high strain rate deformation, combined with the low thermal conductivity of high speed steel, causes rapid temperature elevation in localized deformation bands, reducing their strength and promoting further deformation within these bands, ultimately forming adiabatic shear bands ([Figure 5c: see original paper]). At constant strain rate, the width of the plastically unstable region increases with temperature.

(2) Safe Processing Region (Region 2): Outlined by the green border in [Figure 4: see original paper], this region spans 1050–1150°C at strain rates of 0.01-0.1 s^{-1}, with a maximum power dissipation efficiency of 33%. Micrographs of specimens deformed in this region are shown in [Figure 6: see original paper]. The original microstructure is replaced by recrystallized grains significantly finer than the as-deposited grains, and the lamellar eutectic carbides distributed along grain boundaries in the as-deposited state are partially broken up. The primary softening mechanism in this region is dynamic recrystallization. Comparing [Figure 6a: see original paper] and [6b] reveals that recrystallized grain size increases with deformation temperature at constant strain rate. According to the relationship between grain size and flow stress:

$$D \approx 0.8 k_g \sigma_s^{-p}$$

where k_g and p are material constants, and σ_s is the steady-state flow stress. This equation indicates that grain size increases as flow stress decreases with temperature. Comparing [Figure 6b: see original paper] and [6c] shows that recrystallized grain size decreases with increasing strain rate at constant temperature. At lower strain rates, flow stress decreases, which according to the equation would increase grain size.

The primary objective of forging spray-formed high speed steel is to refine carbide size, morphology, and distribution while eliminating inherent microporosity. Carbide breakdown requires large deformations, which are typically performed in the recrystallization region. Based on microstructural analysis, the feasible hot working window for spray-formed Nb-containing M3:2 high speed steel is 1050–1150°C at 0.01-0.1 s^{-1}. To obtain an optimized microstructure with fine, uniformly distributed spherical carbides, the recommended hot working parameters are 1150°C and 0.1 s^{-1}.

(3) Low Temperature Region (Region 3): Dislocation slip remains an important deformation mechanism at elevated temperatures. Unlike room temperature deformation, dislocations can overcome obstacles through combined action of applied stress and thermal activation (primarily dislocation climb at high temperatures), enabling continued slip. However, dynamic analysis shows that deformation in the low-temperature region (950–1000°C) requires high activation energy (Q = 609 kJ/mol), resulting in limited thermal activation and difficulty for dislocations to overcome obstacles. Additionally, numerous undissolved carbides at low temperatures create significant resistance to dislocation slip, requiring higher applied stress to maintain deformation. High stress leads to massive dislocation multiplication and pile-up at grain boundaries, generating stress concentrations that nucleate microcracks, which propagate and link up during deformation ([Figure 7: see original paper]). [Figure 7b: see original paper] also shows that cracks can directly cut through hard carbide phases during propagation, indicating that cutting of second-phase particles by dislocations is a non-thermally activated process. Thus, thermal activation plays a relatively minor role in helping dislocations overcome obstacles in this region, and crack initiation is primarily caused by stress concentration from dislocation pile-up at grain boundaries.

(4) Low Strain Rate Region (Region 4): Outlined by the red border in [Figure 4: see original paper], microstructures deformed at 0.001 s^{-1} are shown in [Figure 8: see original paper]. Microcracks along grain boundaries are observed at all temperatures. Plastic deformation of polycrystalline materials requires grain boundary coordination. At high temperatures, grain boundary atoms diffuse easily, facilitating grain boundary sliding under applied stress. Therefore, grain boundary sliding must be considered during high-temperature deformation. However, in spray-formed high speed steel, numerous hard second-phase particles (primarily M2C eutectic carbides and MC carbides) distributed along grain boundaries impede grain boundary sliding, creating significant stress concentrations at particle-grain boundary interfaces. Additionally, slow dislocation accumulation at low strain rates reduces the driving force for dynamic recrystallization (the primary softening mechanism), resulting in insufficient stress relief. When stress concentration exceeds the bonding strength between second-phase particles and the matrix, particle debonding creates voids that grow and coalesce into cracks during subsequent deformation. [Figure 8b: see original paper] also shows wedge cracks formed at triple grain junctions due to hindered grain boundary sliding. In summary, cracks in this region primarily form from voids nucleated by grain boundary sliding and their subsequent growth.

2.5 Forged Microstructure and Mechanical Properties

The as-forged microstructure of the spray-formed Nb-containing M3:2 high speed steel (forging ratio 6.25) is shown in [Figure 9: see original paper]. The forged microstructure is uniform and fine, with carbides that are small, predominantly spherical or near-spherical, and uniformly distributed—representing significant improvement over the as-deposited state. [Figure 9b: see original paper] shows that carbides exhibit slight banding in the longitudinal direction, which is the main reason for the lower isotropy of spray-formed high speed steel compared with powder metallurgy high speed steel.

The mechanical properties of the forged Nb-containing M3:2 high speed steel after heat treatment are listed in . The hardness and bending strength of the spray-formed Nb-containing M3:2 high speed steel exceed those of sintered ASP23 high speed steel and are comparable to hot isostatically pressed (HIP) ASP23 powder metallurgy high speed steel, though its impact toughness is lower than that of HIP high speed steel.

Conclusions

1. The as-deposited microstructure of spray-formed Nb-containing M3:2 high speed steel requires subsequent thermomechanical processing to eliminate defects and improve carbide size, morphology, and distribution.

2. Flow stress decreases with increasing deformation temperature and decreasing strain rate, indicating positive strain rate sensitivity and negative temperature sensitivity for spray-formed Nb-containing M3:2 high speed steel.

3. In the low-temperature deformation region (950-1000°C, 0.001-0.1 s^{-1}), thermal activation plays a weak role in helping dislocations overcome obstacles, and cracks initiate due to stress concentration from dislocation pile-up at grain boundaries. In the low strain rate region (0.001 s^{-1}), cracks form primarily through nucleation and growth of voids caused by grain boundary sliding being impeded by grain boundary carbides.

4. The hot processing window for spray-formed Nb-containing M3:2 high speed steel is 1050-1150°C at strain rates of 0.01-0.1 s^{-1}, where dynamic recrystallization is the primary softening mechanism. Plastic instability occurs when the strain rate exceeds 1 s^{-1} in the temperature range of 950-1150°C.

5. After forging and heat treatment, the hardness and strength of Nb-containing M3:2 high speed steel meet or exceed those of hot isostatically pressed powder metallurgy high speed steel of the same composition.

References

1. SONG Xiaolong, AN Jiru, *New Handbook of Metal Materials at Home and Abroad*, 2nd Edition, (Beijing, Chemical Industry Press, 2012)
2. Catalogue Thyssen Edelstahl Service GmbH, GER Pat, G0168/1(1997)
3. P. Beiss, PM (Powder Metallurgy) methods for the production of high speed steels, *Met Powder Rep.*, 38(4), 185(1983)
4. I. C. Ernst, D. Duh, ESP4 and TSP4, a comparison of spray formed with powder metallurgically produced cobalt free high-speed steel of type 6W-5Mo-4V-4Cr, *Journal of Materials Science*, 39(22), 6831(2004)
5. LI Junchen, PENG Xiaodong, LIU Junwei, YANG Yan, ZENG Li, Deformation behavior of alloy Mg-9Li-3Al-2.5Sr at elevated temperature, *Chinese Journal of Materials Research*, 26(3), 309(2012)
6. K. P. Rao, Y. V. R. H. Prasad, K. Suresh, Anisotropy of flow during isothermal forging of rolled AZ31B magnesium alloy rolled plate in three orthogonal directions: Correlation with processing maps, *Mater. Sci. Eng.*, A558, 30(2012)
7. K. Suresh, K. P. Rao, Y. V. R. H. Prasad, N. Hort, K. U. Kainer, Effect of calcium addition on the hot working behavior of as-cast AZ31 magnesium alloy, *Mater. Sci. Eng.*, A588, 272(2013)
8. C. Dharmendra, K. P. Rao, F. Zhao, Y. V. R. H. Prasad, N. Hort, K. U. Kainer, Effect of silicon content on hot working, processing maps, and microstructural evolution of cast TX32-0.4 Al magnesium alloy, *Mater. Sci. Eng.*, A606, 11(2014)
9. O. Sivakesavam, Y. V. R. H. Prasad, Hot deformation behaviour of as-cast Mg-2Zn-1Mn alloy in compression: a study with processing map, *Mater. Sci. Eng.*, A362, 118(2003)
10. N. Srinivasan, Y. V. R. H. Prasad, P. P. Rao, Hot deformation behaviour of Mg-3Al alloy, a study using processing map, *Mater. Sci. Eng.*, A476, 146(2008)
11. T. Zhong, K. P. Rao, Y. V. R. H. Prasad, F. Zhao, M. Gupta, Hot deformation mechanisms, microstructure and texture evolution in extruded AZ31-nano-alumina composite, *Mater. Sci. Eng.*, A589, 41(2014)
12. KONG Fantao, CUI Ning, CHEN Yuyong, XIONG Ningning, The hot deformation behavior of Ti-43Al-9V-Y alloy, *Acta Metall. Sin.*, 49(11), 1363(2012)
13. T. Seshacharyulu, S.C. Medeiros, J.T. Morgan, J.C. Malas, W.G. Frazier, Y.V.R.H. Prasad, Hot deformation and microstructural damage mechanisms in extra-low interstitial (ELI) grade Ti-6Al-4V, *Mater. Sci. Eng.*, A279(1), 289(2000)

14. F. W. Kang, G. Q. Zhang, Z. Li, J. F. Sun, Hot deformation of spray formed nickel-base superalloy using processing maps, *Trans. Nonferrous Met. Soc. China*, 18(3), 531(2008)

15. H. R. Ezatpour, S. A. Sajjadi, M. Haddad-sabzevar, G. R. Ebrahimi, Hot deformation and processing maps of K310 cold work tool steel, *Mater. Sci. Eng.*, A550, 152(2012)

16. S. Venugopal, S.L. Mannan, Y. V. R. H. Prasad, Optimization of hot workability in stainless, *Metall. Trans.*, 23A, 3039(1992)

17. G. Meng, B. L. Li, H. M. Li, H. Huang, Z. R. Nie, Hot deformation and processing maps of an Al-5.7 wt.%Mg alloy with erbium, *Mater. Sci. Eng.*, A517, 132(2009)

18. WANG Hebin, ZHANG Jinxiang, LU Lin, HOU Longgang, CUI Hua, HUANG Jinfeng, ZHANG Jishan, High-temperature thermal deformation and microstructure evolution of spray formed M4 high speed steel, *Chinese Journal of Materials Research*, 27(2), 167(2013)

19. Y. H. Liu, Y. Q. Ning, Z. K. Yao, M.W. Fu, Hot deformation behavior of the 1.15C-4.00Cr-3.00V-6.00W-5.00Mo powder metallurgy high speed steel, *Mater. Des.*, 54, 854(2014)

20. Y. V. R. H. Prasad, H. L. Gegel, S. M. Doraivelu, J. C Malas, J. T. Morgan, K. A. Lark, D. R. Barker, Modeling of dynamic material behavior in hot deformation: forging of Ti-6242, *Metall. Trans. A*, 15, 1883(1984)

21. Y. V. R. H. Prasad, Recent advances in the science of mechanical processing, *Indian J. Technol.*, 28, 435(1990)

22. H. J. McQueen, S. Yue, N. D. Ryan, E. Fry, Hot working characteristics of steels in austenitic state, *J. Mater. Process Technol.*, 53, 293(1995)

23. H. J. McQueen, N. D. Ryan, Constitutive analysis in hot working, *Mater. Sci. Eng.*, A322, 43(2001)

24. R. Raj, Development of a processing map for use in warm-forming and hot-forming processes, *Metall. Trans. A*, 12, 1089(1981)

25. C. Gandhi, On fracture initiation mechanisms and dynamic recrystallization during hot deformation of pure nickel, *Metall. Trans. A*, 13, 1233(1982)

26. Y. V. R. K. Prasad, T. Seshacharyulu, Modelling of hot deformation for microstructural control, *Int. Mater. Rev.*, 43(6), 243(1998)

27. Y. V. R. H. Prasad, K. P. Rao, S. Sasidhara, *Hot Working Guide: A Compendium of Processing Maps*, (Ohio, ASM International, 1997)

28. YU Yipeng, *Study on Microstructures and Properties of Spray Formed Nb-containing M3 High Speed Steel*, PhD thesis, University of Science and Technology Beijing, 2012

29. P. Beiss, Comparison of properties of high speed steel produced by conventional casting, spray forming and powder metallurgy, *Mater. Sci. Forum*, 498, 244(2005)

Note: Figure translations are in progress. See original paper for figures.