UDK 691.714:539.374
Original scientific article/Izvirni znanstveni članek
ISSN 1580-2949
Mater. Tehnol.

IN-SITU TEM STUDIES ON PLASTIC DEFORMATION MECHANISMS IN SLM AUSTENITIC STAINLESS STEEL

IN-SITU TEM ŠTUDIJA MEHANIZMOV PLASTIČNE DEFORMACIJE V SLM AUSTENITNEM NERJAVNEM JEKLU

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Prejem rokopisa – received: 2025-06-24; sprejem za objavo – accepted for publication: 2025-10-07

doi:10.17222/mit.2025.1498

This paper reviews the research on the plastic deformation mechanisms of selective laser melting (SLM) austenitic stainless steel using in-situ transmission electron microscopy (TEM). The SLM technology, which manufactures 3D metal components with ultra-fast cooling rates, demonstrates unique plastic deformation mechanisms, including coordinated effects of dislocation slip, twinning, and phase transformation. In-situ TEM tensile experiments visually reveal these dynamic processes: dislocation cell walls replacing grain boundaries as the primary dislocation sources, "slip-step" characteristics of dislocations at cell walls, and the layer-by-layer growth mechanism of nano-twins. These processes significantly enhance the material's strength-plasticity synergy. At low temperatures, deformation-induced martensitic transformation (DIMT) occurs via the $\gamma \rightarrow \varepsilon \rightarrow \alpha'$ path. Phase nucleation is closely related to dislocation accumulation and the intersection of ε-plates, with in-situ TEM capturing the dynamic process of phase transformation in real-time. This research indicates that stacking fault energy (SFE) is a key factor influencing phase transformation. The unique delayed DIMT phenomenon in SLM austenitic stainless steel is attributed to the high dislocation density, which hinders the movement of partial dislocations, significantly enhancing the material's ductility. The in-situ TEM technique provides key evidence for revealing the plastic deformation mechanisms of advanced metals and offers guidance for high-performance material design.

Keywords: in-situ TEM, SLM austenitic stainless steel, dislocation slip, twinning, phase transformation

V članku avtorji podajajo literaturni pregled raziskav s področja mehanizmov plastične deformacije selektivnega laserskega nataljevanja (SLM; angl.: selective laser melting) austenitnih nerjavnih jekel s poudarkom na *in-situ* presevni elektronski mikroskopiji (TEM). SLM technologija s katero izdelujemo kovinske izdelke zahtevnih 3D oblik poteka pri ultra-hitrem ohlajanju z edinstvenimi mehanizmi plastične deformacije, ki vključujejo koordinirano drsenje dislokacij, dvojčenje in fazno transformacijo. Z *in-situ* nateznimi TEM eksperimenti lahko vizualno odkrijemo in prepoznamo te dinamične procese, kot so: nadomeščanje mej zrn (kot primarni izvor dislokacij) s celičnimi stenami dislokacij, značilnosti stopničastega drsenja dislokacij (angl.: "slip-step" characteristics of dislocations) na celičnih stenah in postopen mehanizem rasti dvojčkov plast za plastjo. Vsi ti procesi pomembno izboljšajo sinergijo med trdnostjo in plastičnostjo materiala. Pri zelo nizkih temperaturah poteka deformacijsko inducirana martenzitna transformacija (DIMT; angl.: deformation induced martensitic transformation) preko poti $\gamma \rightarrow \varepsilon \rightarrow \alpha'$. Fazna transformacija je tesno povezana s kopičenjem dislokacij in presečišči ε-ploščic. Z *in-situ* TEM nateznimi eksperimenti ustvarimo in zajamemo proces fazne transformacije v realnem času. V eni od raziskav avtorji povdarjajo, da je energija napake zloga (SFE; angl.: stacking fault energy) ključni faktor, ki vpliva na fazno transformacijo. Edinstven fenomen zakasnjene DIMT v SLM austenitnih nerjavnih jeklih raziskovalci pripisujejo visoki gostoti dislokacij, ki zavira gibanje parcialnih dislokacij, kar pomembno izboljšuje duktilnost materiala. Tehnika *in-situ* TEM lahko tako zagotavlja ključne dokaze za odkrivanje mehanizmov plastične deformacije kovin in ponuja smernice za načrtovanje naprednih visokozmogljivih materialov.

Ključne besede: *in-situ* natezni preizkusi, presevna elektronska mikroskopija, selektivno lasersko nataljevanje, austenitno nerjavno jeklo, drsenje dislokacij, dvojčenje, fazna transformacija

1 INTRODUCTION

With the growing demand for personalized and highly customized austenitic stainless steel on the markets, aerospace, energy, national defense and other key industries are actively seeking new technologies to replace traditional processes in order to achieve higher levels of process flexibility and versatility.[1–5] For decades, additive manufacturing (AM) has gradually become the pioneering force driving this wave of transformation. Selective laser melting (SLM), as a key method in the field of additive manufacturing (AM), enables direct fabrication of complex components from digital models, greatly enhancing the freedom of customized design and production. Moreover, compared to traditional methods, it significantly simplifies the manufacturing process, improves material utilization, and reduces production costs,[3,4,6–8] thereby exerting a profound impact on the development of the manufacturing industry.[9–11] SLM is characterized by ultra-fast cooling rates. Traditional rapid cooling technologies are limited to producing low-dimensional metal forms, such as metal powders,

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strips, and foils, whereas additive manufacturing can directly produce 3D metal components with extremely high cooling rates,[12,13] as shown in **Figure 1**. When manufacturing austenitic stainless steel through SLM, the laser beam precisely scans each layer of powder, causing the powder particles to fuse directly, layer by layer, to create a 3D component.[14] The austenitic stainless steel produced in this way has a microstructure that is significantly different from products made through traditional manufacturing methods (such as casting and forging), exhibiting unique and exceptional performance characteristics.

The application in fields such as deep-sea exploration, oil and gas storage and transportation urgently requires research into the source of the excellent mechanical properties of SLM austenitic stainless steel.[15] It is currently known that the excellent mechanical properties of SLM austenitic stainless steel are typically attributed to its unique deformation mechanisms, which arise from its characteristic heterogeneous structure and cellular substructures.[16–18] Furthermore, the deformation mechanisms of SLM austenitic stainless steel can be influenced by adjusting the elemental composition and microstructure.[19–23] Therefore, revealing the formation processes and interactions of dislocations, nano-twins, and martensitic phase transformations is crucial for the design of high-performance materials and process optimization, which further increases the demand for microstructural characterization techniques.[24] As a result, advanced micro-experimental characterization technologies that can visually display the dynamic deformation processes of materials are increasingly becoming key tools in the related research.

The in-situ transmission electron microscopy (TEM) tensile technique employs a straining stage as an effective tool for exploring the deformation mechanisms of materials, characterized by its intuitive observation capability and high efficiency. In recent years, it has been proven to be highly effective in revealing the dynamic deformation processes of SLM materials.[25–27] With the help of in-situ transmission electron microscopy (TEM) used for direct observation at the micro and atomic scales,[28–31] we are able to gain insights into the subtle changes in a crystal structure under stress-strain conditions, including dynamic processes such as dislocation generation and movement. This provides the possibility to visually reveal the intrinsic mechanisms of materials during elastic-to-plastic deformation stages, thereby deepening the understanding and recognition of the material's inherent properties.[32] Most importantly, in-situ mechanical studies can eliminate the interference caused by changing samples in ex-situ experiments during the same deformation process, which is crucial for exploring new potential deformation mechanisms. The source of the excellent mechanical properties of SLMed austenitic stainless steel indeed requires complex in-situ characterization techniques to be revealed.[33,34] This paper reviews the recent research on the plastic deformation of SLMed austenitic stainless steel using in-situ transmission electron microscopy, with a detailed discussion of the deformation mechanisms from the perspectives of dislocation slips, twinning, and phase transformation. Finally, the paper explores potential future research directions for the transmission electron microscopy studies of SLM austenitic stainless steel.

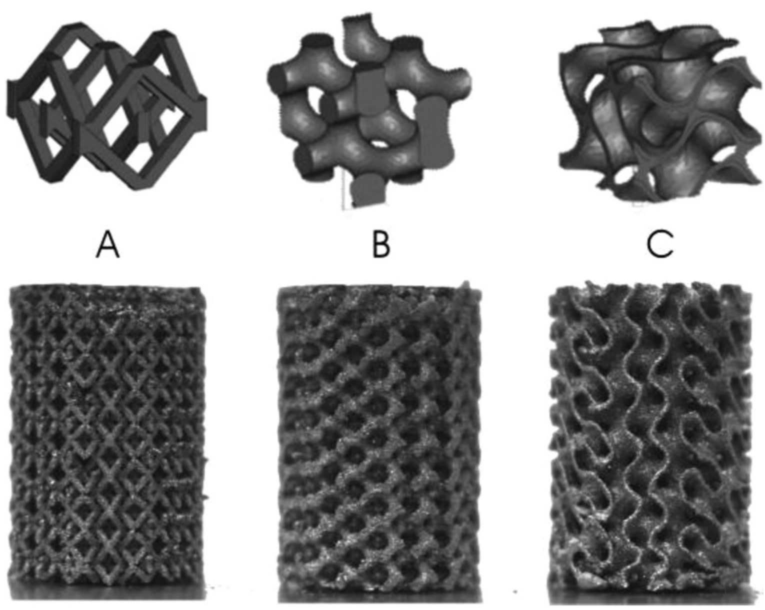

Figure 1: CAD images of designed unit cells and their corresponding as-produced components: (A) octahedron (Ock), (B) cellular gyroid (CG), (C) sheet gyroid (SG)[6]

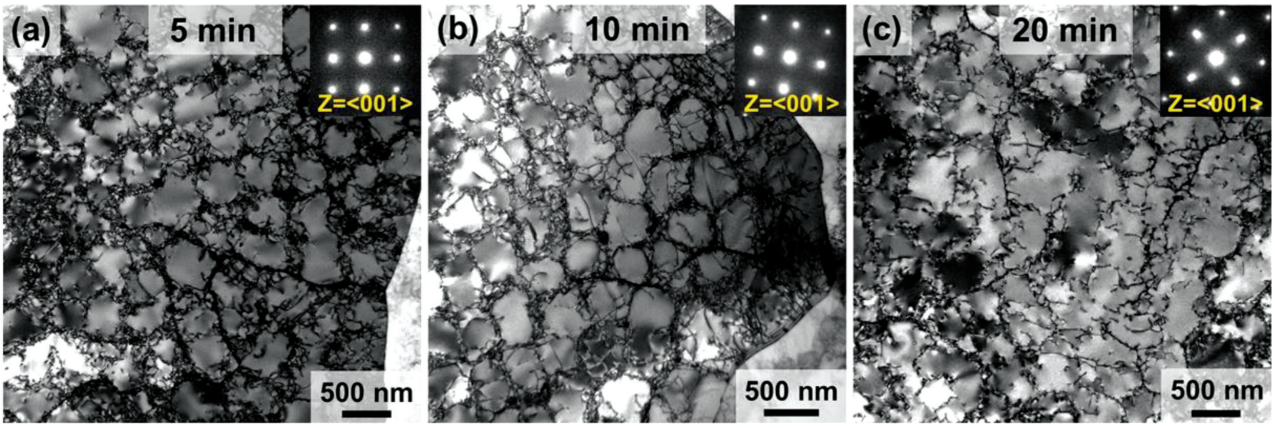

Figure 2: Bright-field TEM images of dislocation-cell structure in SLM-316L sample[52]

2 DISLOCATION-BASED PLASTICITY MECHANISMS

The high strength of SLM-manufactured metal materials is primarily attributed to the formation of cellular substructures, increased dislocation density, and enhanced dislocation activity. The differences in the dislocation-network distribution lead to variations in mechanical properties. Therefore, understanding the plastic mechanisms of dislocations – specifically how they nucleate, proliferate, and interact with substructures or other defect structures – is key to comprehending the mechanical properties of SLM-ed metals.[35–44] In-situ transmission electron microscopy (TEM) mechanical testing provides an opportunity to directly capture dynamic dislocation nucleation and slip processes in small-sized SLM-ed samples, deepening the fundamental understanding of the underlying deformation and strengthening mechanisms of SLM-ed metals. This, in turn, reveals the fundamental reason why SLM-ed metals exhibit superior strength-plasticity synergy compared to samples made with other processing methods.[45–51]

Due to the laser-powder interaction during an SLM forming process, the micro-melt pool undergoes cyclic heating and cooling. Periodic thermal expansion and contraction lead to the formation of subcellular struc-

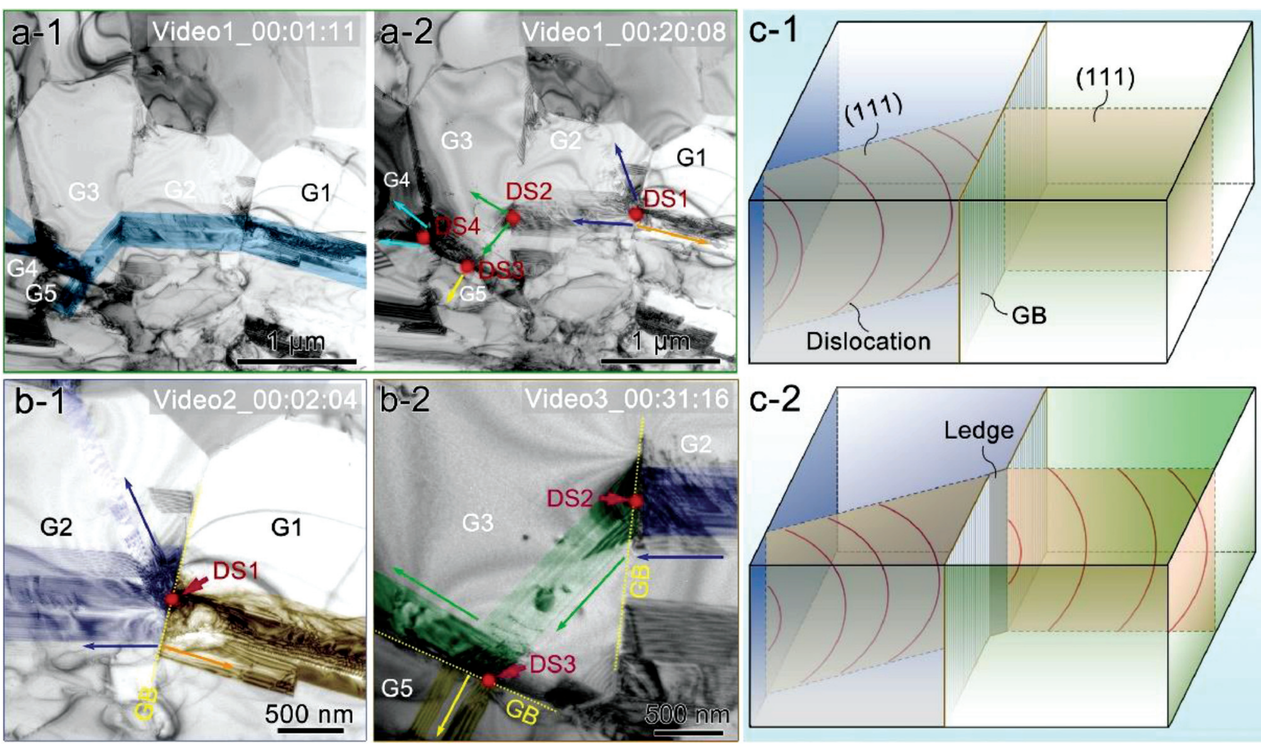

Figure 3: In-situ TEM observation of plastic deformation propagating within grains: (a-1) and (a-2) in-situ TEM images showing the connected slip paths within grains; (b-1) and (b-2) screenshots showing the emission of dislocations from GB DSs; (c-1) and (c-2) schematics showing the formation of boundary DS[49]

tures within the grains, also known as dislocation cells, shown in **Figure 2**.[52] The dislocation walls of these dislocation cells provide nucleation sites for dislocations and can significantly, though not completely, hinder dislocation movement. The fine dislocation cells clearly lead to high strength, similar to the Hall-Petch relationship.[48] Gao believes that when plastic deformation begins, Frank-Read dislocation sources form on dislocation walls and emit dislocations into adjacent subcellular structures. This behavior is similar to that in FG 316L stainless steel; however, in FG 316L stainless steel, grain boundaries play a key role in accommodating plastic deformation and are the primary dislocation source for dislocation emission into the grain interior. The expansion of plastic deformation within the grains occurs through continuous activation of grain boundary dislocation sources. In contrast, for SLMed 316L stainless steel, dislocation walls replace grain boundaries as the primary dislocation source, as shown in **Figure 3**.[49]

Figure 4 clearly shows that when plastic deformation continues, dislocations are temporarily trapped when they reach dislocation walls, and then they dissociate into Shockley partials with a wiggling motion. As the applied stress increases, the dislocations move forward again, as illustrated above. When the external stress is sufficiently high, the leading portion is emitted from dislocation wall A and stops at the nearby dislocation wall B. At this point, the trailing portion remains trapped by dislocation wall A. As the applied stress gradually increases, the leading portion overcomes the obstacle posed by wall B and instantaneously jitter-slides into the adjacent cell. The trailing portion also slides onto wall B.[50] An interesting feature of the dislocation slip here is that the slip motion is not continuous as it exhibits an abrupt, instantaneous behavior. Lee's research describes this type of dislocation movement, characterized by a jittering slip, which was also observed through in-situ TEM deformation experiments in nano-precipitate-hardened and irradiated austenitic steels. This slip behavior may be closely related to the high lattice-friction stress[51]. Additionally, in general, in FCC metals with low stacking fault energy, due to the wider stacking faults, dissociated dislocations have difficulty in cross-slip. As a result, perfect dislocations tend to maintain planar slip or dissociate into partial dislocations[53]. Thus, after dislocation A crosses a dislocation wall, it dissociates into partial dislocations with Burgers vectors of 1/6[211 and 1/6[12–1], respectively. Clearly, in SLMed 316Lstainless steel, dislocation motion is obstructed by these dislocation walls but it does not stop completely. As the strain increases, slip transfers between adjacent subcellular structures. Therefore, the SLM process enables austenitic stainless steel to improve its strength without sacrificing ductility.

In summary, in-situ transmission-electron-microscopy tensile experiments provide an accurate and intuitive validation method for understanding the dislocation slip mechanisms in metals such as 316L stainless steel. These experiments enable real-time capture of disloca-

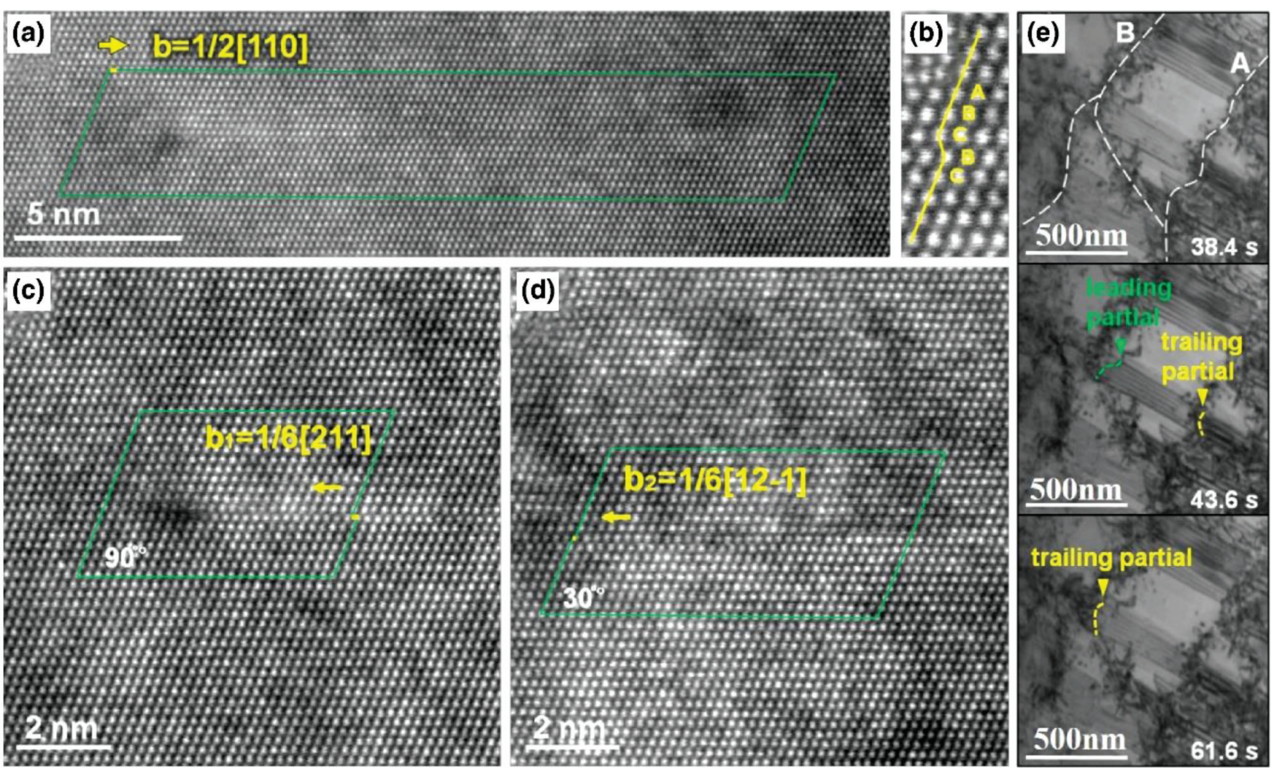

Figure 4: In-situ TEM images showing dynamic evolution of SLM 316L sample: (a–d) formation process of partial dislocations; (e) interaction between dislocations and dislocation walls[50]

tion nucleation, proliferation, and their dynamic interactions with microstructures at the nanoscale, providing key insights into the fundamental nature of their excellent strength-plasticity synergy. The key findings include: (1) Dislocation cell walls replace traditional grain boundaries as the primary dislocation source, which is a fundamental difference from the deformation mechanisms of forged materials; (2) The interaction between partial dislocations and cell walls exhibits a unique "jumping" slip behavior, characterized by a sudden motion after dislocations are temporarily pinned; (3) Strain is transmitted across cells via dislocation walls, and this restricted dislocation movement enables strengthening while maintaining the material's ductility. The real-time analysis of nanoscale dynamic changes enabled by in-situ TEM has made it a hallmark tool for characterizing the microstructural evolution of advanced metallic materials.

3 TWINNING-BASED PLASTICITY MECHANISMS

It is well known that SLMed 316L stainless steel typically exhibits a relatively high porosity. However, during 3D printing, nitrogen can reduce the stacking fault energy, promoting the formation of deformation twins, resulting in a behavior similar to twinning induced plasticity (TWIP). In this process, twinning generates significant plasticity, compensating for the negative effects of porosity and enabling the material to exhibit excellent strength and ductility.[48,53] Therefore, studying the formation conditions and proliferation mechanisms of deformation twins is of great importance for mitigating the effects of porosity and enhancing the strength-ductility synergy of SLMed 316L stainless steel.

Generally, the main factor influencing the formation of deformation twins is the stacking fault energy (SFE). Within a certain range, the lower the SFE, the more favorable the formation of deformation twins, and the thinner the twins formed as the SFE decreases. Unlike the decrease in the stacking fault energy caused by high concentrations of Mn in TWIP steels, in the SLM process, nitrogen (N) as a solute atom can diffuse into the stacking fault region, thereby significantly reducing the SFE of SLMed 316L stainless steel. This promotes dislocation dissociation and stacking fault formation, limits dislocation cross-slip, and facilitates the formation of deformation twins.[48,53] From another perspective, the formation of deformation twins in FCC metals is caused by the highly coordinated sliding of Shockley partial dislocations with specific Burgers vectors on consecutive {111} twin planes.[55] SLMed 316L stainless steel is no exception; partial dislocations emitted from dislocation walls or grain boundaries, when sliding on adjacent planes, can form deformation twins.[50,56] This behavior is similar to that in TWIP steels.

Figure 5 includes snapshots from in-situ TEM strain testing illustrating the process of direct emission of Shockley partial dislocations from a grain boundary source, leading to the formation of deformation twins (DTs). Due to the motion of dislocation A, the contrast at the lower part of planar defect 1 changes from fringe contrast to white contrast. As deformation progresses, the contrast at the upper part of the planar defect also changes from fringe contrast to white contrast through the motion of dislocation B. The contrast change from fringe to white corresponds to the formation of a thin twin crystal through the motion of a single twinning partial dislocation, with a thickness that is a multiple of three atomic layers.[54] Meanwhile, as can be seen in **Figure 6**, in-situ TEM strain testing confirmed that SLMed 316L stainless steel can form nanoscale twins after deformation. Nanoscale twins can also accumulate and grow from dislocation cell walls, rather than necessarily starting from grain boundaries. The thickness of the twins typically ranges from 2 nm to 6 nm, with stable twin thicknesses as small as two atomic layers, reflecting the layer-by-layer growth mechanism of the twins.[50] The occurrence of nanoscale twins leads to an increase in the lattice friction stress, which in turn increases the resistance to dislocation motion. Through a dynamic Hall-Petch effect, similar to that in nanostructured twinned copper and TWIP steels, stable plastic deformation can be achieved through strain hardening.

In summary, in SLMed 316L stainless steel, twinning growth exhibits a unique plastic coordination mechanism. In-situ TEM experiments visually reveal the dynamic formation process of deformation twins: through

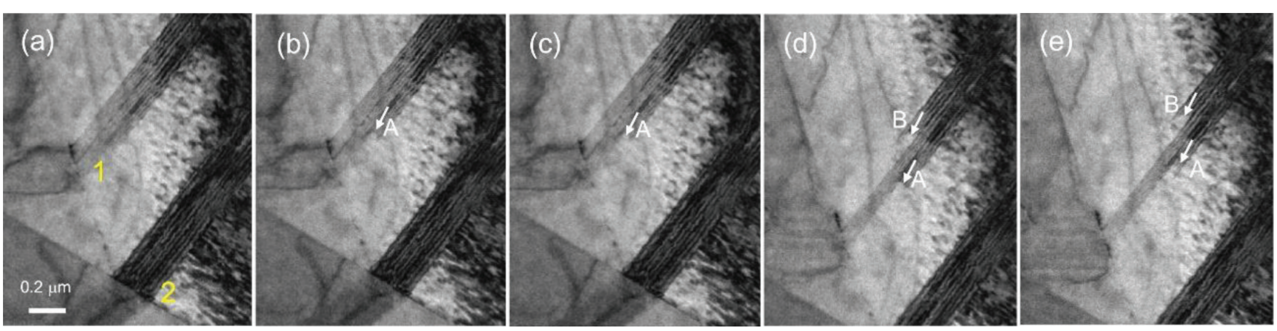

Figure 5: Snapshots of an in-situ TEM straining test providing evidence of the formation of DTs by direct emission of Shockley partial dislocations from grain boundary sources[54]

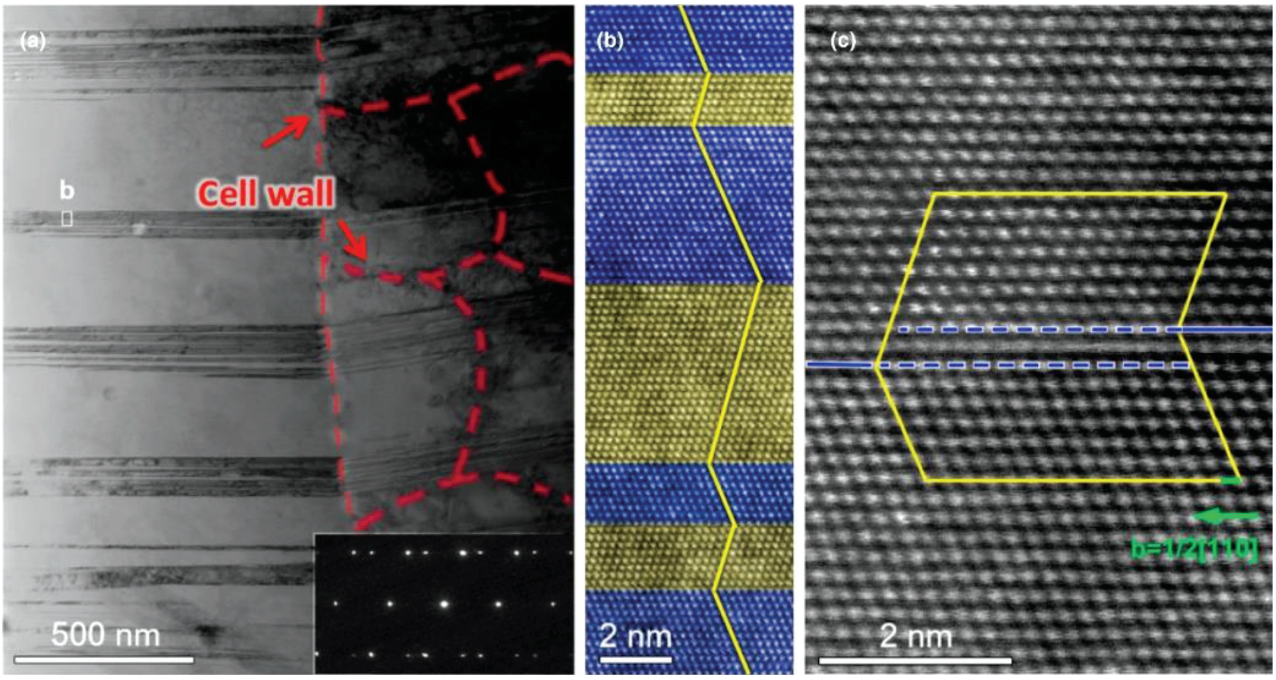

Figure 6: TEM image showing nano-twins formed during deformation of SLM 316L sample[50]

the highly coordinated sliding of Shockley partial dislocations on the {111} twin planes, twins nucleate from grain boundaries or dislocation cell walls and exhibit nanoscale characteristics of layer-by-layer growth. It is notable that the nitrogen atoms incorporated during the SLM process significantly reduce the stacking fault energy, promoting dislocation dissociation and twin formation. This twinning induced plasticity (TWIP) effect effectively compensates for the negative impact of porosity defects on the material's strength-ductility synergy. Additionally, in-situ observations reveal that nanoscale twins can not only nucleate from grain boundaries but also accumulate and grow from dislocation cell walls. Their thickness can range from 2 nm to 6 nm (even as thin as two atomic layers), significantly increasing lattice friction stress. This dynamic Hall-Petch effect impedes dislocation motion through twin boundaries, leading to continuous strain hardening during deformation. This provides microstructural support for the excellent strength-ductility synergy of SLMed 316L stainless steel. These findings not only deepen the understanding of twinning plasticity mechanisms but also highlight the unique advantages of the in-situ TEM technology in analyzing the deformation behavior of advanced metals.

4 PHASE TRANSFORMATION-MEDIATED PLASTIC DEFORMATION

In recent years, the urgent demand for highly customized austenitic stainless steel in key application areas such as liquefied natural gas storage and transportation, as well as components for space exploration, has driven research advancements on the plastic deformation mechanisms of selective laser melted austenitic stainless steel used in special environments. For example, under low-temperature conditions, the plastic deformation mechanism shifts from being predominantly governed by ordinary dislocation slip to being dominated by martensitic transformation. As shown in **Figure 7**,[57] recent studies have indicated the phenomenon of deformation-induced martensitic transformation (DIMT) in SLMed 316 stainless steel at low temperatures. Specifically, external strain or stress introduced by deformation can drive this phase transformation. The deformation-induced phase transformation typically follows two pathways: face-centered cubic (FCC) γ-austenite → hexagonal close-packed (HCP) ε-martensite → body-centered cubic (BCC) α'-martensite, and γ-austenite → α'-martensite. The occurrence of phase transformation can lead to fundamental changes in the original structure. In addition to enhancing the solid solution, it can accommodate a more plastic strain, thereby increasing strength without altering the ductility of the metal.[57,58] In-situ transmission electron microscopy (TEM), as a small-scale micro-characterization technique, can easily capture the martensitic transformation process at the atomic scale, providing valuable insights for a more comprehensive understanding of its transformation mechanisms.

As a diffusionless (i.e., displacive) transformation, the lattice reorganization of martensitic transformation is achieved through an ordered and short-range migration of atoms within the matrix. Current research indicates that the stacking fault energy (SFE) is the primary factor influencing the martensitic transformation. Whether it is the element content, temperature, or strain rate, all these factors indirectly affect the martensitic transformation by

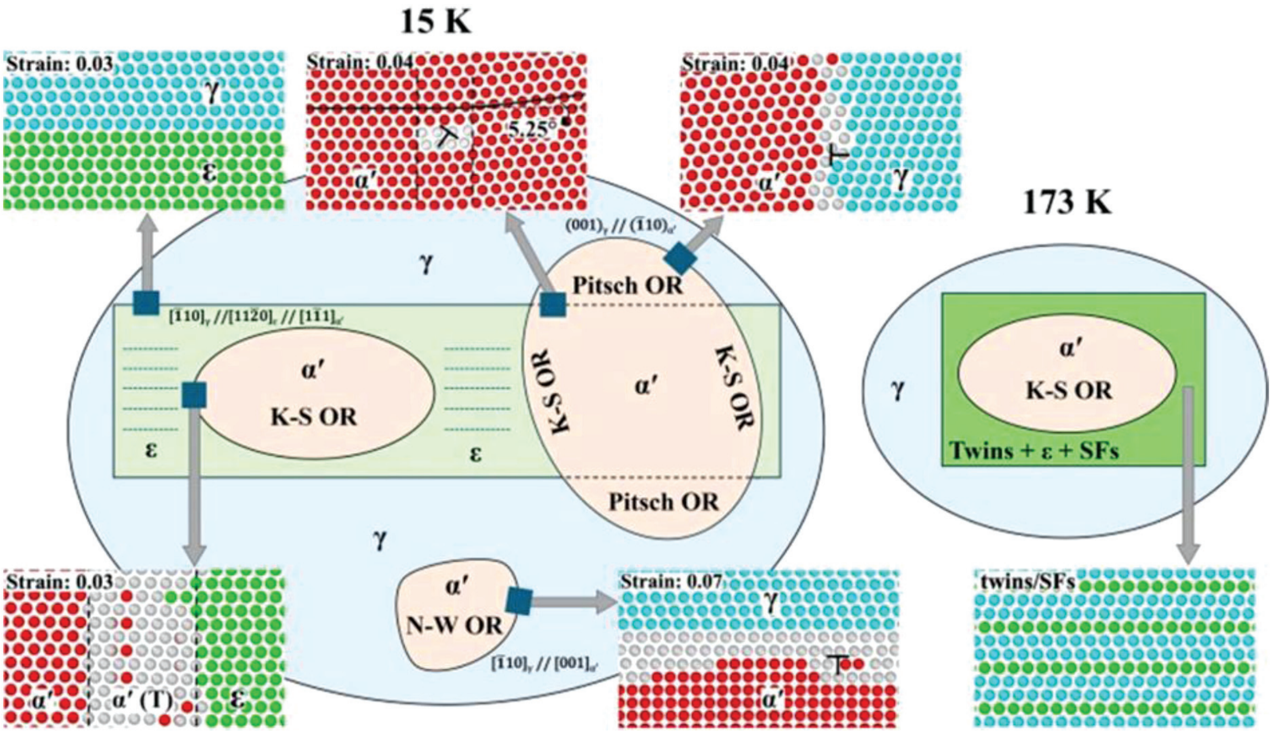

Figure 7: Schematic diagrams of martensitic transformation pathways induced by ultra-cryogenic deformation at 15 and 173 K[57]

altering the material's SFE. As revealed in **Figure 8**, already several years ago, Brooks and others conducted in-situ observations of a martensite formation in stainless steel. Their results indicated that ε-martensite typically appears in areas where appropriate but irregular stacking faults are formed, while the nucleation of α'-martensite is associated with dislocation accumulation on the slip planes. Bracke et al. discovered that α'-martensite nucleates at the intersections of two ε-martensite plates and conducted detailed observations of the crystal structure in this region using transmission electron microscopy (TEM), further confirming that the ε-phase serves as an intermediate phase for the nucleation of α'-martensite in an austenitic matrix.[59] As shown in **Figure 9**, recent work by J. Liu using a Gatan cooling straining TEM stage confirmed the deformation-induced martensitic phase transformation in austenitic stainless steel at low temperatures.[60] This study directly captured the formation process of stacking faults and the subsequent ε-martensite, visually demonstrating the role of stacking faults (SFs) as an intermediate step in the transformation from γ-austenite to ε-martensite. Specifically, when the stacking faults overlap periodically on every other {111}

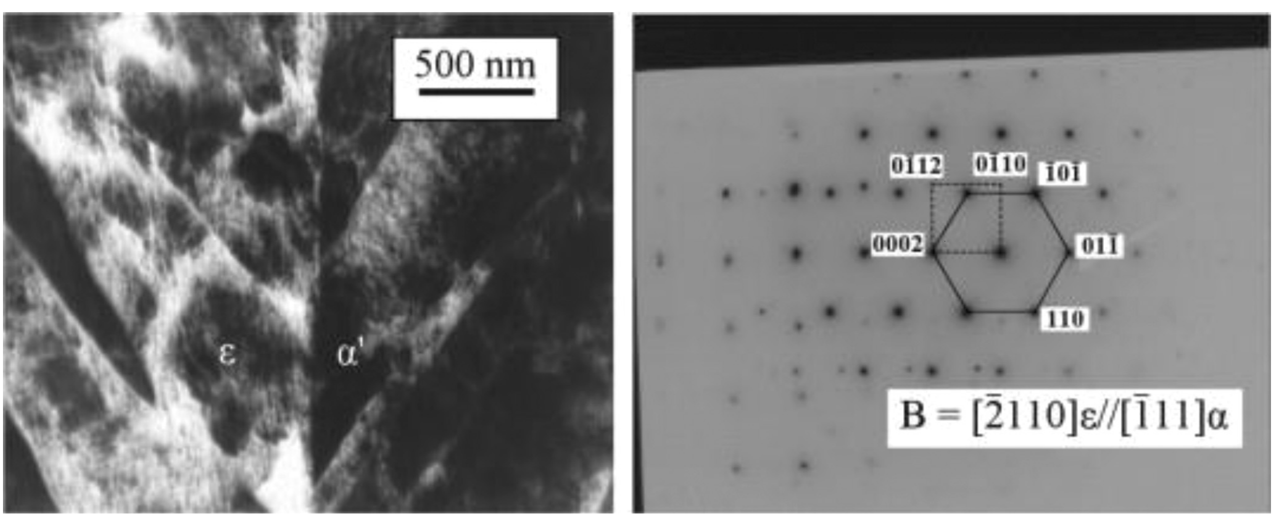

Figure 8: Bright-field TEM image and its corresponding DP of the tensile sample in-situ strained to 5 % at room temperature[59]

plane, a complete ε-martensite phase with a close-packed hexagonal (hcp) crystal structure is formed.[60]

In addition, Ye Tian and others controlled the stacking fault energy (SFE) of austenitic stainless steel by adjusting the element content. As the SFE increased, the transformation mechanism shifted from $\gamma \rightarrow \varepsilon \rightarrow \alpha'$ to $\gamma \rightarrow \gamma^{twin} \rightarrow \alpha'$, and ultimately resulted in only mechanical twins and stacking faults, with almost no α'-martensite present. This indicates that as the SFE continues to increase, the austenite gradually transitions from an unstable state to a stable state, leading to a decrease in the fraction of ε-martensite, while α' typically nucleates along the ε phase, resulting in a gradual reduction of martensite.[61] This is also validated by Suning Li and others through molecular dynamics (MD) simulations and high-resolution transmission electron microscopy observations. The study found that during the transition from twin boundaries and stacking faults to α'-martensite, higher local strain levels were observed compared to those associated with ε-martensite. The redistribution of stress leads to localized strain concentration in the surrounding regions. This strain concentration effect creates a higher energy local environment for the formation of α', reducing the energy barrier for martensite nucleation and promoting the formation of the new α' phase, thereby allowing for strain energy relaxation. Furthermore, higher levels of local strain can also facilitate the formation of α' by providing a distorted lattice, which is critical for the phase transformation.[59]

It is noteworthy that a unique phenomenon of delayed deformation-induced martensitic transformation (DIMT) occurs in SLMed 316L stainless steel. The delayed initiation and slower DIMT rate cause the peak of the work-hardening-rate curve to occur later, resulting in a slowdown of the transformation rate of tough austenite, which can accommodate more plastic deformation. This delayed DIMT can be attributed to the higher dislocation density and elevated nitrogen content, with the dislocation density playing a primary role. A higher dislocation density requires additional energy for the movement of a/6 p Shockley partial dislocations to bypass existing dislocation entanglements within the material. Such high energy demands significantly restrict local dislocation motion, thereby stabilizing austenite and delaying its transformation to martensite, which ultimately results in improved ductility of the material.[62] However, due to the limitations in observation time scales and loading conditions of transmission electron microscopy and in-situ mechanical testing, some transient phase transformations or changes in stacking fault structures may be challenging to dynamically capture using in-situ techniques. Research in these areas may need to rely on future advancements in experimental technologies. Given that SLM austenitic stainless steel exhibits excellent mechanical properties that can meet the specific requirements of oil and gas storage and high-end applications, studying its phase transformation mechanisms under special conditions is the key issue for understanding and controlling the mechanical properties of SLM austenitic stainless steel.

In summary, the phase transformation in SLMed 316L stainless steel exhibits a unique plastic coordination mechanism. Under low-temperature conditions, the plastic deformation mechanism shifts from being dominated by dislocation slip to being dominated by martensitic transformation. This deformation-induced martensitic transformation (DIMT) occurs via pathways $\gamma \rightarrow \varepsilon \rightarrow \alpha'$ or $\gamma \rightarrow \alpha'$. In-situ TEM experiments visually reveal this dynamic phase transformation process: martensite forms through the ordered migration of atoms within the matrix, and its nucleation is closely related to dislocation accumulation and the crossing of ε-laths.

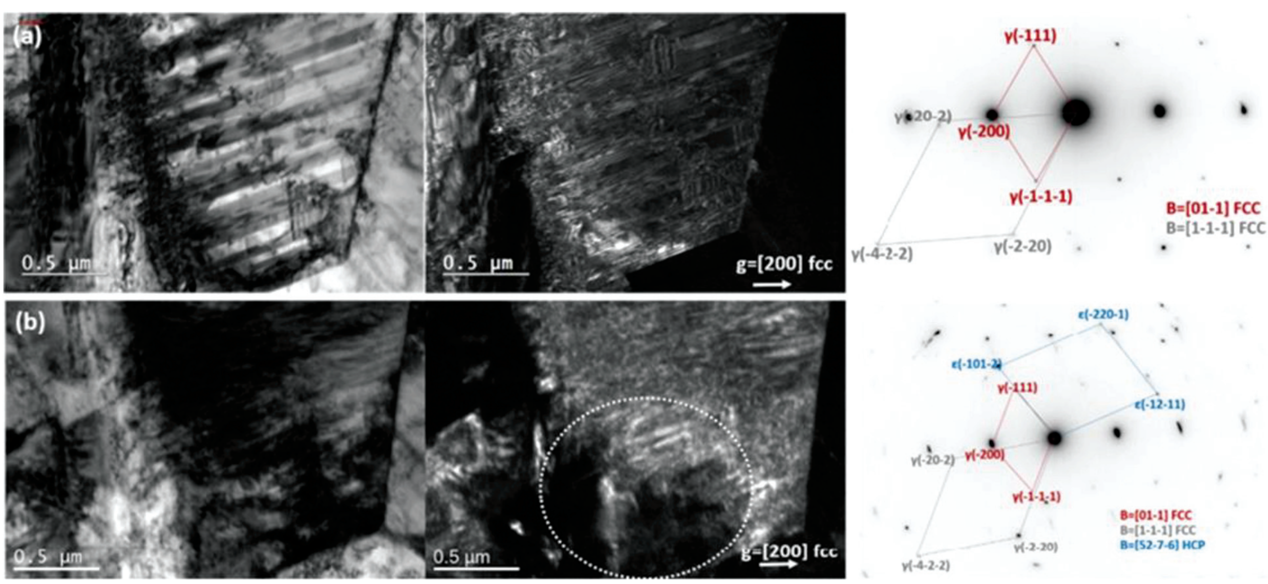

Figure 9: In-situ TEM observation of a phase transformation process[60]

This phase transformation not only enhances solid solution strengthening but also accommodates a significant amount of plastic strain through lattice reorganization, achieving a simultaneous increase in strength and ductility. Research has confirmed that the stacking fault energy (SFE) is a key factor influencing phase transformation. An increase in the SFE leads to a change in the transformation pathway from $\gamma \rightarrow \varepsilon \rightarrow \alpha'$ to $\gamma \rightarrow \gamma$ twin $\rightarrow \alpha'$, ultimately retaining only mechanical twinning. Additionally, the unique delayed DIMT phenomenon observed in SLMed 316L stainless steel is attributed to the high dislocation density impeding the motion of partial dislocations, which delays the transformation of austenite to martensite and significantly enhances the material's ductility. The in-situ TEM technique overcame the limitations of traditional observation methods, capturing the dynamic phase transformation process in real time at the atomic scale. This provides crucial evidence for understanding the plastic mechanisms associated with phase transformations. These findings not only deepen the understanding of the coordinated plastic deformation mechanisms during phase transformations but also highlight the unique advantages of the in-situ TEM technology in analyzing the phase behavior of advanced metals.

5 CONCLUSION

In-situ transmission electron microscopy (TEM) tensile experiments provide a key method for revealing the microscopic plastic deformation mechanisms of SLMed austenitic stainless steel. During this process, dislocation slip, twin growth, and phase transformation exhibit important coordinated interactions. The SLM technology, which fabricates three-dimensional metal components through ultra-fast cooling rates, results in unique heterogeneous structures and cellular substructures that endow the material with excellent mechanical properties. In-situ TEM experiments can capture the dynamic deformation processes of the material in real-time under stress-strain conditions, providing direct evidence for understanding its intrinsic mechanisms.

Dislocation slip is one of the important mechanisms of plastic deformation in SLMed Austenitic stainless steel. Unlike traditional forged materials, in SLMed materials, dislocation cell walls replace grain boundaries as the primary source of dislocations. Dislocations nucleate at the cell walls and propagate into adjacent cellular substructures. While dislocation movement is hindered by the cell walls, it does not completely stop; instead, it occurs through "jumping" slip, allowing for cross-cell transmission. This constrained dislocation movement enhances the strength of the material while maintaining its ductility.

Twin growth also has a significant impact on the plastic deformation of SLMed austenitic stainless steel. During the SLM process, nitrogen atoms dissolve into the material, significantly lowering the stacking fault energy (SFE) and promoting the formation of deformation twins. These twins nucleate from grain boundaries or dislocation cell walls and exhibit a layer-by-layer growth pattern at the nanoscale. The presence of nanoscale twins increases the lattice friction stress, hindering dislocation movement, and facilitates continuous strain hardening of the material during the deformation process through the dynamic Hall-Petch effect.

Phase transformation also plays an important role in the plastic deformation of SLMed austenitic stainless steel. Under low-temperature conditions, the plastic deformation mechanism shifts from being dominated by dislocation slip to being primarily governed by martensitic phase transformation. Deformation-induced martensitic transformation (DIMT) occurs via the paths $\gamma \rightarrow \varepsilon \rightarrow \alpha'$ or $\gamma \rightarrow \alpha'$. This phase transformation not only enhances solid solution strengthening but also accommodates a large amount of plastic strain through lattice reorganization, leading to a simultaneous increase in strength and ductility. In-situ TEM experiments visually reveal this dynamic phase transformation process, providing crucial evidence for understanding the plastic mechanisms associated with phase transitions.

In summary, dislocation slip, twin growth, and phase transformation demonstrate important coordinated interactions in the plastic deformation of SLMed austenitic stainless steel. In-situ TEM experiments capture the dynamic deformation processes of the material in real-time under stress-strain conditions, providing direct evidence to reveal these plastic deformation mechanisms. This understanding is significant for optimizing and controlling the mechanical properties of SLM austenitic stainless steel.

6 PROSPECT

The application of austenitic stainless steel in an SLM process showed significant potential, but further research and optimization are still needed in many aspects. Firstly, the relationship between microstructural evolution and material properties requires deeper exploration to reveal the fundamental reasons behind its excellent mechanical properties and plastic deformation behavior. Secondly, although there has been some progress in optimizing SLM process parameters, how to apply these parameters in industrial production remains a pressing issue to be addressed.[63,64] Furthermore, residual stress control and deformation prediction still need further improvement, especially for complex structures and large components. A combined approach of numerical simulation and experimentation is expected to find more effective stress relief techniques and deformation control strategies. Finally, exploring the application of more high-performance materials in SLM will greatly enrich the application fields of the SLM technology.[65,66]

Conflict of interest

Conflict of interest: Authors declare that there is no conflict of interest.

Acknowledgement

The authors express their gratitude to all members of the laboratory for their help. This research was supported by the National Natural Science Foundation of China (grant number 52175353), Shanxi Provincial Key R&D Program (202102150401002), Shanxi Provincial Basic Research Program Joint Funding Project (202403011211004), and Shanxi Province Scientific and Technological Achievements Transformation and Guidance Project (202404021301039).

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